TRANSACTIONS

of the

American Foundrymen's Society

Proceedings of the

Fifty - Fifth Annual Meeting

Buffalo, N. Y.

April 23 - 26, 1951

VOLUME 59

Published by the American Foundrymen's Society Chicago, Illinois 1951

COPYRIGHT 1951

by the

AMERICAN FOUNDRYMEN'S SOCIETY (INCORPORATED)

PRINTED IN THE UNITED STATES OF AMERICA

The American Foundrymen's Society as a body is not responsible for the statements and opinions advanced in this publication. Nothing contained in any publication of the American Foundrymen's Society is to be construed as granting any right, by implication or otherwise, for manufacture, sale, or use in connection with any method, apparatus, or product covered by Patent Rights, nor as insuring anyone against liability for infringement of Patent Rights.

S I

S

TABLE OF CONTENTS

										Page
Summary of Proceedings of the 55th Annual Meeting .	. 0.	, · · · · · · ·	-(-")		18), 5					vi
President's Annual Address										xiii
Report of the Treasurer			. ,							xv
Financial Statements										xvi
Report of the Secretary										xvii
Membership Reports				٠.					xix	k, XX
Report of the Technical Director										xxi
Minutes of Board Meetings		n 2 11		Ea wa.						xxiv
Eighth Chapter Officers Conference									2	XXXV
Management of Industrial Research										1
Some Thermal Considerations in Foundry Work										7
Basic Cupola Melting and Its Possibilities										20
Modification of Aluminum Alloys										28
An Investigation of the Penetration of Steel into Molding S										35
Mechanical Equipment for Medium-Size Gray Iron Foundry										56
Melt Quality and Fracture Characteristics of 85-5-5- Red Bras	s and 8	8-8-4	Bronz	е.						67
Chill Tests and the Metallurgy of Gray Iron										79
Fatigue Data Summary, Report No. 2										92
Scab Defect on Gray Iron Castings					•				•	99
Metal Penetration				•	•		٠		•	108
Sieve Analyses of Silica Sands					•		•	•	•	117
Oxidation-Reduction Principles Controlling the Composition				Irons				•		121
Melting Aluminum and Magnesium-Base Alloys							•	•	٠	139
Radiography as an Assistant to Foundry Practice							•		•	151
Recent Advances in Dielectric Core Baking					•		٠	•		159
The Contribution of Riser and Casting End Effects to Sound							۰	•	٠	171
										181
Kinetics of Graphitization in Cast Iron							*		*	
Ill Effects and Usefulness of Gases in Metallurgy										201
A New Method of Evaluating Costs in Jobbing Foundries .							*	•		210
Design of Core Boxes and Driers for Use on Core Blowing M								*	•	216
Mechanical Properties of Cast Steel as Influenced by Mass and	d Segre	gation	1 .						•	223
The Relative Effects of Chromium and Silicon Contents on R				lack-l	lear	Mal	leable	e Iro	n	244
Determination of Metal Penetration in Sand Molds								*	•	253
Effect of Vanadium on the Properties of Cast Chromium Moly								•		261
Core Oil Evaluation Method										276
Improvement of Machinability in High-Phosphorus Gray Cast	Iron,	Part I	Ι,		*					282
Compaction Studies of Molding Sands										291
Manufacture of Bronze Boiler Drop Plugs				•					0	296
Core Practice as Related to Malleable Foundry Losses										300
Dimensional Checking and Pressure Testing of Gray Iron	Castin	gs		•						304
Choosing Equipment for Nondestructive Testing				•			•		•	309
Precoat Materials for Investment Casting										315
Importance of Slag Control in Basic Cupola Operation .										323
Development of Foundry Courses in High and Trade School										332
Influence of Silicon Content on Mechanical and High-Temp	peratur	e Prop	perties	of N	odul	ar Ca	st Ir	on		337
										348
Structure and Mechanical Properties of a Mo-Ni-Cr Cast Iron										353
Effects of Certain Elements on Grain Size of Cast Copper-B								*		360
"Spec" Sheets and Their Various Uses										372
Industry Participation in Secondary School Training Progr	ams for	r the	Found	dry						376
Heat Flow in Moist Sand										381
Increasing Riser Efficiency										392
Die Casting Magnesium Alloys										398
Silicon-Chromium Alloy in Complicated Iron Castings .										401
Modern Core Sand Brastice										419

TABLE OF CONTENTS

(Continued from preceding page)

							Pa
Equipment and Methods of Straightening and I	Dimensional	Inspection of M	Malleable I	Iron Casti	ings		41
Solidification of Gray Iron in Sand Molds							42
Solidification of Steel Against Sand and Chill V	Walls						43
Testing of Sand Under Impact							
Absorbing the Technical Trainee							
Refining Secondary Copper Alloys							
Some Effects of Temperature and Melting Varial	bles on Chen	nical Composition	on and Stri	ucture of	Gray	Irons	4
Freezing of White Cast Iron in Green Sand Mole							
Malleable Cast Iron-Annealing Furnaces and	Atmosphere	e					4
Effect of Phosphorus Content on Mechanical P	roperties of	a Nodular Cast	Iron .				5
Isothermal Transformation Characteristics on	Direct Coo	oling of Alloyed	White Iro	n			5
Effect of Sand Grain Distribution on Casting F							
What the College Graduate Can Do For the A							
A Castability and Property Comparison of Seve	eral Magnesi	ium-Rare Earth	Sand Casti	ing Alloys	š .		5.
Ladle Refractories and Practice in Acid Electr							5
Subject Index							5.
Author Index							

(Fiscal Year Ending June 30, 1952)

- *President, WALTER L. SEELBACH, Superior Foundry, Inc., Cleveland.
- *Vice-President, I. R. WAGNER, Electric Steel Casting Co., Indianapolis

Board of Directors

(In Addition to the President and Vice-President)

Term Expires 1952

- T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa.
- L. C. FARQUHAR, SR., American Steel Foundries, East St. Louis, Ill.
- V. J. SEDLON, Master Pattern Co., Cleveland
- *F. G. Sefing, International Nickel Co., New York
- L. D. WRIGHT, U. S. Radiator Co., Geneva, N. Y.

Term Expires 1953

- J. J. McFadyen, Galt Malleable Iron Co., Galt, Ont., Canada.
- J. O. OSTERGREN, Lakey Foundry & Machine Co., Muskegon, Mich.
- *Frank W. Shipley, Caterpillar Tractor Co., Peoria, Ill.
- *James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.
- E. C. TROY, Palmyra, N. J.

ge

18 25

51

62 65

72 88

94 01 09

23

29 32

46 55

60

Term Expires 1954

HARRY W. DIETERT, Harry W. Dietert Co., Detroit

A. L. Hunt, National Bearing Div., American Brake Shoe Co., St. Louis.

JAMES T. MACKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

MARTIN J. O'BRIEN, JR., Symington-Gould Corp., Depew, N. Y.

A. M. ONDREYCO, Vulcan Foundry Co., Oakland, Calif.

*WALTON L. WOODY, National Malleable & Steel Castings Co., Cleveland

Staff Officers

Headquarters, 616 S. Michigan Ave., Chicago 5, Ill.

Secretary-Treasurer, WM. W. MALONEY

Technical Director, S. C. MASSARI

^{*} Member Executive Committee

SUMMARY OF PROCEEDINGS OF THE 55th ANNUAL MEETING

THE 55TH ANNUAL CONVENTION of the American Foundrymen's Society was held in Buffalo, April 23-26, 1951 without exhibits. There were over 3,000 foundrymen in attendance.

Credit for the excellent reception and hospitality received by those attending is given to the Western New York Chapter of A.F.S. under the chairmanship of Alfred A. Diebold, Atlas Steel Casting Co. This committee excelled in its splendid work and hospitality as host to the thousands of members and guests attending the Convention.

During the 4-day Convention 30 technical sessions were held with 70 papers being presented. In addition, there were three Gray Iron Shop Course sessions, three Sand Shop Course sessions and two Brass and Bronze Sand Course sessions. There was a Symposium on Principles of Gating sponsored jointly by Aluminum and Magnesium, Brass and Bronze, Gray Iron, Malleable and Steel Divisions. The papers and discussions presented during this symposium are published separately. A Round-Table Luncheon was sponsored by each Division. The Charles Edgar Hoyt Lecture was given by James C. Zeder, Chrysler Corp., who spoke on "Management of Industrial Research."

A summary of the sessions held follows:

ALUMINUM AND MAGNESIUM SESSION

Monday, April 23, 10:00 A.M.

Presiding—J. J. Warga, Piasecki Helicopter Corp., Morton, Pa. Co-Chairman—P. J. Scherbner, Sperry Gyroscope Corp., Great Neck, Long Island, N. Y.

Melting Aluminum and Magnesium-Base Alloys, L. W. Eastwood, Battelle Memorial Institute, Columbus.

Fluid Mechanics Applied to Founding, W. O. Wetmore and D. S. Richins, U.S. Naval Ordnance Test Station, Pasadena, Calif.

Brass and Bronze Session

Monday, April 23, 10:00 A.M.

Presiding-W. B. Scott, American Brake Shoe Co., National Bearing Division, Meadville, Pa.

Co-Chairman—H. G. Schwab, Bunting Brass & Bronze Company, Toledo, Ohio.

Refining Secondary Copper Alloys, Marvin Glassenberg and L. F. Mondolfo, Illinois Institute of Technology, Chicago, and

A. H. Hesee, R. Lavin & Sons, Inc., Chicago.

Radiography as an Assistant to Foundry Practice, S. A. Brosky,
Pittsburgh Testing Laboratory, and C. B. Johnson, Rockwell

Mfg. Company, Pittsburgh.

Increasing Riser Efficiency, J. O'Keeffe, Jr., Exomet, Inc., Conneaut, Ohio.

HEAT TRANSFER SESSION

Monday, April 23, 10:00 A.M.

Presiding-H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman—E. C. Troy, Foundry Engineer, Palmyra, N. J. Freezing of White Cast Iron in Green Sand Molds, H. A. Schwartz and W. K. Bock, National Malleable & Steel Castings Co. Cleveland.

Heat Flow in Moist Sands, Progress Report, Heat Transfer Research, V. Paschkis, Columbia University, New York.

MALLEABLE SESSION

Monday, April 23, 10:00 A.M.

Presiding-E. C. Zirzow, Deere & Co., Moline, Ill. Co-Chairman-F. J. Pfarr, Lake City Malleable Co., Cleveland. Malleable Core Practice as Related to Foundry Losses, E. J.

Jory, National Malleable & Steel Castings Co., Cicero, Ill. *Modern Core Sand Practice*, R. H. Greenlee, Auto Specialties Mfg. Co., St. Joseph, Mich.

ALUMINUM AND MAGNESIUM ROUND TABLE LUNCHEON Monday, April 23, 12:00 Noon

Presiding—R. F. Cramer, General Electric Co., Schenectady. Co-Chairmen—J. W. Meier, Dept. of Mines and Technical Surveys, Ottawa, Canada.

O. Z. Rylski, Dept. of Mines and Technical Surveys, Ottawa, Canada.

Subject-Centrifugal Casting.

BRASS AND BRONZE ROUND TABLE LUNCHEON

Monday, April 23, 12:00 Noon

Presiding-B. A. Miller, Baldwin-Lima-Hamilton Corp., Eddystone Division, Philadelphia.

Co-Chairman-H. L. Smith, Federated Metals Div., American Smelting & Refining Co., Pittsburgh.

Subject-Problems in Brass and Bronze Foundries.

EDUCATIONAL SESSION

Monday, April 23, 2:00 P.M.

Presiding-W. H. Ruten, Polytechnic Institute of Brooklyn, N. Y.

Co-Chairman-B. D. Claffey, Acme Aluminum Alloys, Inc., Dayton.

Development of Foundry Courses in High and Trade Schools, R. W. Schroeder, University of Illinois, Chicago.

Industry Participation in Secondary School Training Programs for the Foundry, T. W. Russell, Jr., American Brake Shoe Co., New York.

HEAT TRANSFER SESSION

Monday, April 23, 2:00 P.M.

Presiding-H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman-E. C. Troy, Foundry Engineer, Palmyra, N. J. Solidification of Steel Against Sand and Chill Walls, H. F. Bishep, F. A. Brandt and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

Solidification of Gray Iron in Sand Molds," R. P. Dunphy and W. S. Pellini, Naval Research Laboratory, Washington D. C.

MALLEABLE SESSION

Monday, April 23, 2:00 P.M.

Presiding—C. F. Lauenstein, Link-Belt Co., Indianapolis, Ind. Co-Chairman—Fitz Coghlin, Jr., Consultant, Albion, Mich.

Relative Effects of Chromium and Silicon Contents on Rate of Anneal of Black-Heart Malleable Iron," J. E. Rehder, Department of Mines and Technical Surveys, Ottawa, Canada.

Part I-Oxidation-Reduction Principles Controlling the Composition of Molten Cast Irons, R. W. Heine, University of Wisconsin. Madison.

ALUMINUM AND MAGNESIUM SESSION

Monday, April 23, 4:00 P.M.

Presiding-M. E. Brooks, The Dow Chemical Co., Bay City,

Co-Chairman-W. D. Danks, Howard Foundry Co., Chicago: Magnesium as a Die Casting Material, R. C. Cornell, Litemetal Diecast, Inc., Jackson, Mich.

Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys, K. E. Nelson and F. P. Strieter, The Dow Chemical Co., Midland, Mich.

^{*} Published as part of Symposium on Principles of Gating.

nd.

J.

ies

ON

cal

va.

ly-

an

n,

C.,

ls.

ns

id

d.

al

Modification Technique of Aluminum-Silicon Alloys, Official Exchange Paper from The Institute of Australian Foundrymen by R. H. Dyke, Defence Research Laboratories, Marybirnong, Victoria, presented by W. E. Sicha, Aluminum Co. of America,

Brass and Bronze Session

Monday, April 23, 4:00 P.M.
Presiding-B. N. Ames, New York Naval Shipyard, Brooklyn.
Co-Chairman-J. W. Bolton, Lunkenheimer Co., Cincinnati. U. S. Navy Non-Ferrous Development Program, C. L. Frear, Bureau of Ships, Department of Navy, Washington, D. C.

Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze, A.F.S. Research Report, R. D. Shelleng, C. Upthegrove and F. B. Rote, University of Michigan, Ann Arbor, Mich.

MALLEABLE SESSION

Monday, April 23, 4:00 P.M.

Fresiding-W. D. McMillan, International Harvester Co., Chi-

Co-Chairman-Wm. Zeunik, National Malleable & Steel Castings Co., Indianapolis, Ind.

Part II-Oxidation-Reduction Principles Controlling the Composition of Molten Cast Irons, R. W. Heine, University of Wis-

consin. Madison. Malleable Cast Iron Annealing Furnaces and Atmospheres, O. E. Cullen and R. J. Light, Surface Combustion Corp., Toledo,

EDUCATIONAL DINNER

Monday, April 23, 7:00 P.M.

Presiding-G. J. Barker, University of Wisconsin, Madison. Co-Chairman-G. K. Dreher, Foundry Educational Foundation, Cleveland.

Speakers-C. J. Freund, University of Detroit, Detroit. What the College Graduate Can Do for the American Foundryman, and C. L. Carter, Albion Malleable Iron Co., Albion, Mich. Absorbing the Technical Trainee in Industry.

> Non-Ferrous Foundry Industry Dinner (Sponsored by Non-Ferrous Founders' Society) Monday, April 23, 7:00 P.M.

Brass and Bronze Sand Course

Monday, April 23, 8:00 P.M.

Presiding-W. W. Edens, Badger Brass & Aluminum Foundry Company, Milwaukee.

Co-Chairman-G. P Halliwell, H. Kramer & Co., Chicago. Subject: Basic Approach to Sand Control for Copper-Base Alloys, Part I-Structural, Green, Air Set and Dry Properties.

Discussion Leader: H. W. Dietert, Harry W. Dietert Co., De-

GRAY IRON SHOP COURSE

Monday, April 23, 8:00 P.M.

Presiding-W. W. Levi, Lynchburg Foundry Co., Radford, Va. Co-Chairman-E. J. Burke, Hanna Furnace Company, Buffalo. Subject-Air in the Cupola.

Discussion Leader-D. E. Krause, Gray Iron Research Institute, Inc., Columbus, Ohio.

SAND SHOP COURSE

Monday, April 23, 8:00 P.M.

Presiding-F. S. Brewster, Harry W. Dietert Co., Detroit. Subject-Malleable Foundry Sand Control.

Discussion Leader-E. E. Woodliff, Foundry Sand Service Engineering Co., Detroit.

SYMPOSIUM ON PRINCIPLES OF GATING*

(Jointly sponsored by Aluminum and Magnesium, Brass and Bronze, Gray Iron, Malleable and Steel Divisions) Tuesday, April 24, 9:00 A.M.

Presiding-R. F. Thomson, Research Laboratories Division, General Motors Corp., Detroit.

Co-Chairmen-Aluminum and Magnesium Division-W. E. Sicha, Aluminum Co. of America, Cleveland.

Steel Division-W. S. Pellini, Naval Research Laboratory, Washington, D. C.

Brass and Bronze Division-G. P. Halliwell, H. Kramer & Co.,

Gray Iron Division-F. J. Walls, International Nickel Co., De-

Malleable Division-R. P. Schauss, Illinois Clay Products Co., Chicago

Introduction-Purpose of Symposium, R. F. Thomson. (Aluminum and Magnesium Division), Tentative Design of Horizontal Gating Systems for Light Alloys, L. W. Eastwood, Battelle Memorial Institute, Columbus, Ohio.

(Steel Division), Relation of Casting Quality to Gating Practice, W. H. Johnson, H. F. Bishop and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

(Brass and Bronze Division), Considerations in the Feeding of Castings, A. K. Higgins, Allis-Chalmers Mfg. Co., Milwaukee.

(Gray Iron Division), Interim Report on Gray Iron Risering Research, H. F. Taylor and W. A. Schmidt, Massachusetts Institute of Technology, Cambridge, Mass.

MALLEABLE ROUND TABLE LUNCHEON

Tuesday, April 24, 12:00 Noon

Presiding-J. H. Lansing, Malleable Founders' Society, Cleve-

Co-Chairman-R. J. Anderson, Belle City Malleable Iron Co., Racine, Wis.

Subjects-Panel Discussions-Malleable Foundry Refractories, Experimental Work on Core and Mold Materials."

Discussion Leader-Wm. Ferrell, Auto Specialties Mfg. Co., St. Joseph, Mich.

FOUNDRY EQUIPMENT AND SUPPLIES INDUSTRY LUNCHEON

(Sponsored by Foundry Equipment Manufacturers Assn. and Foundry Facings Manufacturers Association)

Tuesday, April 24, 12:00 Noon

Presiding-C. V. Nass, Chicago, President, Foundry Equipment Manufacturers Association.

Co-Chairman-F. B. Flynn, Chicago, President, Foundry Facings Manufacturers Association.

Speakers-A. J. Grindle, Director, Foundry Equipment and Supplies Section, Machinery Div., N.P.A., Washington, D. C.

Thos. Kaveny, Jr., Herman Pneumatic Machine Co., Pittsburgh, Pa., Industry Consultant to Foundry Equipment and Supplies Section, Machinery Div., N.P.A., Washington, D. C.

SYMPOSIUM ON PRINCIPLES OF GATING

(Jointly sponsored by Aluminum and Magnesium, Brass and Bronze, Gray Iron, Malleable and Steel Divisions)

Tuesday, April 24, 2:00 P.M.

Presiding-R. F. Thomson, Research Laboratories Division, General Motors Corporation, Detroit.

Co-Chairmen-Aluminum and Magnesium Division-W. E. Sicha, Aluminum Co. of America, Cleveland.

Steel Division-W. S. Pellini, Naval Research Laboratory, Washington, D. C.

Brass and Bronze Division-G. P. Halliwell, H. Kramer & Co.,

Gray Iron Division-F. J. Walls, International Nickel Co., De-

troit. Malleable Division-R. P. Schauss, Illinois Clay Products Co.,

Chicago

(Aluminum and Magnesium Division), Discussion by T. E. Kramer, Wm. F. Jobbins, Inc., Aurora, Ill.

(Steel Division), Discussion by J. B. Caine, Foundry Consultant, Wyoming Ohio, and J. H. Lowe, Wehr Steel Co., Milwaukee.

(Brass and Bronze Division), Discussion by B. N. Ames, N. Y. Naval Shipyard, Brooklyn, and W. W. Edens, Badger Brass and Aluminum Foundry Co., Milwaukee.

(Gray Iron Division), The Why of Gate and Feeder Design, F. G. Sefing, International Nickel Co., New York. Discussion by H. H. Kessler, Sorbo-Mat Process Engineers, St. Louis, Mo.

(Malleable Division), Trends in Malleable Gating and Risering, E. C. Zirzow, Deere & Co., Moline, Ill.

Developments in Gating of Small to Medium Malleable Castings, C. C. Lawson, Wagner Malleable Iron Co., Decatur, Ill.

^{*} Published separately as Symposium on Principles of Gating.

PATTERN SESSION

Tuesday, April 24, 4:00 P.M.

Presiding-Harry Lees, Whitin Machine Co., Whitinsville, Mass

Co-Chairman-V. C. Reid, City Pattern Foundry & Machine Co., Detroit.

Design of Core Boxes and Driers for Core Blowing, E. Blake, The Osborn Manufacturing Co., Cleveland.

Design of Core Setting and Rubbing Fixtures, F. J. Oklessen, Motor Pattern Co., Cleveland.

SAND SESSION

Tuesday, April 24, 4:00 P.M.

Presiding-J. B. Caine, Foundry Consultant, Wyoming, Ohio. Co-Chairman-R. H. Jacoby, The Key Company, E. St. Louis,

Compaction Studies of Molding Sands, R. E. Grim and W. D. Johns, Jr., State Geological Survey, University of Illinois, Ur-

Effect of Sand Grain Distribution on Casting Finish, H. H. Fairfield, and James MacConachie, Wm. Kennedy & Sons Ltd., Owen Sound, Ont., Canada.

TIMESTUDY AND METHODS SESSION

Tuesday, April 24, 4:00 P.M.

Presiding-M. T. Sell, Sterling Foundry, Wellington, Ohio. Co-Chairman-H. R. Williams, Williams Management Engineering, Milwaukee.

Fatigue Data Summary-Report No. II, M. E. Annich, American Brake Shoe Co., Mahwah, N. J.

CANADIAN DINNER

Tuesday, April 24, 7:00 P.M.

Presiding-J. J. McFadyen, Galt Malleable Iron Co., Galt, Ont., Canada.

Brass and Bronze Sand Course

Tuesday, April 24, 8:00 P.M.

Presiding-J. R. Crain, American Brake Shoe Co., National Bearing Division, Meadville, Pa.

Co-Chairman-G. P. Halliwell, H. Kramer & Co., Chicago. Subject-Basic Approach to Sand Control for Copper-Base Alloys-Part II, Hot and Retained Properties.

Discussion Leader-H. W. Dietert, Harry W. Dietert Co., De-

GRAY IRON SHOP COURSE

Tuesday, April 24, 8:00 P.M.

Presiding-H. H. Wilder, Vanadium Corp. of America, Detroit. Co-Chairman-E. J. Burke, Hanna Furnace Co., Buffalo. Subject-Melting Gray Iron in a Reverberatory-Type Furnace. Discussion Leader-J. G. Winget, Reda Pump Co., Bartlesville, Oklahoma.

SAND SHOP COURSE

Tuesday, April 24, 8:00 P.M.

Presiding-R. H. Olmsted, Whitehead Bros. Co., New York. Co-Chairman-W. D. Dunn, Oberdorfer Foundries, Syracuse,

Subject-Shell Molding and Use of Resin Binders.

Discussion Leaders-B. N. Ames, New York Naval Shipyard, Brooklyn, and W. C. Jeffery, University of Alabama, Birmingham, Ala.

BRASS AND BRONZE SESSION

Wednesday, April 25, 10:00 A.M.

Presiding-F. L. Riddell, H. Kramer & Co., Chicago.

Co-Chairman-H. G. Schwab, Bunting Brass & Bronze Co., Toledo, Ohio.

Effects of Certain Elements on Grain Size of Cast Copper-Base Alloys, R. A. Colton and M. Margolis, American Smelting & Refining Co., Barber, N. J.

Manufacture of Bronze Boiler Drop Plugs, B. F. Kline and R. Davidson, Southern Pacific Railroad Co., Sacramento, Calif., presented by G. P. Halliwell, H. Kramer & Co., Chicago.

FOUNDRY COST SESSION

Wednesday, April 25, 10:00 A.M.

Presiding-R. L. Lee, Grede Foundries, Inc., Milwaukee. Co-Chairman-G. Tisdale, Zenith Foundry Co., Milwaukee.

'Spec' Sheets and their Various Uses, John Taylor, Lester B. Knight and Associates, Inc., Chicago.

A New Method of Evaluating Costs in Jobbing Foundries, A. W. Schneble, Jr., Advance Foundry Co., Dayton, Ohio, and C. E. McQuiston, Ohio State University, Columbus.

GRAY IRON SESSION

Wednesday, April 25, 10:00 A.M.

Presiding-A. P. Gagnebin, International Nickel Co., New

Co-Chairmen-T. E. Eagan, Cooper-Bessemer Corp., Grove City, Pa.

W. B. McFerrin, Haynes Stellite Division, Kokomo, Indiana. Influence of Silicon Content on Mechanical and High-Temperature Properties on Nodular Cast Iron, W. H. White, Jackson Iron & Steel Co., Jackson, Ohio, L. P. Rice and A. R. Elsea, Battelle Memorial Institute, Columbus, Ohio.

Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron, J. E. Rehder, Department of Mines and Technical Surveys, Ottawa, Canada.

Some Effects of Temperature and Melting Variables on Chemical Composition and Structure of Gray Irons, E. A. Lange and R. W. Heine, University of Wisconsin, Madison.

SAND SESSION

Wednesday, April 25, 10:00 A.M.

Presiding-J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.

Metal Penetration-Report, Mold Surface Committee-S. L. Gertsman and A. E. Murton, Department of Mines and Technical Surveys, Ottawa, Canada.

Determination of Metal Penetration in Sand Molds, H. J.

Gonya and D. C. Ekey, Ohio State University, Columbus.

Effect of Fluid Metal Pressure on Penetration Defects, Progress Report-C. C. Sigerfoos, Michigan State College, East Lansing,

DEFENSE PRODUCTION LUNCHEON

Wednesday, April 25, 12:00 Noon

Presiding-W. L. Seelbach, Vice-President, American Foundrymen's Society.

Speakers: A. J. McDonald, Chief, Castings Section, Iron & Steel Division, N.P.A., Washington, D. C.

J. A. Claussen, Chief, Pig Iron Section, Iron & Steel Division, N.P.A., Washington, D. C.

Lt. Commander Wm. A. Meissner, Bureau of Ships, Washington, D. C., Liaison Officer to N.P.A.

Nigel Bell, Light Metals Division, N.P.A., Washington, D. C.

ANNUAL BUSINESS MEETING

Wednesday, April 25, 2:00 P.M.

Presiding-Walton L. Woody, A.F.S. President. President Woody called the meeting to order as the Annual Business Meeting of the American Foundrymen's Society. He

then presented the President's Annual Address. See page xiii. Following this presentation President Woody called on Secre-

tary-Treasurer Wm. W. Maloney who announced the 1951 A.F.S. Apprentice Contest winners as follows: Wood Patternmaking Division

1st-Chas. J. Hofschroer, Kay-Brunner Steel Products, Inc., Alhambra, Calif.

2nd-Eugene A. Busch, Worthington Pump & Machine Co., Buffalo.

3rd-Allen Roy Klein, Olney Foundry, Link Belt Co., Philadelphia.

Metal Patternmaking Division

1st-Marvin L. Lohnes, Caterpillar Tractor Co., Peoria, Ill. 2nd-Jos. E. Collins, Jr., Packard Motor Car Co., Detroit. 3rd-John Wm. Burkholder, Jr., Central Pattern Co., St. Louis. Gray Iron Molding Division

lst-Harold M. Carty, Caterpillar Tractor Co., Peoria, Ill. 2nd-Gerald Schrank, Nordberg Manufacturing Co., Mil-

3rd-Donald N. Marinelli, Fulton Foundry & Machine Co., Cleveland.

B.

nd

ve

h-

te,

R.

nd

ıi.

he

31

ii.

J.

SS

g

n,

Steel Molding Division
1st-John G. Polino, Dominion Engineering Works, Ltd., Montreal, Canada.

2nd-Adolph C. Zinn, Waukesha Foundry Co., Waukesha, Wis. 3rd-Norman G. Bublitz, Bucyrus-Erie Co., So. Milwaukee, Wis.

Nonferrous Molding Division

1st-Hilary J. Heck, Nordberg Manufacturing Co., Milwaukee. 2nd-Gilbert Dandoy, Montreal Bronze, Ltd., Montreal, Canada.

3rd-T. Grant Wilcox, Dominion Engineering Works, Ltd., Montreal, Canada.

The Society arranged to have the five first-prize winners present at the Convention and to receive their awards in person. The first-prize winners were called to the platform and President Woody presented each with the first prize, a check for \$100.00 and an engraved Certificate of Award prefaced by a few words of commendation and encouragement.

President Woody then called on Secretary-Treasurer Maloney who reported on the nominations of Officers and Directors for the coming year and stated that no additional nominees had been received in accordance with the procedure prescribed in Art. X of the Society By-Laws. He therefore cast the unanimous ballot of the membership of A.F.S. for the election of the

President (to serve one year):

W. L. Seelbach, Superior Foundry, Inc., Cleveland.

Vice-President (to serve one year):

I. R. Wagner, Electric Steel Casting Co., Indianapolis, Ind. Directors (to serve three years):

Harry W. Dietert, Harry W. Dietert Co., Detroit.

A. L. Hunt, National Bearing Div., American Brake Shoe Co.,

Dr. James T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.

Martin J. O'Brien, Jr., Symington-Gould Corp., Depew, N. Y. A. M. Ondreyco, Vulcan Foundry Co., Oakland, Calif.

Director (to serve one year):

Walton L. Woody, National Malleable & Steel Castings Co., Cleveland.

President Woody then introduced the Charles Edgar Hoyt Lecturer James C. Zeder, Chrysler Corp., Detroit.

CHARLES EDGAR HOYT ANNUAL LECTURE

Wednesday, April 25, 2:45 P.M.

Management of Industrial Research, James C. Zeder, Vice-President and Director of Engineering and Research, Chrysler Corporation, Detroit.

PLANT AND PLANT EQUIPMENT SESSION Wednesday, April 25, 4:00 P.M.

Presiding-Jas. Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman-H. W. Johnson, Wells Manufacturing Com-

pany, Skokie, Ill.

Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings, L. N. Schuman, National Malleable & Steel Castings Co., Cleveland.

Dimensional Checking and Pressure Testing of Gray Iron Castings, K. M. Smith, Caterpillar Tractor Co., Peoria, Ill., presented by F. W. Shipley.

REFRACTORIES SESSION

Wednesday, April 25, 4:00 P.M.
Presiding-R. H. Stone, Vesuvius Crucible Company, Swissvale, Pittsburgh.

Co-Chairmen-R. P. Schauss, Illinois Clay Products Co., and R. A. Witschey, A. P. Green Fire Brick Co., Chicago.

Ladle Refractories and Practice in Acid Electric Steel Foundry, Clyde Wyman, Burnside Steel Co., Chicago. (Followed by general discussion of ladle refractories and their application).

SAND SESSION

Wednesday, April 25, 4:00 P.M.

Presiding-E. C. Zirzow, Deere & Company, Moline, Ill. Co-Chairman-H. K. Salzberg, The Borden Co., Bainbridge, N. Y.

Core Oil Evaluation Method, A. E. Murton, Department of Mines and Technical Surveys, Ottawa, H. H. Fairfield, Wm.

Kennedy & Sons Ltd., Owen Sound, Ont., and B. Richardson, Steel Castings Institute of Canada, Ottawa.

Recent Advances in Dielectric Core Baking, J. W. Cable, Thermex Division, The Girdler Corp., Louisville, Ky.

A.F.S. ALUMNI DINNER Wednesday, April 25, 7:00 P.M.

PLANT AND PLANT EQUIPMENT SESSION

Wednesday, April 25, 8:00 P.M. Presiding-R. L. Wolf, Stone and Webster Engineering Corp., Boston.

Choosing Equipment for Nondestructive Testing, C. H. Hastings, Watertown Arsenal, Mass.

Motion Picture-"Mechanization in Molding," H. C. Weimer, Beardsley & Piper Div., Pettibone Mulliken Corp., Chicago.

GRAY IRON SHOP COURSE

Wednesday, April 25, 8:00 P.M.

Presiding-K. H. Priestley, Vassar Electroloy Products, Inc., Vassar. Mich.

Co-Chairman-E. J. Burke, Hanna Furnace Co., Buffalo. Subject-Metal Pouring Temperature Control.

Discussion Leader-R. A. Clark, Electro Metallurgical Div., Union Carbide & Carbon Co., Detroit.

SAND SHOP COURSE

Wednesday, April 25, 8:00 P.M. Presiding-R. H. Jacoby, The Key Co., East St. Louis, Ill. Co-Chairman-H. W. Meyer, General Steel Castings Corp., Granite City, Ill.

Subject-Foundry Sand Control.

Discussion Leader-C. B. Schureman, Baroid Sales Div., National Lead Co., Chicago.

GRAY IRON SESSION

Thursday, April 26, 10:00 A.M.

Presiding-Max Kuniansky, Lynchburg Foundry Co., Lynchburg. Va. Co-Chairman-Gosta Vennerholm, Ford Motor Co., Dearborn,

Mich.

Basic Cupola Melting and Its Possibilities, Official Exchange Paper from The Institute of British Foundrymen by E. S. Renshaw, Ford Motor Co. Ltd., Dagenham, England.

Importance of Slag Control in Basic Cupola Operation, R. A. Flinn and R. W. Kraft, American Brake Shoe Co., Mahwah, N. J.

STEEL SESSION

Thursday, April 26, 10:00 A.M.

Presiding-J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.

Co-Chairman-H. W. Dietert, Harry W. Dietert Co., Detroit. Precoat Materials for Investment Casting, W. F. Davenport and A. Strott, Wright-Patterson Air Force Base, Dayton, Ohio.

Contribution of Riser and Casting Edge Effects to Soundness of Cast Steel Plate, H. F. Bishop, E. T. Myskowski and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

An Investigation of the Penetration of Steel into Molding Sands, Official Exchange Paper from Metallografiska Institute by Holger Pettersson, Stockholm, Sweden, presented by J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.

NATIONAL CASTINGS COUNCIL Thursday, April 26, 10:00 A.M.

GRAY IRON ROUND TABLE LUNCHEON Thursday, April 26, 12:00 Noon

Presiding-A. O. Barczak, Superior Foundry, Inc., Cleveland. Co-Chairman-C. A. Harmon, Hanna Furnace Co., Buffalo. Subject-Gray Iron Melting with Materials Available.

Discussion Leaders-J. F. Dobbs, New York Air Brake Company, Watertown, N. Y., A. J. MacDonald, Hanna Furnace Co., Buffalo, and S. A. Kundrat, Homestead Valve Mfg. Co., Coraopolis, Pa.

STEEL ROUND TABLE LUNCHEON

Thursday, April 26, 12:00 Noon

Presiding-V. E. Zang, Unitcast Corporation, Toledo, Ohio. Co-Chairman-D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.

Subjects-Alloy Substitutions and Conversion, Increasing Casting and Total Yield, Sulphur, and Acid vs Basic Practice.

GRAY IRON SESSION

Thursday, April 26, 2:00 P.M.

Presiding-V. A. Crosby, Climax Molybdenum Co., Detroit. Co-Chairman-R. Schneidewind, University of Michigan, Ann Arbor.

Structure and Mechanical Properties of a Mo-Ni-Cr Cast Iron, E. A. Loria, The Carborundum Co., Niagara Falls, N. Y.

Kinetics of Graphitization in Cast Iron, B. F. Brown, Dept. of Engineering Research, North Carolina State College, Raleigh, N. C. and M. F. Hawkes, Dept. of Metallurgical Engineering, Metals Research Laboratory, Carnegie Institute of Technology, Pittsburgh.

Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron, F. B. Rote and G. A. Conger, University of Michigan, Ann Arbor, and K. A. DeLonge, International

Nickel Co., Bayonne, N. J.

STEEL SESSION

Thursday, April 26, 2:00 P.M.

Presiding-F. Kiper, Ohio Steel Foundry Co., Springfield, Ohio. Co-Chairman-G. W. Johnson, Vanadium Corp. of America,

Effect of Vanadium on Properties of Cast Chromium-Molybdenum Steels, N. A. Ziegler, W. L. Meinhart and J. R. Gold-

smith, Crane Co., Chicago.

Mechanical Properties of Cast Steel as Influenced by Mass and Segregation, J. F. Wallace, Watertown Arsenal, Mass., Lt. Col. J. H. Savage, Ordnance Corps, and H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.

Usefulness and Ill Effects of Gases in Metallurgy, E. Spire,

Canadian Liquid Air Co. Ltd., Montreal, Canada.

GRAY IRON SESSION

Thursday, April 26, 4:00 P.M.

Presiding-H. H. Wilder, Vanadium Corporation of America,

Co-Chairman-J. L. Brooks, Muskegon Piston Ring Co., Sparta,

Improvement of Machinability in High-Phosphorus Gray Cast Iron-Part II, W. W. Austin, Southern Research Institute, Birmingham, Ala.

Chill Tests and the Metallurgy of Gray Iron, D. E. Krause,

Gray Iron Research Institute, Columbus, Ohio.

Silicon-Chromium Alloy in Complicated Iron Castings, R. A. Clark, Electro Metallurgical Div., Union Carbide & Carbon Co., Detroit.

SAND SESSION

Thursday, April 26, 4:00 P.M.

Presiding-C. A. Sanders, American Colloid Co., Chicago. Co-Chairman-H. W. Dietert, Harry W. Dietert Co., Detroit. Silica Sands-Sieve Analyses, A. I. Krynitsky and F. W. Raring, National Bureau of Standards, Washington, D. C., presented by

Scab Defect on Gray Iron Castings, A Progress Report by A.F.S. Committee on Physical Properties of Iron Molding Materials at Elevated Temperatures.

Testing of Sand Under Impact, W. H. Moore, Meehanite Corp., Cleveland Heights, Ohio.

ANNUAL BANQUET

Thursday, April 26, 7:00 P.M.

Presiding-Walton L. Woody, President, American Foundry-

The Annual A.F.S. Dinner of the Society's 55th Annual Convention was called to order by President Walton L. Woody, pre-

Following the singing of the National Anthem by those in attendance, President Woody introduced L. C. Wilson, Chairman of the A.F.S. Board of Awards. Mr. Wilson introduced A. L. Boegehold who presented the John A. Penton Gold Medal Award to V. A. Crosby.

PRESENTATION OF JOHN A. PENTON GOLD MEDAL

Mr. Boegehold presented the award with these prefacing

"I have known and admired the 1951 John A. Penton Gold

Medal recipient for more than 20 years. V. A. Crosby's character is so exemplary that he convinced me that he has no vices or undesirable traits.

"For a quarter of a century he has steadfastly promoted a better understanding by the Foundry Industry of the science and technology of cast metals. A.F.S. convention programs back through the years show V. A. Crosby as chairman of a technical session, author of a paper or chairman of a committee. The 'Alloy Cast Iron Handbook,' the "Cast Metals Handbook' and the 'Cupola Operations Handbook' were all recipients of his guidance and contributions.

"Numerous local chapters of A.F.S. from Boston to Seattle and in Canada have repeatedly heard his lectures on many phases of cast iron. The Detroit Chapter chose him as Chairman in

1941 and 1942.

"Other technical societies receive the benefit of his and others' work in the A.F.S. through his membership on committees of the AIMME, ASM, ASTM and SAE. In the latter society he is the arbiter of Specifications for Automotive Castings by being chairman of that committee.

"For two years the A.F.S. Gray Iron Division Program and Papers Committee operated under his direction, and for the last

two years he has been Chairman of that division.

"The year of his birth, 1893, has been made famous by a considerable number of famous metallurgists. He was born under the sign of Zodiac which I am sure has a metallurgical aspect, so in spite of the fact that he graduated from 'Ole Miss' with a B.S. in Chemistry and had a scholarship to L.S.U. for post graduate work, he accepted a metallurgical position at the Dodge Bros. in Detroit.

Following he served in the USAF and then became Chief Metallurgist of Packard Motor Car Co. Next he took charge of melting at Studebaker Corp. and finally in 1934 he joined Climax Molybdenum Co. where he is now Metallurgical Engineer in charge of sales, service and development in the Detroit

"Because of my close association with Mr. Crosby, I have known for a long time that his work deserved public recognition by the Society so it was with great pleasure that I read the announcement that the Board of Awards has voted him the John A. Penton Gold Medal recipient for "Outstanding contributions to the dissemination of information to the Foundry Industry."

"It gives me great pleasure Victor Anderson Crosby to present to you on behalf of the American Foundrymen's Society the John A. Penton Gold Medal and a certificate for Honorary Life

Membership in this great organization."

Mr. Crosby's Acceptance of John A. Penton Award

S S A In

in

ho

Mr. Crosby accepted this award with the following remarks: "It is said in the Good Book that it is more blessed to give than to receive, but on this particular occasion I must admit that I am delighted to be the recipient rather than the donor. I can assure you that I was surprised and very happy when advised that I had been selected by the Board of Awards to be the recipient of the John A. Penton Gold Medal.

"I prefer to accept this honor-a recognition which comes to only a few people-not as a personal reward, but rather as an acknowledgment of the combined efforts of all the men who have worked with me over the many years. I refer particularly to the technical staff of the Climax Molybdenum Co., to the A.F.S. headquarters technical staff, and to the personnel of the

various technical committees.

"It is with deep gratitude and humility that I accept the John A. Penton Gold Medal, and I shall always strive to be worthy of the esteem and honor it signifies."

PRESENTATION OF JOHN H. WHITING GOLD MEDAL

The John H. Whiting Gold Medal Award was presented to Alfred Boyles by Dr. Clarence H. Lorig with these prefacing

"This year's recipient of the John H. Whiting Gold Medal is a young man who has devoted the major share of his professional career to work in the field of gray cast iron. He has attained national and international recognition for his efforts in advancing our knowledge of this important metal. Thus, through his achievements, his devotion to his work, and his loyalty to the industry, he has fulfilled, in good measure, all the requirements for the award.

"Alfred Boyles was born, raised, and educated in North Caro-

r

ce

ck

al

ne

nd

is

id

es

in

S

of

g

d

st

1-

h

si

e

ef

of

d

n

e

ŧ

e

D

ina. He majored in chemistry at the University of North Carolina and then obtained his early technical training in the steel industry, having been employed at Tennessee Coal, Iron, and Railroad Co. for several years as analytical chemist, test engineer, and metallographer. Here he instituted microscopical techniques to solve metallurgical problems that were ingenious in their concept. This ingenuity was again exemplified in his work at Battelle where he was employed as research engineer for a period of six years. It was then that he pioneered a study on the freezing of cast iron and the formation of graphite in gray cast iron. In 1941, he became associated with the United States Pipe and Foundry Co. at Burlington, N. J. where he holds the position of research metallurgist. In that capacity, he has responsibility for metallurgical developments in the field of cast iron. His phenomenal mechanical design aptitude, so successfully used in the solution of problems, has been a source of amazing interest to all who know and work with him.

"A few years ago he wrote the book 'Structure of Cast Iron,'

which is a classic among textbooks in metallurgy

"It is therefore, significant that the Board of Awards should have emphasized this devotion to cast iron in their citation which reads, Fundamental Studies on the Mechanism of Graphi-

tization of Gray Cast Iron.'

"Alfred Boyles' interests go beyond the foundry and cast iron. Since his college days, he has maintained an active interest in amateur astronomy and color photography. It was gratifying to learn that distinction came to him early through the recognition received for some of his early contributions in the field of astronomy.

"He serves on various technical committees in this Society and in ASTM."

With these remarks Dr. Lorig presented the award to Dr. Boyles.

ACCEPTANCE OF JOHN H. WHITING GOLD MEDAL

Dr. Boyles accepted this award from Dr. Lorig and expressed his appreciation for this singular recognition. He also expressed with deep humility appreciation to his associates in the industry.

PRESENTATION OF THE PETER L. SIMPSON GOLD MEDAL

The Peter L. Simpson Gold Medal award was made by R. A.

Flinn to T. W. Curry with these prefacing remarks:

"From the age of nine, T. W. Curry worked during summer vacations in the various departments of the Danville Iron and Steel Corp. In 1932 he was graduated from the Pennsylvania State College with a B.S. degree in Metallurgical Engineering. After graduation he resumed employment with the Danville Iron and Steel Corp. where he developed high-strength cast iron for jaw crusher frames.

"He was foundry superintendent for the Kennedy Van Saun Foundry for a short term. In 1943 he became associated with the Lynchburg Foundry Co. where he is now employed as Metal-

lurgist.

"He has contributed numerous technical papers on foundry practice to the literature and has made many talks at A.F.S.

Conventions, Chapters and Regional Conferences.

"I am happy to present to you, Thomas W. Curry, this Peter L. Simpson Gold Medal on behalf of the A.F.S. Board of Awards for 'Outstanding work and contributions in the field of sand technology in the Foundry Industry.'"

ACCEPTANCE OF THE PETER L. SIMPSON GOLD MEDAL

Thomas W. Curry accepted this award with the following remarks:

"To be the recipient of the Peter L. Simpson Gold Medal Award is one of the happiest experiences of my life. However, I approach this moment with a feeling of humility and deep appreciation to accept this honor.

"It is a priceless gift made possible for me only by the perseverence and hard work of my associates at Lynchburg Foundry Co. and the full and willing cooperation of the members of this

Society to share their store of foundry knowledge.

"I hope that I shall be able to pattern my work in the Foundry Industry and my associations with the members of this Society in such a manner that both may be in keeping with the faith and tradition of the Peter L. Simpson Award. In so doing, I hope that I shall be worthy of this honor which you have conferred upon me.

"I wish to express my sincere appreciation and thanks to the

Members of the Board of Awards, and the members of the American Foundrymen's Society,"

Presentation of Honorary Life Membership to E. W. Beach

An Honorary Life Membership in A.F.S. was presented to E. W. Beach by James Thomson. In the absence of Mr. Beach due to illness the certificate was received by G. W. Cannon on behalf of Mr. Beach. Mr. Thomson made these presentation remarks:

"In a Society like ours, composed of so many elements, bestowing recognition on a single individual for what he has done on behalf of the Foundry Industry becomes a most difficult task. I am sure that any Board of Awards is constantly faced with the choice of recognizing technical and practical accomplishment, or of acknowledging the grace of men who are outstanding in character.

The recipient of this award is a veteran foundryman who was apprenticed to the Waterbury-Farrell Foundry and Machine Co., Waterbury, Conn., his home town, in 1891. Thereafter he served as president, vice-president or general manager with a number of foundry concerns before joining Campbell, Wyant & Cannon Foundry Co. of Muskegon, Mich., in 1924. For almost a quarter of a century he was the Executive Engineer and Metallurgical Consultant for the C.W.C. organization and during this period made many notable contributions to the science of ferrous metallurgy. For example, he is credited with first applying automotive cylinder casting patterns to molding machine production, and the A.F.S. citation is largely concerned with this period.

"E. W. Beach is notable for other accomplishments, however. Among them is the deep respect in which he has long been held by his associates. He is a Charter Member of the Western Michigan Chapter, and he has served on the A.F.S. Plant and Plant Equipment Committee ever since it was first established. In fact, he has been that committee's guiding hand. Mr. Beach and I have worked together on this committee since 1934 and his advice and counsel have been of inestimable value to the committee due to his long and intimate acquaintance with the engineering phases of the Foundry Industry.

"He was active in revitalizing the Hackley Memorial Training School Foundry in Muskegon, now recognized as one of the outstanding foundry trade school enterprises. He has also long been

active in photography.

"Mr. Beach is entitled to recognition by this Society for the combination of these factors. However, I think all who know him will think longer of his never-failing courtesy, his kindliness under any circumstances, his cool head under extreme provocation, and above all his respect for the other fellow. This is the kind of man that makes a Society like ours grow as a living thing and not just a business organization.

"It is indeed unfortunate that Mr. Beach could not receive this Honorary Life Membership Certificate in person due to persistent ill health since his retirement from active business two years ago. On behalf of the American Foundrymen's Society, however, I present this award to Edward Wells Beach through his long time business associate at C.W.C., George W. Cannon."

ACCEPTANCE OF E. W. BEACH

Mr. Cannon read this telegraphed statement from Mr. Beach: "My progress since release from four months hospitalization does not warrant my undertaking the journey to Buffalo. That I cannot be present to receive this honor which has been awarded me by the American Foundrymen's Society in person is a disappointment beyond measure.

"Membership in the Society has been deeply treasured through a long period of years, a span which has seen the development of an associated group of men intent on advancing their own individual skill in their chosen industry by means of mutual exchange of ideas for improvement in production methods and quality, to a growth that has taken it from the status of a minor association to that of a major engineering society of the world. That I am to receive individual recognition for the small part I have contributed to its work, is an honor indeed.

"That the Awards Committee has chosen for its presentation my friend and coworker of many years in committee duties, A.F.S. Director James Thomson, and to accept it for me my former employer of 25 years, George W. Cannon, are indeed added honors. My appreciative thanks to the Awards Committee and the Society for the award and token of an Honorar Life Membership."

Fo

the

M

TI

fir

for

to

tes

the

vit act ever bin ab from Forch to

rie bee

abl

PRESENTATION OF HONORARY LIFE MEMBERSHIP TO E. J. McAfee

A.F.S. Director V. J. Sedlon presented an Honorary Life Membership to Edward J. McAfee with these prefacing remarks:

"For one who has been a patternmaker all his life and is engaged in the business of making patterns for foundrymen, this moment is indeed a happy one. It is my honor and privilege to represent A.F.S. in its honoring of a fellow patternmaker.

"Mr. McAfee has been a patternmaker since 1907. He was born and raised in the great Pacific Northwest of Oregon and Washington and has been connected with the Puget Sound Naval Shipyard since 1916 first as a patternmaker, then as a supervisor, and as a Master Patternmaker since 1938. He is a patternmaker who projects his thinking beyond today into the possibilities of tomorrow.

"Typical of his willingness to impart his knowledge to others, in the true spirit of A.F.S., Mr. McAfee made numerous talks during the past five years before A.F.S. Chapters, schools, and various organizations, particularly on the development and use of plastics as patternmaking materials. He has authored a number of articles in the American Foundryman and is co-author of 'The Pattern Maker's Manual,' issued by the Puget Sound Naval Ship Yard. He has served not only on various committees of the A.F.S. Pattern Division, but he is now engaged in writing for A.F.S. a new patternmaking manual.

"In addition to his many contributions to the technology of foundry patternmaking, he has for many years been prominent in encouraging young men to enter the foundry industry. He has carried on this work through the A.F.S. Apprentice Contests and as a Director of the Washington Chapter of A.F.S. Through this work, I have come to know him well and I feel that this recognition is one of the most deserved in many years.

"It is my pleasure to present Honorary Life Membership in the American Foundrymen's Society to Edward J. McAfee 'for outstanding service to the Society and to the Foundry Industry in the development of patternmaking materials and applications' and to offer him this certificate of Life Membership as a token of our respect and esteem."

ACCEPTANCE BY E. J. MCAFEE

Mr. McAfee graciously accepted this award with these remarks:

"It is needless to say that I feel highly honored and grateful to be a recipient of this Honorary Life Membership in the American Foundrymens' Society. The notification of my selection came as a complete surprise and I feel highly indebted to the Society as the efforts I have put forth on its behalf fall far short toward making me eligible to receive this honor. However, it will be an incentive to me in my future endeavors to somewhat justify the confidence placed in me.

"I wish to thank the Board of Awards and the Officers of the American Foundrymen's Society for this honor."

Presentation of Honorary Life Membership to W. L. Woody

An Honorary Life Membership certificate was presented by Past President Ralph J. Teetor to A.F.S. President Walton L. Woody on his completion of a successful year as President of the Society. President Woody graciously accepted the Honorary Life Membership certificate and expressed his appreciation.

GUEST SPEAKER

Following presentation of Awards, the guest speaker of the evening, Dr. Kenneth McFarland, Superintendent of Schools, Topeka, Kansas, spoke on the Biblical theme "Which Knew Not Joseph."

PRESIDENT'S ANNUAL ADDRESS

W. L. Woody*

IT IS CUSTOMARY FOR THE PRESIDENT of American Foundrymen's Society to make an annual address to the membership convened at the Annual Business Meeting of the Society during its Annual Convention. The main purpose of this Annual Business Meeting is, first, to give the membership a preview of their officers for the coming year; second, to recognize and do honor to winners in the annual A.F.S. Apprentice Contest; and, third, to present to the membership one of the outstanding scientific men of our day with a diseussion on a subject of interest to foundrymen and vital to all businesses. As I stand before you, I am acutely aware that a "swan song," a recitation of past events, interests very few and makes music only to the bird that is singing. What I have to recite in detail about a specific undertaking in no way should detract from the recognition of the fact that the American Foundrymen's Society exists solely for and does a fine job in providing the time and places for the interchange of technical and scientific information relative to the Foundry Industry.

try

reful er-

on he ort it at

he

ry

A.F.S. Technical Activities

Some 100 technical committees and subcommittees, most of them within the Society's eight major technical divisions, worked effectively and carried out their purposes and functions admirably during the year. The culmination of their efforts is evidenced in the some 100 technical papers and discussions presented during this Convention.

The A.F.S. Chapter meetings and the Regional Conferences have been excellent in their conception and thorough in their execution. Almost all A.F.S. Chapters and Regional Conferences have been visited by Society officers or Board members during the year. The work of these men has been most effective and appreciated.

The A.F.S. Headquarters Staff has efficiently carried on their excellent work. Particularly effective has been the direction of the eight A.F.S.-sponsored re-

search projects and coordinating the scientific work of the Society.

Your Society makes available to the membership preprints of the technical papers presented at the annual Convention. These technical papers are then published in the annual A.F.S. Transactions which is highly recognized by the Foundry Industry throughout the world. The Society also publishes in an able manner the more note worthy papers in the American Foundryman. The Society's list of publications is constantly growing larger.

The Society's finances have been judiciously budgeted and budgets have been more than met and finances are in excellent condition.

Fine cooperation has existed between the American Foundrymen's Society and the foundry trade associations, the Foundry Equipment Manufacturers Association and the Foundry Facings Manufacturers Association. The American Foundrymen's Society has also worked for and with the Foundry Educational Foundation and the National Castings Council in carrying on the purposes of these organizations.

During the year three new Student Chapters have been added to A.F.S. namely, one at Northwestern University, Evanston, Ill., another at Penn State College, State College, Pa., and the third at University of Alabama, Tuscaloosa, Ala. No finer way could be devised to have able young men become interested in the Foundry Industry.

In all this work emphasis has been on the value of this society to the growth of the individual member.

A.F.S. Headquarters Housing Project

This year, in addition to these all important phases of A.F.S. work, the Society undertook another project about which I want to go into in some detail.

Mark Twain said, "The weather was something everybody talked about but nobody did anything about." It seemed that every time the A.F.S. Finance Committee discussed the rent item in the budget they would say, "Why doesn't a large Society like this own their own building?" But nobody did anything about it.

About 14 months ago this question was discussed

^{*} Vice President of Operations and Director, National Malleable & Steel Castings Co., Cleveland.

Delivered during the Annual Business Meeting of the Society at the 1951 A.F.S. Convention, April 25, 1951 in Buffalo.

again and this time something was done about it. President E. W. Horlebein asked the Vice President to propose a plan as a feeler to the A.F.S. Alumni Group. This was done at the Alumni Dinner during the 54th Annual Convention in Cleveland in 1950. It was proposed that \$100,000 be raised for this purpose and that it be sponsored by the Alumni Group. The result was that \$8,000 was pledged before that Convention was over.

On June 22, 1950 this matter was presented to the Chapter Officers Conference and the support of the Chapters was asked. By the time of the Annual Board Meeting, July 28, 1950, \$18,000 had been pledged and

A.F.S. Board approval was assured.

In September, 1950 a letter was mailed to the entire A.F.S. membership regarding the plan. The September, 1950 issue of AMERICAN FOUNDRYMAN carried the first announcement of the approved project. At the September, 1950 A.F.S. Board meeting, the President was empowered to appoint a Building Committee.

Ralph J. Teetor, a Past President of the American Foundrymen's Society, was asked and accepted the President's invitation to serve as Chairman of this important Committee. Mr. Teetor's associates in the Foundry Industry apparently hold him in the same high regard as the President of the Society does for when they were asked to serve with him on this Committee every one of the 15 men asked, accepted the invitation. By this time \$43,000 had been pledged toward the building fund.

The Committee immediately took action and by the

January, 1951 meeting of the A.F.S. Board of Directors \$70,000 had been pledged to the building fund. The Board then asked the Committee to complete the campaign by Convention time (April 1951), which I am happy to report has been more than accomplished.

It was important, of course, to accomplish the purpose of this undertaking. However, and more important, this project has been a large factor in a great upsurge of interest in the American Foundrymen's Society. In the Bible it says, "Where a man's heart is, there will his treasures be also." It is equally true that where your treasure is placed, great interest has first been aroused. It is this aroused interest in A.F.S. which is a by-product far greater than the project itself. This gives your Officers and Directors, and I am sure each member of A.F.S. a sense of great satisfaction with this year's work in this Society.

I take this occasion to thank the many thousands who have helped in all the work of A.F.S. this year and those that have given of their time and money to this particular project. A Society where so many are actively interested is bound to become more useful to the industry. Next year, with able and energetic men as your officers and a great piece of work for the Foundry Industry planned, I am sure, this aroused interest in the work of A.F.S. will indeed be very helpful.

In closing, may I congratulate you on the high caliber of the men you have selected for your next year's officers and directors and again thank those who have

served so well this year.

REPORT OF THE TREASURER

ON JULY 18, following completion of the Annual Audit by George V. Rountree & Co., Certified Public Accountants of Chicago, the Treasurer forwarded to members of the Board complete financial data as follows:

(1) Auditors' Report as of June 30, 1951.

rs

m

r-

t-

p

0

S.

at

st

S.

ct

m

n

15

11

0

e

0

n

le

d

y

e

(2) Consolidated Statement of Income and Expense for 1948-49, 1949-50 and 1950-51.

(3) Proposed Budgets (included in the above Consolidated Statement) for the Fiscal Years 1951-52 and 1952-53.

The financial statements for 1950-51 reported the following. Income \$382,923.28 against a Budget of \$370,500; Expenses \$431,-353.24 against Budget \$442,730. Thus, while we budgeted \$72,230 Excess Expense over Income for the fiscal year, the actual Excess Expense totaled \$48,429.96. It is apparent, therefore, that Income exceeded Budget while Expenses were held below Budget.

Income and Expense

Major variations between Budget and actual Income and Ex-

pense will be discussed separately in this report.

The Auditors statement includes an operational statement of Income and Expense by activities. While this statement does not provide sufficient detail for comparative analysis, it shows that only two Society activities (Member Service and American Foundryman) produce an Excess of Income in a non-exhibit year. All other activities are Expense-incurring. It is obvious that our effort should be directed toward narrowing the gap between Income and Expense in any non-exhibit year and it is hoped that the income from American Foundryman some day will produce this result.

It is interesting to note that income from American Foundryman exceeded the Budget with respect not only to display advertising but also to classified advertising and to subscriptions and sales. Total Budget for these three items was \$128,000; the

Actual was \$132,708.

A comparison of Income and Expense per member for the past two non-exhibit years may be of interest, as follows:

1948-49	1950-51
Total Membership on June 30 10,062	9.340
Total Income\$331,535	\$382.923
Income per Member 32.90	41.00
Total Expense 402,248	431,353
Expense per Member	46.20

Balance Sheet

The financial position of the Society on June 30, 1951, was some \$86,000 better than on June 30, 1950. This is, however, entirely due to acquisition of the Building Fund. In other words, if we eliminate the Fund Principals of the Building Fund and the Awards Fund, the General Fund Principal would be \$238,204 on June 30 this year against \$290,618 a year ago. This is typical of the effect of exhibit and non-exhibit year operations.

Comparative Condensed Balance Sheets for the past four years appear separately. These statements also show a total of Cash and Investments as follows:

June	30,	1951	\$103,913	\$308.805	\$412.718
June	30,	1950	229,497	118,805	348,302
June	30,	1949	85.785	133 875	219.590
Iune	30.	1948	106.394	178,805	285,199

It will be noted that the Accounts Receivable total of \$49,975

on the Comparative statement is not comparable with the total of \$22,011 the previous year because the 1951 figure includes \$35,680 in pledges received for the Building Fund.

It will be noted that Inventories were reduced some \$8,000, that the Deferred and Prepaid items are about the same as the previous year, and that Furniture and Fixtures have been re-

duced slightly.

On the Liabilities side, note that Accounts Payable are below \$20,000, a good condition, and that a reserve of \$1,344 has been set up by the Auditors against possible exchange lcsses (6%)

on deposits in Canadian Banks.

The Deferred Dues Income item, \$109,735, compares with \$108,081 a year ago and is the result of our Dues Income now being placed on an accrual basis for accounting purposes. The Treasurer would be remise if he did not again point out the inconsistency of maintaining part of our books on a cash and part on an accrual basis. From an administrative standpoint, we still do not believe that any advantage is gained by deferring \$108,000 from 1949-50 into 1950-51, and then deferring \$109,000 from 1950-51 into 1951-52.

The method of accruing dues no longer provides any ready determination of the rate at which dues come in, and we again submit that the professed advantage from placing Dues Income on an accrual basis became practically non-existent the moment we adopted our present method of dues pro-rationing. It was claimed by the auditing concern who set up this accrual system that there would be less bookkeeping involved; on the contrary, the amount of bookkeeping has very considerably increased. We, therefore, recommend that the Finance Committee take the time during 1951-52 to consult with the Staff and the present auditor toward the end of possible discontinuation of the accrual method of handling dues Income.

General

During the past year the Society invested in United States Government Securities to a total of \$190,000 (par value). This includes \$125,000 for the General Reserve Fund and \$55,000 for the Building Fund. These investments were made from surplus cash funds. The Secretary-Treasurer is authorized to invest additional Building Fund cash in increments of \$500.00 or more as cash may be accumulated in that fund.

While not a matter of consideration by the retiring Board of Directors, the Treasurer reports that the 1951-52 Finance Committee has agreed to the submission of a two-year Budget covering the period to June 30, 1953. It is believed that a 2-year Budget properly reflects the activities of the Society in view of the staging of an exhibit every second year, the revenue from which must carry the Society's expected Excess Expense of the succeeding non-exhibit year. Better planning, with less tendency toward "feast and famine" operations, should result.

Conclusion

The Society's finances appear to be in good shape for the end of a non-exhibit year, with the Fund Principals (exclusive of all Building Funds) totaling \$263,349 against \$199,930 at the close of the previous non-exhibit year on June 30, 1949. It has been shown that Expenses are under constant control of the Finance Committee and Treasurer, and all Staff members responsible for expenditures are kept continuously cost-conscious, in keeping with the requirements of their activities.

Respectfully submitted, Wyl. W. MALONEY Treasurer

so tr in re

to no do wi

to 12 Sin 13 ter

fo

M M No Av To Ra

cor of it

in an \$7! an to

Mo Ch of is shi

me Ch tha

Condensed Statement of Income And Expense

(All Expenses Distributed to Major Activities) Fiscal Year July 1, 1950—June 30, 1951

INCOME		
	Total I	Expense
	Amount	Per cent
Membership Dues	\$177,853.66	46.1
General Publications	10,820.22	2.7
Special Publications	22,914.20	5.8
AMERICAN FOUNDRYMAN	164,078.49	42.5
Convention	7,077.31	1.8
Miscellaneous	4,887.39	1.1
TOTAL INCOME	\$387,631.27	100.0
Excess Expense over Income	52,413.28	
TOTAL	\$440,044.55	
EXPENSE		
	Total E	
	Amount	Per cent
Membership Department		6.4
Research	42,510.06	9.7
Other Technical Activities	12,272.04	2.8
General Publications	25,996.80	5.9
Special Publications	32,182.41	7.3
AMERICAN FOUNDRYMAN	154,916.46	35.2
Chapter Operations	65,050.26	14.8
Convention	31,325.57	7.1
General Administration	33,308.32	7.6
Retirement Expense	14,237.27	3.2
TOTAL	\$440.044.55	100.0

Comparative Condensed Balance Sheets

As of June, 1948 to 1951, inclusive (4 years)

Assets	June 30, 1948	June 30, 1949	June 30, 1950	June 30, 1951
Cash	\$106,394.93	\$ 85,785.57	\$229,497.13	\$103,913.45
Accounts Receivable	10,003.03	4,405.03	22,011.00	49,975.09
Inventories	33,167.40	35,567.65	35,354.99	27,817.59
Investment Securities	178,805.60	133,805.60	118,805.60	308,805.60
Deferred & Prepaid Items	19,721.40	19,533.00	31,530.03	31,258.11
Furniture & Fixtures (net)	10,744.57	9,791.42	16,451.18	15,858.16
TOTAL	\$358,836.93	\$288,908.27	\$453,649.93	\$537,628.00
LIABILITIES				
Accounts Payable	\$ 20,899.06	\$ 17,085.31	\$ 24,665.99	\$ 19,560.48
Accrued Items	1,099.19	470.91	148.34	141.81
Social Security Payroll Deductions	276.99	204.27	53.12	34.37
Deferred Dues Income	61,439.16	71,217.06	108,081.29	109,735.45
Reserve for Canadian Exchange				1,344.30
Fund Principals	275.122.53	199,930.72	320,701.19	406.811.59
TOTAL	\$358,836.93	\$288,908.27	\$453,649.93	\$537,628.00

REPORT OF THE SECRETARY

This report sums up eight major activities of the Society, with some comments on problems affecting several. The report is confined largely to non-financial matters and to general administrative matters, financial and technical activities being discussed in separate reports of the Treasurer and Technical Director, respectively.

1. Membership

During the past year, Membership in A.F.S. showed a net gain of 294 members, or approximately 3.2 per cent over the total of 9046 on June 30, 1950. The budget estimate was for neither a gain nor a loss. It seems evident that the definite downward trend of the previous year has now been reversed, which should leave no bars against more aggressive membership work.

While the membership total is up, it is important to point out that the number of Sustaining members decreased from 197 to 186 (a loss of 11), and the Company members decreased from 1259 to 1242 (a loss of 17)—a total loss of 28 firm memberships. Since the number of Personal (\$20) memberships increased 13, from 2227 to 2240, there would seem to be some continuing tendency to reduce membership expenditures by converting to lower classes.

A comparison of membership gains and losses over the past four fiscal years, as shown below, brings out two significant points:

1947-48 1948	-49 1949-50	1950-51
--------------	-------------	---------

Members Gained (all Classes) 1,979	1,755	1,787	2,316
Members Lost (all causes) 1,259	2,095	2,804	2,022
Net Gain (or Loss) 720	(340)	(1.017)	294
Average Gain (Loss) per Month 60	(28.3)	(84.7)	24.5
Total as of June 3010,403	10,063	9,046	9.340
Rate of Turnover12.1%	20.8%	31.0%	21.6%

First, during 1950-51 we showed the largest membership gain in four years, and although the net gain was less than in 1947-48, we showed a net gain instead of the net loss of the past two years. This again emphasizes the reversal of a downward membership trend. Second, we have added to this statement a new comparison, namely, Rate of Turnover, which is important to consider for the future of the Society's membership. The Rate of Turnover for 1949-50 was very heavy (31%) and in 1950-51 it was greater than in either 1948-49 or 1947-48. The point should be emphasized that our membership will not increase if we continue to lose as many members as we gain each year.

We would like the Directors to keep these two facts in mind in considering certain elements that affect our ability to solicit and sell memberships. The Society now has spent approximately \$75,000 in staging eight Annual Chapter Officers Conferences, and in these Conferences considerable effort has been directed toward the end of increasing membership by impressing the importance of Society activities on the local Chapter officers.

The National Office today has little to do with membership solicitation work, which is largely in the hands of the Chapter Membership Committees, because of the expressed desire of the Chapters and the National Board of Directors. That delegation of membership work is eminently desirable if the responsibility is accepted by the Chapters, but we submit that the membership work in many Chapters is very loosely conducted, particularly in the effort to hold present members. For example, we set a total of 985 new members as a "Target" for the Chapters during the fiscal year 1950-51, and show an actual gain of 2316 members. However, the net gain was only 294 and only eight Chapters made their targets for the year. It is very obvious that the failure of the other Chapters to do so is simply because of the number of delinquents dropped and not held.

The Secretary's report for June 30, 1950, stated, "There must be greater recognition of the status of our Membership as Circulation, and conscious effort to gain and maintain greater horizontal coverage for the magazine." From this standpoint, a high circulation (i.e., membership) turnover is undesirable, and to this end we have several recommendations to make:

- (1) We repeat our request of last year that the Chapter Contacts Committee might well concentrate on membership work in 1951-52 almost to the exclusion of any other point of discussion with our Chapters.
- (2) We recommend that the National Membership Committee, a body which has largely existed in name only, be made an active corresponding body under Chairman Wagner, with at least a monthly letter from the Chairman to all Chapter Membership Committee Chairmen.
- (3) We believe that greater effort should be made toward the conversion of Personal (\$20) memberships to Company memberships and Company memberships to Sustaining memberships.
- (4) The National Office intends to obtain lists of prospective names from the Chapters for direct membership solicitation from the National Office, a point brought out at the recent Chapter Officers Conference.
- (5) The National Office intends to renew membership targets for 1951-52 with a goal of 10,000 by June 30, 1952, and we urge all members of the Chapter Contacts Committee to emphasize these targets to the Chapters.
- (7) We recommend that the President and Vice-President call a meeting in Chicago this Fall, inviting five or six nearby Chapter Chairmen to discuss specific ways and means for increasing our membership. No business organization, seeking to increase its sales (e.g., membership) could soundly delegate those activities to some group and then have no control whatever over the sales efforts of that group. Yet that is exactly what A.F.S. does on membership work.

A few general membership particulars: The pro-rationing of membership dues continues. Members entering military service will continue to be carried without loss of membership and without further payment of dues. Student members will be carried for one year at the student member dues, even though they leave School and enter employment that year. It should be pointed out that no complaints on the handling and processing of memberships were expressed at the recent Chapter Officers Conference. Our time for processing correct dues payments continues to be seven days.

One particular problem must some day be solved, for in our opinion it will affect the future character of our members, if not the Society itself, and this is the problem of whether the individual member or his firm should be encouraged to pay the membership dues. It is our belief that the Society must some day take a definite stand, for at the present time some 50 per cent of all Personal memberships are paid by company check. If this percentage increases to 80 or 90 per cent, it might be asked whether we then would have a technical society or the basis of a trade association, for in the latter type of organization all dues are paid by the companies. We believe that this is a major policy problem for consideration by the Board Policy Sub-Committee on Membership.

2. Chapters

No new regular Chapters were established during 1950-51, although there has been some activity in New England, which has been explored by Vice-President-elect Wagner with no definite developments to be reported as yet. Three new Student Chapters were formed, at the University of Alabama, Northwestern University, and Pennsylvania State College.

It continues to be our pclicy not to create Chapters, but to encourage and cooperate with any Chapter movement. However, in any case where the territory of a proposed Chapter will overlap a present Chapter, the existing Chapter must be notified immediately of any new Chapter movement.

The 8th Annual Chapter Officers Conference was held June 25-26, 1951, with a total attendance of 101, including 83 Chapter delegates—both being new highs for these events. All 40 Chap-

ters were represented, all but two with at least two delegates, and 5 with three delegates. The total of 83 included 36 Chairmen, 36 Program Chairman-Vice-Chairmen, 8 Secretary-Treasurers, 2 Secretaries and 1 Director (through misunderstanding). Chairman Walter L. Seelbach presided and 14 National Officers and Directors, another new high, were present. Those attending have expressed themselves as receiving a great deal of value from the meeting.

During the past year the National Officers, under Vice-President Seelbach, continued to devote considerable time and effort to visiting Chapters and maintaining closer contacts for the national organization. President Woody attended all but one regional conference and also visited the Chapter Board of Directors on the west coast. The Secretary spent nearly four weeks on the west coast in October and, following that trip, recommended close contact with the west coast Chapters for the next several years.

The two major Chapter problems today are (1) developing greater activity in membership work, and (2) the continuous development of programs that really serve the membership. We suggest that the Chapter Contacts Committee consider their Chapter visits in the nature of "inspection visits" and suggest also that the Committee members, following a Chapter visit, might well address personal letters to the management of companies that do not now support Chapter activities.

3. Convention

A total of 2816 were registered at the 55th Annual Convention of the Scciety in Buffalo, N. Y., April 23-26, including 1935 members, 446 non-members, 199 ladies, and 186 complimentary registrations. The registration fees of \$2 for members and \$5 for non-members seems to have been generally accepted, with few complaints noted. The housing of this Convention was again rather difficult to handle, which should lead us in future years to consider carefully the staging of any Convention in a city having only one outstanding hotel.

The local Convention Committees under the Chairmanship of Chapter Chairman A. A. Diebold carried on their activities in a highly enthusiastic and excellent manner and in no small part contributed to the success of th's Convention. Numerous compliments have been received on the smoothness with which the Convention was run, due largely to the presence of 12 staff members, most of whom are now experienced in the staging of conventions.

An excellent technical program was provided and two innovations were the Defense Production Luncheon on Wednesday and the Foundry Equipment & Supplies Luncheon on Tuesday, both addressed by officials of the National Production Authority. Both sessions were sellouts and probably should be continued during the period of the present emergency.

Two Convention problems need solving: (1) We believe that the Annual Banquet should be carefully considered in order to reduce the time given for speeches and presentations prior to introduction of the main guest speaker. The 1951 Banquet program was much too long and a number were forced to leave before the guest speaker finished, at nearly 11:00 p.m. (2) A four-day Convention program in a non-exhibit year is much too tight, and we strongly recommend that any future non-exhibit Conventions be held over a period of five days, with the Banquet on the third day.

No list of registered attendance is being prepared for the Buffalo Convention.

4. International Foundry Congress-1952

During the past year, considerable thought has been devoted to the International Foundry Congress to be held May 1-7, 1952, in Atlantic City, N. J. A meeting was held in Cleveland in April and, following the Buffalo Convention, the Secretary visited Europe and contacted the foundry organizations in France, Italy, Switzerland, Belgium, The Netherlands, England and Ireland. He found the European foundrymen enthusiastically planning to attend, if possible, under present restrictions on U. S. dollars and further restrictions on the issuance of visas for travel to this country.

While in Europe it was agreed by all that a two-week pre-Convention tour, followed by the Convention week in Atlantic City, would be desirable, and it is estimated that approximately 100 European foundrymen may participate in such a tour. Details will be handled by Tom Makemson, Secretary of the Institute of British Foundrymen, at Manchester, England, and arrangements already have been made for the boat. The tour will be handled by Thomas Cook & Son. President-elect Seelbach and the Secretary contacted ECA in Washington to explain in what manner ECA may be of assistance with this tour.

The fact that President and Mrs. Seelbach attended the International at Brussels in September undoubtedly increased the European interest in the 1952 big event. I do not believe any incumbent president of A.F.S. has attended an International in Europe since 1929.

5. American Foundryman

The fiscal year 1951-52 was the first year during which advertising was solicited by outside publishers representatives, operating strictly on a commission basis. In spite of the fact that all three representatives originally selected were changed during the year, the first year of the 3-year budget for AMERICAN FOUNDRYMAN was met and surpassed.

Closer contacts have been maintained with the members of FEMA and the FFMA, and the number of pages of advertising (448) exceeded the total in any previous year. The Society's membership is now audited by the Audit Bureau of Circulation (ABC), which means that henceforth our circulation will have greater standing with advertisers and advertising agencies. Much still remains to be done in increasing our horizontal circulation coverage, and to this end we must seek membership by a greater number of companies even more than by individuals.

Advertising rates in AMERICAN FOUNDRYMAN were raised some 12 per cent as of July 1, 1951, the full effect of which will not be felt until July 1, 1952. Some discussion has taken place recently toward the end of making the advertising rates in AMERICAN FOUNDRYMAN more comparable with those of The Foundry; we believe that the time for more definite discussion should be when we have made the budget for the second year of the present 3-year budget. We must first overcome the advertising inertia which the magazine has experienced during the past six years, and in the judgment of the Staff, we will do so when we reach an average of 50 pages of advertising per month, but not until that time.

During the past year we carried out considerable effort toward circulation support by mailing copics of the magazine to a selected list of non-member firms, changing the List every four months. Some complaints over this practice were registered at the eighth Chapter Officers Conference, it being felt that these mailings and our subsequent solicitation for members or subscriptions, affected the Chapters' membership solicitation work. From now on we will allow a period of 90 days between the drepping of a member and his being solicited for a subscription, and we will not follow up at all any Canadian members lost.

Certain policies which govern the magazine were brought out in the Secretary's report last year. Since then these policies have been affirmed by the Publications Committee and might well be repeated:

Advertising is sold only on its merits, and no pressure of any kind is exercised under penalty of discrimination in other activities.

No attempt is made to claim greater readership than is known and verifiable.

No special rates or rebates are permitted.

Arrangement of editorial comment is primarily for reader convenience.

Editorials are prepared by men of industry, not by the Staff alone.

Transactions material is not duplicated in the magazine in order to save production expense.

A.F.S. does not pay for any article published.

Articles are not published merely to please advertisers and gain advertising.

6. Technical Activities

The Technical Director's report fully covers these activities. However, a few general comments are called for.

The backbone of A.F.S. is its technical work, and in recent years it has been recognized that this technical work extends not only to the work of our technical committees, but also to the rendering of assistance to our Chapters and the preparation of their technical meeting programs. This includes a greater par-

ONS

In-

and

our

eel-

lain

ter-

any nal

ver-

hat

ing

CAN

of

ing

ty's

ion

ave ion iter me be itly

we be

resing six W.C but ard a our at ese ıbrk. the on, îuc ave be

is

der

nd

ent not the of ticipation by the National Office in the organization and operation of Regional Conferences. We believe this is a sound approach which can be continuously developed so long as realistic consideration is given the matter of Staff assistance.

A full-time Staff man, Kenneth N. Morse, has now been employed to head up our new Safety & Hygiene Program. Mr. Morse will be working under the direct supervision of the Technical Director.

7. Publications

Technical publications also will be covered in the report of the Technical Director. The one problem facing the Staff in the matter of special publications today involves the failure of our committees to complete the manuscripts for some projected ten or twelve important publications.

8. General Administration

Certain decisions have been made regarding the acquirement of a Permanent Headquarters Building for A.F.S. An Action Sub-Committee of the Housing Committee is developing plans. A total of approximately \$140,000 has been raised to date, of which all but \$38,000 represents paid contributions. One hundred per cent of our regular Chapters and two of our Student

Chapters contributed, as well as 375 individuals and 379 companies. A total of 9 individuals and 26 companies each contributed \$1000 or more to the Fund, and we should like at this time to thank all members of the Board of Directors for their contributions, individual or company.

During the past year, no Staff turnover in executive positions occurred, with the exception of the dropping of the Cupola Research worker last October and the hiring of the Safety & Hygiene administrator. No temporary clerical help was hired last summer, for the first time in five years, due to a more competent membership handling and the pro-rationing of dues.

Recognition

We of the Staff wish to extend our appreciation for counsel and guidance to the Officers and Directors with whom we have worked closely during the past 12 months. The year 1950-51 proved to be another eventful period in the progress of the Society, and under the leadership of President Woody has undoubtedly advanced A.F.S. considerably in its recognition by the castings industry.

Respectfully submitted, WM. W. MALONEY Secretary

5-Year Comparative Membership Report

As of June 30, 1947-51, Inclusive

	June 30, 1947	June 30, 1948	June 30, 1949	June 30, 1950	June 30, 1951
Sustaining Members	193	199	221	197	186
Company Members	1,548	1,525	1,380	1,229	1,218
Personal Members	7,358	7,860	7,508	6,571	6,836
Student & Apprentice Members	89	307	430	550	499
Honorary Life Members	64	63	69	75	71
International Members	430	449	455	424	396
TOTAL	9,683	10,403	10,063	9.046	9.206
New Members, 12 Months	2,164	1,979	1,755	1,787	7,316
Resignations	186	286	303	484	319
Delinquent Dropped	829	948	1,759	2,286	1,807
Removed by Death	5	25	33	34	32
Net Gain (Loss) for Year	1,145	720	(340)	(1,017)	158
Members in Chapters	8,809	9,519	9,253	8,306	8,534

A. F. S. Membership Distribution

June 30, 1951

Chapter	Sustaining	Company	Personal	Personal (Affiliate)	Personal (Associate)	Student & F Apprentice	Ionorary Life	Tota
Birmingham	5	27	61	145	6	9		255
ritish Columbia	0	12	13	22	2	1	0	50
anton District	1	25	25	66	1	0	0	118
entral Illinois	2	7	25	108	2	9	1	154
Central Indiana	4	31	50	124	4	4	0	217
	1	9	35	51	4	2 `	0	102
Central Michigan	1	26	42	62	3	4	0	138
Central New York	-				6	0	2	170
Central Ohio	4	22	46	90			_	
hesapeake	3	22	54	46	36	2	7	170
hicago	22	96	177	390	24	12	6	727
Cincinnati District	12	41	74	114	11	15	3	270
etroit	9	50	118	236	14	8	11	446
astern Canada	6	40	62	147	14	12	0	281
astern New York	1	13	16	49	4	8	0	91
fetropolitan	9	55	112	126	20	9	4	335
fexico City	0	4	29	13	0	5	0	51
fichiana	2	25	64	102	2	1	0	196
fo-Kan	1	5	34	8	2	2	0	52
Jortheastern Ohio	19	74	102	283	11	14	7	510
Forthern California	4	18	74	100	9	3	0	208
Vo. Illinois & So. Wisconsin	4	15	20	53	1	2	1	96
	1	23	40	63	2	ī	0	130
orthwestern Pennsylvania	-				3	2	2	323
ntario	4	70	112	130	_	0	-	
regon	2	8	21	66	4	-	0	101
hiladelphia	14	44	97	119	7	3	6	290
uad City	5	36	52	105	7	1	2	208
ochester	1	9	42	36	1	0	0	89
aginaw Valley	5	16	17	210	2	22	1	273
Louis District	1	34	105	104	8	2	0	254
outhern California	4	42	116	89	12	5	0	268
ennessee	0	18	39	52	3	2	0	114
exas	5	26	52	63	5	1	1	153
imberline	1	6	21	29	4	2	0	63
oledo	î	18 .	19	58	1	0	1	98
	0	9	39	25	o	0	i	74
ri-State			48	70	9	0	î	159
win City	5	26			8	.1	1	
Vashington	0	6	26	12				54
Vestern Michigan	3	27	39	151	4	2	1	227
Vestern New York	0	36	77	82	3	11	1	210
Visconsin	14	78	87	290	12	25	3	509
TOTAL IN REGULAR CHAPTERS	176	1149	2282	4089	271 .	202	65	8234
STUDENT CHAPTERS								
fassachusetts Inst. of Technology	-	_	-	_	2	40	-	42
lichigan State College	_	_	-	_	_	42	_	42
lissouri School of Mines	_	-	_	-	4	60	-	64
orthwestern University	_	_	_	_	-	20	_	20
hio State University			_	_	_	20	_	20
	_					10		10
regon State College	_	_				22		22
enn. State College	_	-	_			12		12
exas A & M College	-		_	_	_		_	
niversity of Alabama	_	_	-	_	_	25	_	25
niversity of Illinois	-		-	_	-	29	_	29
niversity of Minnesota	_	_	_	_	_	14	_	14
TOTAL IN STUDENT CHAPTERS	_	_	-	_	6	294	_	300
TOTAL IN ALL CHAPTERS	176	1149	2282	4089	277	496	65	8534
nternational	0	24	348	10	12	2	0	396
on-Chapter	10	69	110	48	30	3	6	276
GRAND TOTAL	186	1242	2740	4147	319	501	71	9206
	100	A day 5 day	41.10		0.0			0200

REPORT OF THE TECHNICAL DIRECTOR

THIS PROGRESS REPORT covers the period beginning July 1, 1950 and ending June 30, 1951.

Technical Activities

Our technical committee personnel roster has been in a state of revision, in order to eliminate members of various committees who have found it impossible to be active in the work of the committee. In addition, of course, it is always considered desirable to add new members to the committees, who are both qualified and willing to devote their time and effort to our technical affairs. Also, in some instances, it has become desirable to abolish a committee when its originally intended purpose has been completed and further work considered unnecessary.

At the end of each two-year period it likewise becomes necessary for each of the Divisions involved to select a new Chairman and Vice-Chairman. Several such changes have taken place this year. Under these circumstances, it is apparent that it will be necessary to publish a new Committee Personnel Roster, since the various committee chairmen and members find them particularly useful especially when they are attempting to integrate their work with that of other committees.

Following action of the Board of Directors at its meeting last January, we have been fortunate in obtaining the services of Kenneth E. Morse to direct the Safety, Hygiene and Air Pollution activities of the Society. Our initial efforts in this field will be the preparation of a suitable brochure and the appointment of a fund solicitation committee as well as a working committee, to acquire the funds necessary for the prosecution of this work, in addition to the technical activities involved.

In the latter field, our first efforts will be confined to a careful review of our present Recommended Practices and their modernization, in contemplation of publishing revised editions.

A.F.S.-Sponsored Research

At the Annual Meeting of the Board of Directors in 1950 a Research Committee of the Board, comprised of Messrs. Dunbeck, Farquhar and Zang, was appointed for the purpose of reviewing the several Society-sponsored research projects and to recommend budgetary appropriations for these projects. At the meeting of this Committee held July 9, 1951, the Technical Director prepared a summary of A.F.S. Research Projects.

A.F.S.-sponsored research projects now being conducted are as follows:

Aluminum & Magnesium Division—"A Study of the Principles of Gating," conducted at Battelle Memorial Institute, Columbus, Ohio.

Brass & Bronze Division—"The Fracture Test as an Indication of the Melt Quality of Tin Bronze," conducted at University of Michigan, Ann Arbor, Mich.

Malleable Division—"The Effect of Melting Conditions on Anneability, Hot Crack Formation, and Fluidity of Malleable Iron," conducted at University of Wisconsin, Madison, Wis.

Sand Division—"Properties of Sand at Elevated Temperatures," conducted at Cornell University, Ithaca, N. Y.

Steel Division—"Influence of Mold Conditions on the Development of Hot Tears In Steel Castings," conducted at Armour Research Foundation, Chicago.

Gray Iron Division—"Gating and Risering of Gray Iron Castings," conducted at Massachusetts Institute of Technology, Cambridge, Mass.

Centrifugal Casting—"The Study of Feasibility of Centrifugally Casting Light Alloys," conducted at Canadian Bureau of Mines, Ottawa, Ont., Canada.

Heat Transfer Investigations—"Fundamental Studies of Heat Transfer," conducted at Columbia University, New York.

1951 Convention Program

Judging from the numerous statements of commendation made

by members during and since the Annual Convention, it is believed that this unquestionably was one of the most successful and smoothly operated Conventions we have had up to the present time. The attendance exceeded our last non-exhibit Convention by approximately 800 registrants, the total being slightly over 3,000. Fifty different functions were scheduled during the Convention, in which 90 authors or discussion leaders participated. Of the formal papers presented, 43 were published as separate preprints, including five in AMERICAN FOUNDRYMAN in lieu of separate preprints. A larger number of the papers could not be preprinted in advance of the Convention due to the fact that the manuscripts were not received from the authors and subsequently approved by the Program and Papers Committee, in sufficient time to allow for their editing and production.

Approximately 3,500 requests were received from members of the Society for one or more preprints. When the requests were received before the specified deadline date, preprints were mailed to members approximately one month in advance of the Convention, to afford them an adequate opportunity to review the papers and come prepared to enter into an intelligent discussion following the presentation of the paper at the technical session.

Three functions were scheduled during the Convention which are worthy of special mention, namely, the Foundry Equipment and Supplies Luncheon, the Defense Production Luncheon and the Symposium on the Principles of Gating. The latter was unusual in that it was the only technical function scheduled at that time, and because of its general interest to all foundrymen, resulted in a remarkable attendance, estimated at 1200, even though it started one hour in advance of the normal time for technical sessions, and lasted throughout the entire day. A total of 14 authors participated in this Symposium, and all of the papers, together with both written and oral discussions, are published as a separate book covering the entire proceedings of the Symposium.

Publications

The Annual Transactions of the Society for 1950, Volume 58, comprising 760 pages, was published and shipped during the month of December, 1950, to all who had ordered it on a prepublication basis. A total of 1122 copies of this volume has been sold and in addition 207 furnished gratis, on request, to Sustaining and Honorary Members of the Society, making a total of 1329 copies distributed, of the 1500 copies produced.

In accordance with our established policy, these Transactions were sold at a pre-publication price of \$6.00 to members of the Society, distributed gratis to Sustaining and Honorary Members who requested them, a relatively small number sold to members at the post-publication price of \$8.00, and to non-members of the Society at \$15.00 each.

As part of the Annual Transactions it has been the policy of the Society to provide a stenotype recorder at each of the technical sessions during the Convention, for the purpose of recording any oral discussion which followed the presentation of the paper. These stenotype records are then reproduced and sent to participants for correction, subsequently edited by Jos. E. Foster, Technical Assistant, at which time any irrelevant or undesirable discussion is deleted, and the balance published in the Transactions immediately following the technical paper itself.

The actual recording of the oral discussion by the stenotypist, together with the expense of setting the additional type and its printing in the Transactions, represents an estimated cost of \$700 to \$1,000 each year. Usually such oral discussion adds relatively little, as compared with written discussions prepared in advance, which are well planned and intelligently presented.

Under these circumstances, it is the recommendation of the writer that we dispense with the services of the stenotypist recorders at the Annual Conventions and instead confine discussions of technical papers primarily to written discussions and

R

ta w al sl

da So ta

such brief summary as a pre-appointed representative from the appropriate Division might prepare in written form. In other words, it would be left to his judgment to include only important points which might be developed during oral discussion, and possibly, if deemed advisable, actually communicate with the discusser in an attempt to induce him to prepare it in written form.

During the fiscal year the following publications were produced and made available for sale:

	lumber rinted	Member Price	Non- Member Price
a. Engineering Properties of Cast Iron b. Symposium on Interpretation and Application of Sand Test Data for	1	\$2.25	\$3.50
the Production of Quality Stee Castings	500	3.00	4.75
No. 1 Gray Iron Research Progress Report No. 2	t	2.00	4.00
d. Foundry Work (a high school text-book)		2.00	2.00

* This book is under contract for publication and distribution by John Wiley & Sons of New York. Because of the inadequate marketing facilities of the Society for a book of this kind, and our desire to obtain maximum distribution, it was decided to contract with a well recognized publisher for the production and sale of this book, at an estimated price of \$1.60 per copy. The first edition will comprise 10,000 copies.

Following failure on the part of the author originally invited to write this book, we were fortunate enough to obtain the services of Edwin W. Doe, Chairman of the Metal Shops Dept., Brooklyn Technical High School, Brooklyn N. Y., who has done a commendable job in preparing this text in a very short time.

Most recent advice received from Prof. Peter E. Kyle of Cornell University, the author of the college foundry textbook, is that the manuscript will be ready by the end of the summer. This likewise will be submitted to several recognized publishers of textbooks, for bids in anticipation of a contract for its publication and sale.

In addition to the first four publications mentioned above, our own printing facilities as well as staff personnel, have been utilized for the production of the A.F.S. Alumni Directory and the Chapter Officers Manual.

The Sand Testing Handbook is in the last stages of revision and it is practically certain that the new edition will be published and offered for sale before the end of the calendar year.

The Glossary of Foundry Terminology is in process of editing, and while it still requires a considerable amount of work, substantial progress has been made towards its completion. It is hoped that within the coming year this work can be completed and made available.

Advance or Pre-publication Order Forms for the Annual Transactions of the Scciety for 1951, Volume 59, have been sent to the entire membership. Subscriptions to the Transactions will again be on a prepaid, pre-publication basis at a price of \$6.00 to members and \$15.00 to non-members of the Society, with gratis copies to all Sustaining and Honorary Members who request them. Subsequent to the close of the pre-publication offer, Sept. 30, members will be charged \$8.00 per copy and non-members \$15.00 for this publication. Only a clothbound edition will be produced and it is expected that it will be ready for shipment during December, 1951.

As is customary, the Chas. Edgar Hoyt Annual Lecture, this year presented by Mr. James C. Zeder, Vice-President and Director of Engineering and Research of the Chrysler Corporation, has been published and is now available for sale. Th.s paper is included in this volume of the annual Transactions (p. 1).

The manuscripts for a number of other special publications are anticipated in the near future. In addition, the CAST METALS HANDBOOK as well as the CUPOLA HANDBOOK must be completely revised and published, since our present supply is relatively low. In order to assume this load, it will be necessary to increase the technical staff or publication of these works must of necessity be materially delayed. Inasmuch as our present policy is to price all publications so as to yield a modest profit, it appears highly

desirable to acquire such necessary personnel to expedite this work. Beyond this, as a technical society, we have a definite obligation to the Industry to publish these technical works promptly when the manuscripts are available, and thereby avoid premature obsolescence of the information they contain. The Publications Committee of the Board of Directors will make a separate recommendation regarding this situation.

Technical Inquiries

Technical inquiries continue to be received at National Headquarters and when possible, are answered promptly reflecting the best available information. During the year a total of 401 individual requests was received. Lack of a library and a well qualified technical librarian, of course, make it difficult and time consuming in many instances, to prepare a reply which will be of value to the recipient.

Apprentice Contest

Each year the Society sponsors an Apprentice Contest. For the second consecutive year the contest has included metal patternmaking as well as wood patternmaking, and molding in the fields of light metals, copper base, gray iron and steel. A total of 246 entries was received from 76 companies and 8 Chapters. This represents a reduction from the 1950 total of 267, but same unquestionably is attributable to the large number of potential participants who have entered the armed forces and some who have been advanced prematurely within their own organizations, thus making them ineligible under the rules of the contest.

The Apprentice Contest Committee of the Educational Division deserves a great deal of credit, beginning with the preparation of the contest drawings, to the final judging of the entries.

Chapter Speakers List

As an aid to the Chapters in preparing a good technical program for the year's meetings, a new speakers list is made available each year to all Chapter representatives who attend the Chapter Officers Conference, and will be sent to others when requested. The revision of this list each year is based in a large part upon the Speaker Evaluation Reports received from the Chapters during the year, so as to avoid continued listing of speakers who have failed to perform acceptably at Chapter Meetings. New speakers of real merit are, of course, added when the list is being revised.

Attesting to the value of this list, it is noteworthy that only 29 per cent of the speakers named by the Chapter representatives as reflecting the two best they have had at their Chapter during the preceding year, were not included in the official Chapter Speakers List, as prepared by National Headquarters.

A.F.S. Film Directory

As a further service to the Chapters, we have again prepared a new and revised list of films relating to the Foundry Industry, which are known to be available and of probable interest for showing at Chapter Meetings.

The A.F.S. colored, sound film on "The Principles of Gating" which has resulted from the research project sponsored at Battelle Memorial Institute, under the direction of the Aluminum and Magnesium Division Research Committee, has had a total of 81 showings since October, 1950. Of this total, 28 were before Chapters, 8 at schools, 3 before regional conferences, 8 at other associations, and 34 to individual company employee groups. A considerable number of additional requests for future showings are already at hand. The Technical Director personally showed this film before 13 Chapters, including a trip to the Southern and Northern California and Timberline Chapters.

The interest in this film has been beyond expectation and certainly justifies the production of a new film, which it is contemplated will have its initial showing before the Annual Convention of the Society in Atlantic City in May, 1952. Three individual companies have made outright purchases of copies of this film at a price of \$400 each; one in this country and two in Switzerland.

The previous black and white silent film has been shown before approximately 10 societies and company groups in England, France, Belgium and Germany and requests are continuing to be received for its use. NS

his

ite

rks

oid

he

a

nal

ct-

of

2

nd

ill

or

at-

he

tal rs. ne ial ho ns,

he en ge he of er en ly a-er al

d

t-maler A gs d n

d

Recorded Lectures for Chapter Use

A program has been initiated involving the preparation of good technical presentations on various subjects, together with an appropriate number of good lantern slides. These technical talks will be recorded on magnetic tape so that when a Chapter wishes a technical presentation on a given subject, we will be able to send them a spool of magnetic tape and the lantern slides. The first talk so prepared covers the developments to date revealed by the Brass and Bronze research sponsored by the Society at the University of Michigan. An initial sample presentation of this talk was made before the Chapter Officers Conference in June, with favorable response. It is hoped that over a period of time we may develop a library of such technical presentations, primarily for the use of the Chapters, but also for student groups and individual plant gatherings.

Other Technical Activities

Close liaison with other technical societies continues, the Technical Director holding memberships in the following: American Society of Mechanical Engineers American Society for Metals American Institute of Mining & Metallurgical Engineers
American Ordnance Association
American Society for Testing Materials
Society of Automotive Engineers
Society for Non-Destructive Testing
Society for Advancement of Engineering Education

Society for Advancement of Engineering Education
At the Annual Meeting of the American Society of Mechanical
Engineers, the writer was elected Chairman of the Metals Engineering Division for the coming year and also has continued as
a member of the Professional Divisions Committee of ASME.

Even though the Technical Director has continued as a member of the Technical Advisory Committee of the Munitions Board of the Armed Forces, only one meeting of this group has been held during the past two years and it is the writer's opinion that the value of this Committee is relatively little because it has not been given an opportunity to effectively function.

In closing, the writer wishes to express appreciation for the cooperation and understanding attitude of the Board of Directors in assisting to make our technical activities successful during the past fiscal year.

Respectfully submitted, S. C. MASSARI Technical Director

A

at

Ca H Ca cu

Bo

Co

bu

Ex

bu

Minutes Special Meeting 1950-1951 Board of Directors

Drake Hotel, Chicago-Friday, July 27, 1951

Present: President Walton L. Woody. presiding Vice-President Walter L. Seelbach

Directors

ICALOID.		
Terms exp. 1951)	(Terms exp. 1952)	(Terms exp. 1953)
T. H. Benners, Jr.	T. E. Eagan	J. J. McFadyen
N. J. Dunbeck	L. C. Farquhar	F. W. Shipley
Robert Gregg	V. J. Sedlon	James Thomson
M. J. O'Brien, Jr.	F. G. Sefing	E. C. Troy
V F. Zang	I. D. Wright	

Absent: Director J. O. Ostergren

President Woody declared that this special meeting had been called strictly for the purpose of organizing the Executive Committee in compliance with provisions of the Society's By-Laws.

He announced that it was his intention, as President, to establish regular meetings of the Board of Directors and to hold meetings of the Executive Committee only in cases of emergency requiring immediate decision. He stated that it would not be the intention to screen through the Executive Committee first all subjects to be discussed in regular meetings of the Board.

Accordingly, he requested that the Directors vote by letter ballot to elect four members of the Board who would serve, together with the President and Vice-President Seelbach as an executive Committee of six members, under the By-Laws. Letter ballot then took place, with the result that the 1950-51 Executive Committee was announced as follows:

President Walton L. Woody, Chairman Vice-President Walter L. Seelbach Director and Past-President E. W. Horlebein Director N. J. Dunbeck Director F. G. Sefing

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved: WALTON L. WOODY, President

Director V. E. Zang

Minutes Special Meeting of 1950-1951 Executive Committee

Drake Hotel, Chicago - Friday, July 27, 1950

Present: President Walton L. Woody, presiding Vice-President Walter L. Seelbach

Directors:

Past-President and Director E. W. Horlebein

Director N. J. Dunbeck Director F. G. Sefing Director V. E. Zang

The By-Laws of the Society, Article X, Sections 1-9, inclusive, were read, describing the method by which a Nominating Committee would be appointed and fulfill its functions for the election of a Vice-President and new Directors of the Society at the 1951 A.F.S. Convention.

The names of various candidates were selected from the list of candidates submitted by eligible Chapters and the following seven members were appointed by the Executive Committee to serve, together with past-Presidents E. W. Horlebein and W. B. Wallis, who form a Nominating Committee of nine members:

Past-President E. W. Horlebein, Pres., Gibson & Kirk Co., Baltimore, Md.—Chairman

Past-President W. B. Wallis, Pres., Pittsburgh Lectromelt Furnace Co., Pittsburgh, Pa.

John Cauffiel, Owner, St. Clair Bronze & Aluminum Co. (Rep. Toledo Chapter and Brass & Bronze—Aluminum) C. K. Donoho, Chief Met., American Cast Iron Pipe Co., Birmingham, Ala. (Rep. Birmingham Chapter and Gray Iron-Steel)

Francis T. O'Hare, Pres., Central Brass & Alum. Co., St. Louis, Mo. (Rep. St. Louis Chapter and Brass & Bronze-Aluminum) H. M. Patton, Fdry. Supt., American Hoist & Derrick Co., St.

Paul, Minn. (Rep. Twin City Chapter and Gray Iron-Steel)
Vaughan C. Reid, Vice-Pres., City Pattern Foundry & Machine
Co., Detroit (Rep. Detroit Chapter and PatternmakingSupplies)

S. D. Russell, Pres., Phoenix Iron Works Co., Oakland, Calif. (Rep. Northern California Chapter and Gray Iron)

George F. Watson, Asst. Chief Chem., American Brake Shoe Co., Mahwah, N. J. (Rep. Metropolitan Chapter and Gray Iron) It was announced that the date of the meeting of the Nominating Committee would be selected by the Chairman at a later date and notified to the Committee personnel.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted, Wm. W. Maloney Secretary-Treasurer

Approved:

WALTON L. WOODY, President

Minutes First Meeting 1950-1951 Board of Directors

Drake Hotel, Chicago-Saturday, July 28, 1950

Present: President Walton L. Woody, presiding Vice-President Walter L. Seelbach

Directors:

IONS

etter

erve.

s an

etter

ecu

ting

rm-

on-

uis.

m)

St.

ine

ng-

lif.

Co.

on)

in-

ter

ing

(Terms exp. 1951) (Terms exp. 1952) (Terms exp. 1953) J. J. McFadyen F. W. Shipley T. E. Eagan T. H. Benners, Jr. L. C. Farquhar N. J. Dunbeck James Thomson V. J. Sedlon Robert Gregg M. J. O'Brien, Jr. E. C. Troy F. G. Sefing V. E. Zang L. D. Wright E. W. Horlebein

Secretary-Treasurer Wm. W. Maloney Technical Director S. C. Massari

Absent: Director J. O. Ostergren

Election of Secretary-Treasurer

At the outset, the President appointed the following Nominating Committee of three to place in nomination the names of candidates for the election of Secretary and Treasurer: E. W. Horlebein, Chairman, T. E. Eagan, and L. C. Farquhar. The Committee promptly placed in nomination the name of the incumbent Secretary-Treasurer, and on motion duly made, seconded and unanimously carried, Secretary-Treasurer Wm. W. Maloney was declared re-elected for the fiscal year 1950-51.

Fixing of Salaries and Compensations

In executive session with members of the Staff excluded, the Board of Directors established salaries and compensations for the members of the National Office Staff for the year 1950-51.

Budget of Estimated Income and Expense

The Treasurer then was requested to present the budget of estimated income and expense for the fiscal year July 1, 1950 to June 30, 1951, inclusive, as recommended by the Finance Committee. President Woody announced that in setting this budget, the Committee had been aiming at building up a reserve of approximately \$450,000 on hand at the end of each non-Exhibit year, that the Society had not yet reached this point, but that the goal had been kept in mind in recommending a hudget

Each item of Income and Expense as recommended was presented, discussed and various questions answered. Several actions

then were taken, as follows:

(a) On motion by Director Eagan, seconded by Director Troy and carried, the President was authorized to appoint a Research Committee of Board members to review requests of the various Committees for research funds and be prepared to make recommendations thereon to the Board of Directors; to establish, subject to Board approval, procedures to be followed in handling the Society's Research Projects; to keep in touch with research progress and to prepare a semi-annual report accounting for expenditures with recommendations as to future procedures; to be responsible to the Board of Directors that research work as financed is carried to logical and final conclusions. It was understood that the purpose of this Board Research Committee was to keep the Board more fully acquainted with procedures, progress, and need for the various A.F.S.-Sponsored Research Programs.

(b) Following discussion, on motion by Director Dunbeck, seconded by Vice-President Seelbach and carried, the budget of \$33,250 for A.F.S.-sponsored research during 1950-51 was approved. Directors Troy and Sefing dis-

senting.

(c) On motion by Director Zang, seconded by Director Gregg and carried, the Finance Committee's budget of \$115,580 for salaries of members of the National Office Staff during 1950-51 was approved.

(d) On motion by Director Shipley, seconded by Director Zang and carried, the budget of Estimated Income and Expense for 1950-51 as recommended by the Finance Committee and as revised by the Board, was declared approved.

Standing and Special Committees

The President declared that he would entertain a motion authorizing his appointment of Standing and Special Committees as provided in the By-Laws of the Society. Motion was made by Director Horlebein, seconded by Director Sedlon and carried, providing such authorization.

President Woody stated that he would present to the Board of Directors at its next meeting the full complement of Standing and Special Committee, not hitherto appointed. Certain com-

mittees, however, were dealt with as follows:

(a) Finance Committee—To consist of President Woody, Vice-President Walter L. Seelbach, Director and Past-President E. W. Horlebein.

(b) National Membership Committee—Vice-President Walter L. Seelbach, Chairman; all Chapter Membership Committee Chairmen to serve as members.

(c) Chapter Contacts Committee—Vice-President Walter L. Seelbach, Chairman; all other members of the Board of Directors to serve as members, in accordance with requests of the Chairman.

(d) Auditing Committee—Not appointed since its functions are adequately fulfilled by the Finance Committee.

National Castings Council Representatives

The President recommended, and on motion duly made, seconded and carried, the Board approved appointment of Vice-President Walter L. Seelbach to serve with President Woody as the second of two A.F.S. representatives to the National Castings Council

Recommendations of 1949-50 Board of Directors

- (a) Recommendations of the old Board covering recommendations of the Publications Committee were accepted by the new Board, on motion duly made, seconded and carried.
- (b) On motion duly made, seconded and carried, the recommendation of the 1949-50 Board, namely, that the 1952 International Foundry Congress and Exhibit be held in Atlantic City, was approved.

1951 Convention Registration Fees

The Secretary was directed to present some of the numerous comments received concerning Registration Fees charged at the 1949 and 1950 Conventions. It was pointed out that the Budget of Income and Expense for 1950-51 as approved, contemplated certain revisions from Registration Fees on the basis of a charge of \$2 to members and \$5 to non-members. Following discussion was made by Director Eagan, seconded by Director Farquhar and unanimously carried establishing the fees of \$2 and \$5 for members and non-members respectively at the 1951 non-Exhibit Convention in Buffalo.

It was pointed out that fees for the 1952 International Foundry Congress and Exhibit would be determined by the Board of Directors at a later date. The Directors requested the Staff to particularly publicize the fact that no Registration Fee would be required of anyone desiring to attend the Annual Business Meeting of the Society in 1951 at Buffalo.

Sand Consultant—The Technical Director pointed out that Dr. H. Ries had been for years retained by the Society as a consultant to the Sand Division and requested approval of a certain expenditure for that purpose in 1950-51. On motion by Director Wright, seconded by Director Sefing and carried, the recommendation was approved.

The secretary requested Board approval for holding of the 8th Annual Chapter Officers Conference in June 1951. Following discussion, motion was made, seconded and unanimously carried

THAT the 1951 Chapter Officers Conference be authorized, with invitations extended to Chapter Officers and to National Directors on the same basis as in 1950.

A.F.S. Building Project

President Walton L. Woody reported that a plan for acquiring a permanent home for A.F.S. had been presented to the A.F.S. Alumni Group on May 11 during the Cleveland Convention. The plan as proposed was to raise approximately \$100,000 to buy or build a permanent home, and thus eventually accomplish considerable savings from the large sums expended by the Society for housing the National Headquarters.

By the time of the Chapter Officers Conference June 26-27, the President stated, over \$15,000 in pledges had been received from members of the Alumni Group. Therefore, it was suggested to the Chapter Officers Conference that the Chapters might wish to support the program themselves. Full discussion was held and numerous questions answered, with the result that the Conference went on record as fully in favor of advancing the program once the project had been approved by the A.F.S. and the Chapter Boards of Directors. Mr. Woody had requested the Chapter Officers to meet with their respective Boards and to indicate their reactions prior to the Annual Meeting of the A.F.S. Board.

He stated that by July 27, 1950 a total of over \$9,000 had been pledged by only eight Chapters of the 40 Chapters of the Society, but that practically every Chapter responding since the Conference had approved the project. He also stated that to date approximately \$30,000 had been pledged by members of the Alumni Group, the Chapters and a few personal contributions. In the Event of Board approval, appropriate committees are to be appointed on such phases as Solicitation, Financing and Location, Plans and Construction, etc., all subject to Board approval.

President Woody stated that the question of deductibility of contributions for Income Tax purposes had been discussed with the Society's attorney, whose reply dated July 19 included the following paragraph:

"It is our opinion that, so long as objects and activities of the Society are as at present, that is, of a scientific and educational nature, and if the proposed building is limited to use as a home for the Society, contributions by individuals to the fund for such a building are deductible for income tax purposes by the contributor.

After further discussion, the following resolution was made by Director Eagan, and seconded by Director Farquhar:

RESOLVED that the Officers of the Society be, and they are hereby authorized to create a fund for the purpose of acquiring a property to be used exclusively as the home of the Society and solely for its Charter purposes.

RESOLVED FURTHER that said fund shall be solicited from the membership of the Society.

RESOLVED FURTHER that when said fund is sufficient for the purpose, the Officers of the Society are further hereby authorized, in the name of the Society, to take such steps as are necessary to the acquisition of said property. Motion was unanimously carried.

Safety & Hygiene

President Woody and the Secretary reported on progress to date in reestablishment of a strong Safety and Hygiene Committee within the Society. It was pointed out that the Society had sounded out a number of large companies on the question of support for the project and, their responses being encouraging, the A.F.S. Safety & Hygiene Committee invited representatives of these companies to a meeting in December, 1949. The meeting held then recommended to the A.F.S. Board that a program involving the raising of \$50,000 a year for five years be instituted, that a full-time Staff man be employed to carry on the project, and also outlined a series of projects for initiation.

The A.F.S. Board at a January, 1950 meeting, approved the recommendation to re-establish a strong Safety & Hygiene project, but revised the recommendations and authorized the raising of \$35,000 a year for a ten-year period, with employment of a

full-time Staff member to head up the work.

With this background sketched in, President Woody then stated that it was his desire for the new Board of Directors to re-affirm the actions of the previous Board to the end that the project might be initiated approximately October 1, 1950.

The Directors expressed the opinion that the name "Safety & Hygiene Project" should be revised to cover air pollution as well, and the name of "Safety & Hygiene and Air Pollution Project" was substituted.

Several newer members of the Board suggested that more study should be made of all phases of solicitation and expense of such a project, and, following discussion, motion was made by Director Eagan, seconded by Director Troy and carried, authorizing the President to appoint a committee "to develop all phases of solicitation and operation under the name of a Safety & Hygiene and Air Pollution Project." It was intended that this Committee shall report to the Board at its next meeting for finalization of the program.

Cupola Research

The Technical Director presented that phase of his annual report dealing with the Cupola Research Project, pointing out that the recent death of R. G. McElwee, Chairman of the Project since 1943, called for re-alignment of cupola research activities, especially to the end that closer control be established over expenditures and evaluation of results. He therefore recommended that a full-time research worker be dispensed with on completion of certain work now in process, that future work on cupola research be carfied on at some institution in the same manner as other A.F.S.-sponsored research project, and that if the Board so desire, additional funds be raised by some accepted method which would make possible continuation of the work. In the meantime, he recommended that the Cupola Research Committee determine, for recommendations to the next meeting of the Board of Directors, an evaluation of its work to date and recommendations for the future.

On motion made, seconded and carried, the Board voted to accept the recommendations of the Technical Director on cupola research.

Director Eagan stated that there was a need for a committee of the Board to develop procedures and evaluate results to date, as in the case of other research projects. He therefore moved that the present Cupola Research Committee be expanded through appointments by the President and that this committee be requested to report to the Board at its next meeting on the scale and basis of operation for future cupola research activities. Motion was seconded and carried.

Without vote, the Board suggested that the Annual Report of the Technical Director be forwarded to all Company and Sustaining members, being analyzed in advance for possible revision.

Approval of Resolutions

The Secretary requested Board approval of the following resolutions required for conducting financial and various other affairs of the Society, and on motions duly made, seconded and carried, the resolutions were approved as follows:

- (a) RESOLVED that resolutions required by the Harris Trust & Savings Bank of Chicago, authorizing the withdrawal of funds, are hereby approved and the Secretary authorized to certify thereto.
- (b) RESOLVED that checks for the withdrawal of funds deposited in the name of the Society with depository banks, including all General Checking accounts and Interest Savings Accounts, and for the withdrawal of all securities held in the various funds of the Society by the Trust Department of the Harris Trust & Savings Bank of Chicago, shall require the signatures of any two of the following Officers: President, Vice-President, Secretary-Treasurer, Technical Director.
- (c) RESOLVED that the Secretary be authorized to rent a safety deposit box at the Harris Trust & Savings Bank of Chicago for the safe-keeping of Society securities, and that any two of the following have authority to obtain access to such safety box: President, Vice-President, Secretary-Treasurer, Technical Director.
- (d) RESOLVED that resolutions required by the Royal Bank of Canada, Montreal, Quebec, and by its subsidiary banks at Toronto, Ontario and Vancouver, B. C., establishing checking accounts in the name of the Society and authorizing the withdrawał of funds therefrom, are hereby approved and the Secretary authorized to certify thereto.
- RESOLVED that the Treasurers of the Eastern Canada, Ontario and British Columbia Chapters of the Society are

the

ONS

ell.

ect"

ore

e of

by

or-

fety

this

for

ual

out

ject

ies,

ex-

ded

ion

re-

28

ard

nod

the

nit-

the

ind

ola

tee

ite,

ved

led

tee

the

ies.

of

uson.

her

nd

of zed

08-

ks,

est

ies

ust

hi-

ol-

ry-

ety

go

wo

ch

er,

of

at

ng

p-

la

authorized to make deposits, in the Royal Bank of Canada and its subsidiary banks, of funds received from Canadian members in payment solely of membership dues in the Society, and the Secretary is authorized to make all necessary accounting arrangements.

- (f) Resolved that checks for the withdrawal of funds deposited in the name of the Society with Canadian depository banks, shall require the signatures of any two of the following Officers: President, Vice-President, Secretary-Treasurer, Technical Director.
- (g) RESOLVED that the Treasurers of the Eastern Canada, Ontario and British Columbia Chapters of the Society shall each give evidence to the Secretary of the existence of Indemnity Bonds covering each of them individually and separately, in amounts deemed satisfactory to the Society, premiums thereon to be paid by the respective Chapters.
- (h) RESOLVED that the present Blanket Indemnity Bond be renewed covering all Staff members for Five Thousand Dollars (\$5,000.00), expect that the Indemnity Bonds be purchased covering the Secretary-Treasurer and Technical Director for Fifty Thousand Dollars (\$50,000.00) each, premiums to be paid by the Society.
- (i) RESOLVED that the Secretary be authorized to execute all contracts for the administration of Society affairs, subject to specific approval by the Board of Directors. In the case of A.F.S.-sponsored research projects, approval of the project by the Board of Directors includes the authority for the Secretary to execute contracts for performance of such projects on a bid basis.
- (j) RESOLVED that the Secretary be authorized to reimburse traveling expenses for members in attendance at any regularly called Board of Directors, Executive Committee, or Technical Committee meeting, with the following exceptions: No expenses shall be paid to Directors or Committee members for attendance at meetings held during the week of the Annual Convention of the Society, unless specifically authorized by the Board of Directors. When meetings are held in conjunction with other committees of associations, the Secretary is authorized to determine what portion of the expense of such attendance shall be paid by the Society.
- (k) RESOLVED that the Secretary be authorized to negotiate the

compensations of necessary Staff employees below \$3,600.00 per year, and that Finance Committee approval be required on all compensations of \$3,600.00 per year or more.

1951 Nominating Committee

For the information of the Board, President Woody reported that the Executive Committee, in special session the afternoon of July 27, had appointed seven members of the Society to serve with the two immediate past-Presidents to form the 1951 Nominating Committee, as follows:

- Past-President E. W. Horlebein, Pres., Gibson & Kirk Co., Baltimore, Md.-Chairman
- Past-President W. B. Wallis, Pres., Pittsburgh Lectromelt Furnace Co., Pittsburgh, Pa.
- John Cauffiel, Owner, St. Clair Bronze & Aluminum Co. (Rep. Toledo Chapter and Brass & Bronze-Aluminum)
- C. K. Donoho, Chief Met., American Cast Iron Pipe Co., Birmingham, Ala. (Rep. Birmingham Chapter and Gray Iron-Steel)
- Francis T. O'Hare, Pres., Central Brass & Aluminum Co., St. Louis, Mo. (Rep. St. Louis Chapter and Brass & Bronze-Aluminum)
- H. M. Patton, Fdy. Supt., American Hoist & Derrick Co., St.
 Paul, Minn. (Rep. Twin City Chapter and Gray Iron-Steel)
- Vaughan C. Reid, Vice-Pres., City Pattern Foundry & Machine Co., Detroit (Rep. Detroit Chapter and Patternmaking-Supplies)
- S. D. Russell, Pres., Phoenix Iron Works Co., Oakland, Calif. (Rep. Northern California Chapter and Gray Iron)
- George F. Watson, Asst. Chief Chem., American Brake Shoe Co., Mahwah, N. J. (Rep. Metropolitan Chapter and Gray Iron)

President Woody announced that the next meeting of the Board of Directors would be held in White Sulphur Springs, West Virginia, at The Greenbrier, on the dates of Saturday and Sunday, September 30 and October 1, respectively.

There being no further business to come before the Board, the meeting was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:

WALTON L. WOODY, President

Minutes Meeting of the A. F. S. Board of Directors

White Sulphur Springs, West Va .- September 30, 1950

Present: President Walton L. Woody, presiding Vice-President Walter L. Seelbach

Directors:

(Terms exp. 1951) (Terms exp. 1952)
Robert Gregg T. E. Eagan
E. W. Horlebein L. C. Farquhar, Victor E. Zang V. J. Sedlon
F. G. Sefing F. C. C. Troy

Secretary-Treasurer Wm. W. Maloney

Absent: Directors T. H. Benners, Jr.; N. J. Dunbeck; M. J. O'Brien, Jr.; J. O. Ostergren; L. D. Wright.

Reading of Minutes—A quorum of the Board having been established, the Minutes of the July 28 meeting of the Board of Directors were read in full and, on motion duly made, seconded and carried, accepted.

Membership

The Secretary reported that the membership in the Society on August 31, 1950, was 8887, representing a net loss of 159 from the total of 9046 members on June 30. Total circulation of AMERICAN FOUNDRYMAN on August 31 was 9253. The Secretary

called attention to the need for strong membership activities by the Chapters so as to overcome the losses of the past twelve months. He suggested that the 'ociety should shoot at a mark of 10,000 members by June 30, 1951, and the Board, without motion, approved this suggestion.

A number of suggestions then were made by the Directors for more aggressive membership work. The secretary was requested to give each Director a history of Chapter memberships for June 30, 1949 and 1948. It was also suggested that the Chapters be notified of the free distribution of AMERICAN FOUNDRYMAN to selected foundries for follow-up purposes. The Directors were asked to concentrate on the importance of membership work on all Chapter visits.

Report on Finances

The Treasurer reported on finances of the Society as of August 31, 1950. He stated that Income for the first two months of the current fiscal year totaled \$146,241.09, or 33 per cent of the year's budget of \$442,730. He reported that expenses for the same period totaled \$99,942.86, showing an excess income of \$46,298.23 for the months of July and August.

On motion made, seconded and carried the report of the Treasurer was approved.

D

A O C be m

te

D

th

N

A.F.S. Building Project

The Secretary reported that contributions and pledges to the Building Fund as of September 25 totaled approximately \$35,000, of which some 50 per cent represented the pledges of individual members.

President Woody then requested authorization to appoint an Alumni Committee on finances and location, and this authorization was given on motion duly made, seconded and carried.

Safety & Hygiene

President Woody asked Vice-President Seelbach, as Chairman of the Safety & Hygiene Committee of the Board, to present the recommendations of his Committee for financing the A.F.S. Program. Chairman Seelbach stated that his Committee believes it highly essential that A.F.S. issue a statement on the scope of its safety and hygiene and air pollution work. The following recommendation was presented for approval:

It is the recommendation of this Committee that the Safety and Hygiene and Air Pollution Program of A.F.S. be devoted to research and the accumulation of engineering data for the development of recommended practices in the field of Foundry Safety, Hygiene and Air Pollution.

On motion duly made, seconded and carried the Committee's recommendation was approved as a statement of A.F.S. policy.

Chairman Seelbach then stated that his Committee had considered a number of suggestions for solicitation of Safety & Hygiene Funds, and recommended the following basis of solicitation:

(1) Ferrous foundries, 1/4¢ per ton per year\$18,000

(2) Nonferrous foundries, ½¢ per ton per year ... 6,900 (c) Equipment and supplies firms, 50¢ per \$1,000.00 foundry sales per year 12,500

TOTAL - \$37,400

Considerable discussion followed, including the suggestion that a minimum of approximately \$10.00 per Company per year be established in soliciting funds.

It was also felt that the funds for the Program should be solicited on the basis of a lump sum of at least \$350,000 for a maximum period of ten years, contributions to be payable in one, two or three years.

In final analysis and without vote, it was understood that the Board Committee on Safety & Hygiene (Chairman Walter L. Seelbach, F. W. Shipley and James Thomson) would be responsible for setting up and administering the final basis for solicitation.

The Minutes of the Safety & Hygiene Committee meeting held December 9, 1949, were read in full, and it was suggested that these Minutes should accompany any material of solicitation. It was also understood that the Board's statement on the scope of the program should accompany any solicitation.

President Woody emphasized the importance of obtaining the support of the various members of trade associations and read a letter he proposed to send to those various groups. On motion by Director Horlebein, seconded by Vice-President Seelbach and carried, the President was authorized to present the suggested letter to all members of the National Castings Council, accompanied by (a) Minutes of Safety & Hygiene meeting of December 9, 1949, (b) Minutes of A.F.S. Board meeting of January 1950, and (c) September 1950 recommendations of Safety & Hygiene Committee of A.F.S. Board.

Board Research Committee

The President announced that the Chairman of the Board Research Committee, Director Dunbeck, had found it impossible to be present and that no report of this Committee could be made at the time.

Cupola Research Program

The Secretary stated that the Cupola Research Committee met in Chicago in September but was unable to agree on a recommendation to the Board. He reported that the Cupola Research worker, Mr. Giszczak, would prepare a complete progress report on Cupola Research activities prior to November 1, together with personal recommendations as to future Committee work. The Cupola Research Committee then expects to review the report and recommend to the Board future activities of the project.

Without vote the President tabled further discussion on Cupola Research to the next meeting of the Board of Directors, in January.

Report on President's Western Chapter Visits—President Woody reported briefly on visits with officers and directors of the Five Pacific Coast Chapters and expressed the opinion that the Chapters were all active and interested in advancing the Society and the local Foundry Industry.

Status of Members Entering Armed Forces

The Secretary presented for discussion the question of the status of members entering the Armed Services and pointed out that, at the beginning of World War II, the A.F.S. Board had voted to waive "for duration" the dues of such members called into active service. On motion duly made, seconded and carried, the Board declared that the dues of members entering the Armed Services hereafter would be waived on the same basis as in World War II, and the Secretary was instructed to publicize this action at the earliest possible date.

Gray Iron Division Research

The Secretary read a letter from Technical Director Massari, reporting on the activities of the Gray Iron Division Research Committee to date and recommending that the Board approve continuation of the current research of Gating and Risering of Gray Iron Casings for one year. The Secretary pointed out that the funds for continuing the project were in the budget for 1950-1951, approved by the Board of Directors July 28.

On motion by Director Horlebein, seconded by Director Sefing and carried, the Gray Iron Research Project was approved for another year with the recommendation that any future requests for research project approvals be first submitted to the Board Research Committee for recommendations.

Board Policy

Director Sefing requested that the work of setting down definite Board Policies be further advanced by listing major A.F.S. problems for study to the end of formulating policies subject to Board approval. It was agreed that Vice-President Seelbach should be added to the Board Policy Steering Committee for purposes of continuity, and President Woody also requested that Director Sefing be added to the Committee. Director Horlebein agreed to serve as Chairman during 1950-1951. The other members of the Committee, as previously, to be President Woody and Director Dunbeck.

Appointment of Committees

The following Committees of the Board were announced by President Woody:

National Membership Committee

Chairman W. L. Seelbach

All Chapter Membership Chairmen also serving

Chapter Contacts Committee

Chairman W. L. Seelbach

All A.F.S. Directors also serving

By-Laws Committee

Chairman H. Bornstein

N. J. Dunbeck

F. W. Shipley

L. C. Wilson

J. A. Wotherspoon

D. C. Zupge

Secretary Wm. W. Maloney, ex officio

Annual Lecture Committee

Chairman Gosta Vennerholm

H. Bornstein

H. Bornsto

C. F. Joseph

Fred J. Walls R. E. Ward

International Relations Committee

(Still to be appointed in expanded form)

Auditing Committee

(Not appointed)

Publications Committee

Chairman H. M. St. John

C. H. Lorig

NS

ect.

on

ors,

ent

of

hat

the

the

out

nad

led

ed,

the

as

ari, rch

of

hat

ing

sts

ard

efi-

S.S.

to

ch

for

ed

le-

er

dy

by

W. D. McMillan H. J. Rowe

Fred J. Walls
Finance Committee

Chairman Walton L. Woody

E. W. Horlebein

Walter L. Seelbach

Research Committee

Chairman N. J. Dunbeck

L. C. Farquhar

V. E. Zang

Board Safety & Hygiene & Air Pollution Committee

Chairman W. L. Seelbach

F. W. Shipley

James Thomson

Board Policy Steering Committee

Chairman E. W. Horlebein

Walton L. Woody

Walter L. Seelbach

N. J. Dunbeck F. G. Sefing

Executive Committee (as elected by the Board)

President Walton L. Woody, Chairman

Vice-President W. L. Seelbach

E. W. Horlebein

N. J. Dunbeck

F. G. Sefing V. E. Zang

The President directed the Secretary to suggest promptly a date for the next meeting of the Board of Directors in January.

There being no further business to come before the Board, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY

Secretary-Treasurer

Approved:

WALTON L. WOODY, President

Minutes

Meeting of the A. F. S. Board of Directors

Blackstone Hotel, Chicago, Ill.—January 25-26, 1951

James Thomson

Present: President Walton L. Woody, presiding Vice-President Walter L. Seelbach

Directors:

(Terms exp. 1951) (Terms exp. 1952) (Terms exp. 1953)

T. H. Benners, Jr. T. E. Eagan J. J. McFadyen N. J. Dunbeck L. C. Farquhar Frank W. Shipley

Robert Gregg V. J. Sedlon

E. W. Horlebein F. G. Sefing M. J. O'Brien, Jr. L. D. Wright

Victor E. Zang

Secretary-Treasurer Wm. W. Maloney

Technical Director S. C. Massari.

Absent: Directors J. O. Ostergren and E. C. Troy

Reading of Minutes—A quorum of the Board having been established, the Minutes of the September 30, 1950, meeting of the Board of Directors were read in full, and on motion duly made, second and carried, accepted.

Report of the Secretary .

(a) Membership—The Secretary reported that membership in the Society as of December 31, 1950, was 9089, representing a net gain of 43 over the total of 9046 members on June 30, 1950. He reported a net loss of members during the month of July, August and September, but a net gain during the months of October, November, and December. It was indicated that all Chapters have been urged to devote considerable time to membership matters in the hopes of bringing the total of 10,000 members by June 30, 1951.

The Secretary also stated that a list of Company and Sustaining members resigned during 1949 had been furnished the Chapters along with the suggestion that the list be solicited for reinstatement.

(b) Chap ers.—The Secretary stated that no new Chapters, regular or Student, were organized during the period July 1-December 31, 1950, but that a new Student Chapter is in formation at Northwestern University, Evanston, Ill., with several other colleges and universities expressing an interest in similar groups.

(c) 1951 Convention—The Secretary reported on progress of the 1951 Convention scheduled at Buffalo, N. Y., April 23-26, and commented on various phases of the program. He stated that the Western New York Chapter was cooperating closely with the National Office Staff in the matter of local Convention Committees and that applications for hotel accommodations had been mailed to the entire membership January 12.

(d) Defense Service—The Secretary recommended that a committee of the Board be appointed to meet with the Staff at an early date to study and develop future defense service activities of the Society. While no action was taken, Board discussion indicated active interest in the subject.

The Report of the Secretary on motion duly made, seconded and carried was accepted and is made a part of these Minutes.

Report of the Treasurer

The Treasurer reported that Income for the first six months of the current fiscal year, July 1-December 31, 1950, totaled \$245,441.28, and that Expense totaled \$210,546.74, against a budget of \$442,730 for the fiscal year. Income and Expense Statement presented showed that the Income to December 31 exceeded forecast by some \$15,000, that Expense was some \$8,000 below the forecast, and that Excess Income earned for the period was \$34,894.54, against a forecast of \$11,873.50. Individual items of the Income and Expense budget were examined in detail with no recommendations resulting.

The Treasurer's Report called attention to a balance of \$192,-574.53 cash as of December 31, 1950, including \$18,103.34 held in three Canadian banks of deposit. It was pointed out that this amount of cash was in excess of cash requirements for the period January 1, 1951, to December 31, 1951, following which date exhibit space rentals could be expected from the 1952

International Foundry Congress.

Individual items of the Balance Sheet were then examined, with no recommendations resulting.

The Treasurer also presented an Operational Form of statement for the first six months of the current fiscal year, showing an actual Excess Income of \$33,549.22, against a previous forecast of \$456.89 Excess Expense. Accordingly, it was concluded that the Society's finances as of December 31, 1950, were proceeding as expected for a non-exhibit year, with Income slightly exceeding estimates, Expense running somewhat below estimates, and controllable items of Expense under control.

In addition, the Treasurer presented for the record a statement of Publications Sales for the six months ending December 31, 1950, a statement of Official Travel and Expense, and a detailed breakdown of Acct. 27.1 Board and Executive Committee Travel and Expense, prepared at the request of the President, and also a statement of Income and Expense as of November 30, 1950.

The Report of the Treasurer on motion duly made, seconded and carried was accepted and is made a part of these Minutes.

M

thi

tio

Ch

Sec

vis

bet

and

Ch

tha

pro

the

tha

wit

rec

mi

tha

lan

Ch

he

Me

Pol

icy;

the

Bos

to

Boa

P

Cor

mee

bee

tee

pro

the

tha

pro

mig

of p

Ruf

solie

that

for

Pro

botl

fron

subs

M

ber

seco

app

(1

D

Report of the Technical Director

The Technical Director presented a complete report covering activities of technical committees, A.F.S. research, the 1951 Convention program, publications, the annual apprentice contests and other activities. The report detailed progress on the nine research projects of the Society without recommendations.

The Technical Director stated that the Program and Papers Committees of the various divisions and General Interest Committees were carrying out their assigned jobs for presentation of an excellent program at the 1951 Convention. This program included an all-day session on Tuesday, April 24, for a Symposium on Gating and Risering of Castings, jointly sponsored by five of the Society's technical divisions.

It was pointed out that a total of 125 entries had been received (as of Jan. 25) in the 1951 apprentice contests, including 44 in gray iron molding, one in steel molding, four in non-ferrous molding, and 58 in patternmaking. The Technical Director stated that national apprentice awards would be made at the time of the Annual Business Meeting in Buffalo on April 25 and that first-prize winners in each of the contests would be brought to the Convention by A.F.S. to receive such awards in person.

The Technical Director announced publication and distribution of the 1950 annual Transactions (vol. 58) comprising 760 pages. In the matter of special publications, it was pointed out that three publications had been completed, three were in the final stage, and three still to be completed.

The Report of the Technical Director on motion duly made, seconded and carried was accepted and is made a part of these Minutes.

Recommendations of the Finance Committee

Minutes of the meeting of the Finance Committee held January 12, 1951, were read in full and on motion of Director Farquhar, seconded by Director Eagan, and carried, were accepted for the record.

Considerable discussion arose over the recommendation of the Finance Committee that some \$110,000 of General Fund cash as of December 31, 1950, be invested only in U. S. Government Bonds and that existing cash in the A.F.S. Building Fund be similarly invested. Some Directors suggested the advisability of purchasing preferred stocks and non-speculative common stocks as a safeguard against future inflation and consequent further depreciation of the dollar.

Members of the Finance Committee expressed the view that they were unanimously disinclined to risk Society funds in view of the present high market for industrial stocks, and that they were unanimously disinclined to recommend the purchase and resale of stocks as a revenue producing action. It was the concensus that future investments of Society funds be made a matter of Board policy, following full discussion by an appointive Board Policy Committee. On motion by Director Eagan, seconded by Director O'Brien, and unanimously carried, the Board accepted the recommendation of the Finance Committee that for the time being surplus General Fund and Building Fund Cash be invested only in U. S. Government Bonds.

Report on American Foundryman

The Secretary presented a detailed report on the AMERICAN FOUNDRYMAN magazine, showing a gross billing of \$51,583 for the six months ending December 31, 1950, an increase of \$4,955.77 over gross billing of \$46,627.23 for the same period in 1949-50. The report stated that since the appointment of advertising representatives in March 1950, greater representation and constant contacts with advertisers and agencies had materially improved the advertising outlook, with over 415 pages of advertising under contract as of December 31, 1950, compared with a total of 425 pages run during the entire fiscal year 1949-50.

The Secretary reported that in keeping with the advertising policy voted by the Board of Directors, numerous steps had been taken to supplement the efforts of advertising representatives. He reported that the January 1951 and February 1951 issues would exceed any normal or non-Convention issues published since advertising was first accepted in May 1945, and he therefore recommended continuation of present advertising representation for the time being.

The report included a comparison of advertising rates in

foundry publications, 1945-1951, and expressed the thought that in view of the probability of increasing production costs during the next several years, consideration should be given the matter of advertising rates. Accordingly, the Secretary recommended that the basic black-and-white single-time page rate for advertising in American Foundryman be increased from the present \$250 to \$280, effective July 1, 1951, with adequate protection provided advertisers of record on that date. On motion by Director Horlebein, seconded by Director Shipley, and unanimously carried, the recommendation was approved.

At the same time the President suggested that advertising rates again be reviewed at the meeting of the Finance Committee to be held in June or July 1951 for budget-making purposes. The report on American Foundryman advertising is made a part of these Minutes.

Report on National Castings Council

Director Horlebein, as President of the National Castings Council, was asked to report on activities of the Council prior to his early departure from the Board meeting. President Horlebein stated that no meeting of the Council had been held during the current fiscal year, the next meeting being scheduled in Cleveland on February 2, 1951, and that a report would be presented on that meeting at a later date.

Report of Research Committee of the Board

Director Dunbeck, as Chairman of the Research Committee of the Board, presented the minutes of a meeting of his Committee held November 6, 1950. These minutes, previously placed in the hands of all members of the Board of Directors, embodied as "Recommendations" certain understandings of the Research Committee with regard to its functions and accepted obligations. No action was taken by the Board on these specific "Recommendations," but it was the consensus that they represented a proper understanding of the Board's desires.

Chairman Dunbeck called attention to the Board resolution

Chairman Dunbeck called attention to the Board resolution of June 26, 1946, and stated that that resolution, in the opinion of his Committee, provided "sufficient protection for the conception, planning, direction, and reporting of A.F.S. divisional projects." During ensuing discussion the several members of the Board clarified to the apparent satisfaction of all that so long as future research projects were under the joint direction of the Technical Director and the various divisional research committees, and subject to review by the Research Committee of the Board, the interests of the Society would be protected.

On motion by Director Shipley, seconded by Director O'Brien and carried, the minutes of November 6, 1950, meeting of the Research Committee of the Board were accepted.

Report of Safety & Hygiene & Air Pollution Committee of the Board

Vice-President Seelbach, as Chairman of the Safety & Hygiene & Air Pollution Committee of the Board, stated that his Committee had held no meetings since the last meeting following the September 30, 1950, meeting of the Board, and that little progress had been made in establishing the program since that date for certain acceptable reasons. He then presented a suggested timetable of activity for establishing the Society's Safety & Hygiene & Air Pollution Program during the year 1951.

Considerable discussion ensued on whether the timetable should call for solicitation of funds prior to the hiring of a staff program director. President Woody urged that the program be gotten under way during the current fiscal year, and after full discussion motion was made by Director Farquhar, seconded by Director Dunbeck and unanimously carried.

That the A.F.S. Secretary and the Safety & Hygiene & Air Pollution Committee be authorized to take such steps as may be necessary to activate the program with General Funds of the Society, to a maximum use of \$20,000, such sums expended to be re-imbursed from funds solicited and obtained for the Safety & Hygiene & Air Pollution Program at a later date.

(Following the Board meeting, Chairman Seelbach's Committee met and revised the timetable of activity in view of the above motion. Minutes of this meeting are appended and are made a part of the Board Minutes.) ng

er

ed

r-

nt

0-

OT

T-

ng

ee

in

1e

e-

d

in

as

h

S.

1-

a

n

n

ie

18

ie

n

at

ξ-

y

e fE

11

y

ir

15 ıl

h

d

n

e

Report of Chapter Contacts Committee

Vice-President Seelbach, as Chairman of the Chapter Contacts Committee, reported that Officers and Directors of the Society thus far have visited 25 of the regular Chapters and that additional visits were being scheduled to all except the Mexico Chapter during the balance of the current fiscal year. The Secretary reported that in addition members of the Staff have visited 17 Chapters and also were planning on additional visits between now (Jan. 25) and June.

Chairman Seelbach thanked the Directors in their interests and the time devoted to maintaining closer relations with the Chapters and requested all to complete their visits to those Chapters which they had agreed to contact. The Secretary asked that all give particular attention to the matter of membership promotion in their visits toward the end of materially increasing the membership during the next five months. It was agreed that the Secretary would prepare and provide each Director with a list of past and future Chapter visitations as a matter of record and reminder.

Report of the Board Policy Steering Committee

Minutes of the meeting of the Board Policy Steering Committee held January 12, 1951, were read and it was announced that the Committee again would meet on February 2 in Cleveland for a discussion of specific subjects. Director Horlebein, as Chairman of the Committee stated that sub-committees would be appointed to develop policies on the following subjects: Membership and Chapter Policy, Technical Policy, Educational Policy, Convention and Exhibit Policy, and Administrative Policy; Policy on AMERICAN FOUNDRYMAN and publications to be developed by the Publications Committee; Financial Policy to be developed by the Finance Committee.

During the discussion that followed it was the concensus of the Board that as policies are developed and approved by the Board, it must be understood that such policies shall be adhered to until such time as a revision thereof is approved by the Board of Directors.

Report on A.F.S. Building Fund

Past President R. J. Teetor, as Chairman of the A.F.S. Housing Committee, was invited to personally present his report at the meeting on January 26, and announced that over \$72,000 had been contributed as of that date. He reported on results obtained from solicitation by 16 members of the Housing Committee and stated that this Committee would enlarge its list of prospects during the coming months, as well as following up on the original list of 200 company names provided. He suggested that an early meeting of the Housing Committee might be advisable to review the Program in the interest of further progress, and offered the thought that a smaller Committee might prove advisable.

During the discussion that followed, it was felt that a number of prospects might be reached during the 1951 Convention in Buffalo. It was agreed without vote that general industry-wide solicitation for the Building Fund should end May 1 in order that the Program might not drag out and in order to make way for solicitation of funds for the Safety & Hygiene & Air Pollution Program in the early Fall. The Chairman and the Secretary both were urged to seek a large volume of small contributions from a larger number of contributors, rather than to seek very substantial amounts from a few.

Recommendations of Board of Awards

Minutes of the meeting of the Board of Awards held December 14, 1950, were read in full and on motions duly made, seconded and carried, the following award recommendations were approved by the Board of Directors:

(a) THAT the Peter L. Simpson Gold Medal of A.F.S. be awarded to Thomas W. Curry, Lynchburg Foundry Co., Lynchburg, Va., "For his outstanding work and contributions in the field of sand technology in the foundry in-

(b) THAT the John A. Penton Gold Medal of A.F.S. be awarded to Victor A. Crosby, Climax Molybdenum Co., Detroit, Mich., "For outstanding service to the Society and for his contributions to the dissemination of information relating to ferrous foundry metallurgy."

(c) That the John H. Whiting Gold Medal of A.F.S. be awarded to Alfred Boyles, United States Pipe & Foundry Co., Burlington, . J., "For fundamental studies on the mechanism of graphitization of gray cast iron.'

(d) That Honorary Life Membership be awarded to E W. Beach (retired) formerly of Campbell, Wyant & Cannon Foundry Co., Muskegon, Mich., "For outstanding contributions to the Society and to the industry in the field of foundry engineering."

(e) THAT Honorary Life Membership be awarded to E. J. McAfee, Puget Sound Naval Shipyard, Bremerton, Wash., "For outstanding service to the Society and to the foundry industry in the development of pattern-making materials and applications."

Some dissatisfaction was expressed by the Board over that part of the Board minutes involving a recommendation to the next Board of Awards for certain types of recognition. Accordingly motion was made by Director Eagan, seconded by Director Sefing and carried that the Board of Directors desires to go on record with the Board of Awards that it does not believe it a good policy for one Board of Awards to make recommendations to a future similar body.

Report of the 1950-51 Nominating Committee

For information only, the report of the 1950-51 Nominating Committee, which requires no approval by the Board of Directors under the by-laws, was presented, naming the following nominees:

For President to serve one year

Walter L. Seelbach

For Vice-President to serve one year

I. R. Wagner, Director

Electric Steel Castings Co., Indianapolis, Ind.

For Directors—each to serve three years
M. J. O'Brien, Wks. Mgr., Symington-Gould Corp., DePew, N. Y .- representing Steel and Western New York (for re-

H. W. Dietert, Pres., H. W. Dietert Co., Detroit-representing Suppliers and Michigan

A. L. Hunt, Wks. Mgr., National Bearing Div., American Brake Shoe Co., St. Louis-representing Non-Ferrous and Missouri

Dr. J. T. MacKenzie, Technical Dir., American Cast Iron Pipe Co., Birmingham, Ala.-representing Gray Iron-Steel and South.

A. M. Ondreyco, Plant Mgr., Vulcan Foundry Co., Oakland, Calif.-representing Gray Iron and Pacific Coast

For Director to serve one year
President Walton L. Woody on completion of his present term of office as prescribed in the A.F.S. by laws.

Future Convention Cities

The Secretary requested discussion on the selection of Convention cities for future years beyond 1952, stating that it was his recommendation that a decision be reached on the 1953 Convention city. He reported the interest of Eastern Canada and Texas Chapters in 1951 and stated that an official invitation from the Chicago Chapter had likewise been received.

Some discussion ensued on the advisability of holding nonexhibit Conventions in cities other than major foundry industrial centers. It was pointed out that a Chicago Convention probably would be an economical event and that it was desirable to obtain hotel guarantees for 1953 at the earliest possible date.

Future Board Meeting-President Woody stated that he did not feel it necessary to hold another Board meeting prior to the Buffalo Convention, April 23-26, but stated his intention of calling a meeting of the Executive Committee during the Convention on Tuesday, April 24. It was stated that a meeting of the Board of Directors might be called early in June and that Directors would be notified well in advance of such an event.

There being no further business to be presented, the meeting was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:

WALTON L. WOODY, President

A

LU

in

w cr te

ol

th

116

Cl

w

to

8t

25

ga

to

Co

Co

ter

for

H

cei

off

me

tio

int

gre

Pre

Int

Fo

ani

Ch

MA

ady

ful

the

Cir equ adv

195 buc

tion pla

ship

of a

Pro

Dir

omi

rece

as s

cert

and

nex

of \$

all

100

had

be b

B

Minutes Meeting of the Executive Committee

Hotel Statler, Buffalo, N. Y .- April 24, 1951

Present: President Walton L. Woody, presiding

Vice-President Walter L. Seelbach

Directors: N. J. Dunbeck
E. W. Horlebein
F. G. Sefing

V. E. Zang Secretary-Treasurer Wm. W. Maloney

Technical Director S. C. Massari

The Minutes of the meeting of the Board of Directors of January 25-26, 1951, were read in excerpt and accepted, having previously been approved by Letter Ballot of the Board of Directors.

President Woody again emphasized the desirability of greater guidance by the Board of Directors and National Office in the organizing and staging of regional conferences by the Chapters. The Technical Director described the development of taperecorded programs by the Staff, which can be used to provide emergency programs when speakers cancel at the last minute, and which can bring to the Chapters a practical and short presentation of Convention papers.

The Secretary stated that, in line with the President's suggestion, the Staff had proposed to the Northern California Chapter the organizing of a regional conference in Oakland next January, and had proposed to the Quad City Chapter the reinstatement of the Quad City Regional Conference this Fall. In each case, it is intended that the National Office will play a considerable part in the organization and handling of the programs.

The Executive Committee approved this step, and on motion by Director Horlebein, seconded by Director Sefing and carried, recommended that the National Office be authorized to work with the Chapters in organizing regional conferences in 1951-52, and that the proposal be presented at the 1951 Chapter Officers Conference, so as to obtain full approval by the Chapters.

In connection with the development of tape-recorded programs and similar material, President Woody expressed the thought that several thousand dollars should be set up for expenditures of this type without the requirement for formal Board action. On motion by Director Horlebein, seconded by Vice-President Seelbach and carried, the purchase of tape-

recording equipment and the arrangement of tape-recorded programs was approved and recommended for Board action at the July Board meeting.

The Secretary presented a Report on Membership as of March 31, 1951, and emphasized the point that, as included in the 1950-51 Budget, some 1,000 copies of American Foundryman were being distributed monthly to selected non-member foundries in order to induce bona fide circulation of American Foundryman and to suggest eventual membership by those receiving the magazine. Director Horlebein suggested that the mailing list for such copies be checked with the Chapters so as to obtain better lists of names in some cases.

The Secretary-Treasurer presented a statement of finances as of March 31, 1951 which, after detailed examination of Income and Expense, was accepted.

Some discussion took place for more aggressive development of the Safety & Hygiene and Air Pollution Program, particularly with regard to the selection and hiring of a Staff man as previously approved by the Board. Vice-President Seelbach urged more intensive work on the program, particularly toward formation of a solicitation committee and development of a solicitation brochure. The Staff pointed out that concentration on the Building Program and 1951 Convention had prevented further developments by the Staff on the Safety & Hygiene Program.

The Secretary pointed out that, in accord with the By-Laws of the Society, official auditors should be appointed by the President to perform the audit of the Society's financial books as of June 30, 1951. President Woody appointed George Rountree & Co., Chicago, as the official auditors for the current fiscal year.

The President reported on the Building Fund to date and stated that the Fund totaled over \$125,000 as of April 24, well beyond the original goal of \$100,000. He stated it was the intention to cease general solicitation on May 31 and to pursue the matter of selection of an architect, and determination of where to build at an early date through the A.F.S. Housing Committee.

There being no further business to be presented, the meeting was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Minutes Annual Meeting 1950-1951 Board of Directors

The Homestead, Hot Springs, Va .- Thursday, July 26, 1951

Present: President Walton L. Woody, presiding Vice-President Walter L. Seelbach

Directors:

(Terms exp. 1951) (Terms exp. 1952) (Terms exp. 1953)
T. H. Benners, Jr. T. E. Eagan
N. J. Dunbeck L. C. Farquhar
Robert Gregg V. J. Sedlon James Thomson
M. J. O'Brien, Jr. F. G. Sefing E. C. Troy

V. E. Zang

Secretary-Treasurer Wm. W. Maloney

Technical Director S. C. Massari

Absent: Directors E. W. Horlebein, J. J. McFadyen and L. D. Wright

Present (as observers):

Vice-President-elect I. R. Wagner
Directors-elect: H. W. Dietert
J. T. MacKenzie
A. M. Ondreyco

Absent: Director-elect A. L. Hunt

Reading of Minutes—The minutes of the Executive Committee held April 24, 1951, at Buffalo, N. Y., were read and on motion duly made, seconded and carried, were approved.

The Secretary was requested to send a copy of the Chapter Manual to all hold-over and new Directors who did not attend the Chapter Officers Conference in June.

Reports of Officers and Board Committees

Report of the Secretary

The Annual report of the Secretary covered the following major activities: Membership, Chapters, Convention, International Foundry Congress, American Foundryman, Technical Activities, Publications, and General Administration.

Membership—The Secretary pointed out that membership in the Society as of June 30, 1951, totaled 9.340, representing a net gain of 294, or approximately 3.2 per cent over the total of 9,046 on June 30, 1950. He stated that 2,316 members were added during the year, and 2.022 members lost for all causes, or a turnover rate of 21.6 per cent. The Secretary stated that con-

siderable effort should be given toward reduction of this rate of turnover.

The Secretary emphasized the importance of membership work during 1951-52 and made seven specific recommendations involving the cooperation of the Chapter Contacts Committee with the Chapters in bringing about a material membership increase. He stated that targets again would be set for each Chapter toward a goal of 10,000 members by June 30, 1951, pointing out that only 8 Chapters made their targets during 1950-51 and that of the total goal of 985 net additional members, the actual net gain was 294, or approximately 30 per cent of the total.

The Secretary reported the addition of three new Student Chapters during 1950-51, at the University of Alabama, Northwestern University, and Pennsylvania State College, bringing the total of Student Chapters to 11. He also reported a successful 8th Annual Chapter Officers Conference held in Chicago, June 25-26, with a total attendance of 101, including 83 Chapter delegates from all 40 Chapters, plus 14 National Officers and Directors. Both attendance figures represented new highs for the Conferences.

Convention—The Secretary reported that the 1951 non-exhibit Convention in Buffalo, April 23-26, had developed a total registered attendance of 2,816, and that the registration fees of \$2.00 for members and \$5.00 for non-members was generally accepted. He stated that the technical program was excellent and well received, that many visitors felt the Convention activities were smoothly run, and that the two special sessions addressed by officials of NPA were both sellouts.

International Foundry Congress—The Secretary reported on his trip to Europe in May 1951 and stated that European foundrymen were enthusiastically planning to attend the 1952 International Foundry Congress in Atlantic City, and that many were interested in a two-week pre-Convention tour prior to the Congress. He stated that arrangements were being made for such a tour with the IBF and Thomas Cook & Son, and reported that President-elect and Mrs. W. L. Seelbach would be attending the International Congress at Brussels in September 1951.

It was announced that the slogan for the International Foundry Congress would be "EVERY Foundry in '52" and that announcements pertaining to the event would be made at all Chapters during the year.

American Foundryman—Reporting on American Foundryman, the Secretary stated that the first full year in soliciting advertising through publishers representatives had been successful in exceeding the advertising income budget for the year, that the Society's membership is now audited by the Audit Bureau of Circulation, that closer contacts are being maintained with the equipment and supplies associations, and that new and increased advertising rates became effective July 1, 1951. He pointed out that American Foundryman advertising is being solicited on a three-year budget covering the period July 1, 1950, to June 30, 1953, and every effort would be made to meet the approved budget.

Editorially, the Secretary stated that acceptance of the publication was constantly growing among the membership and that plans for the coming year included the development of more practical articles of greater interest to the bulk of the membership.

Technical Activities—The Secretary reported the acquirement of a new full-time staff man to head up the Safety & Hygiene Program, namely, Kenneth M. Morse, and stated that he would be working under the direct supervision of the Technical Director.

Publications—On publications, the Secretary stated that recommendations of the Publications Committee had not been received in time for the Board meeting, but would be submitted as soon as available. He pointed out, however, the necessity for certain staff additions to take care of a growing backlog of new and revised publications scheduled for publication within the next two years.

Building Fund—The Secretary's report announced the total of \$140,000 contributed to the Building Fund to date, of which all but \$38,000 is in cash contributions. He pointed out that 100 per cent of the regular Chapters and two Student Chapters had contributed, as well as 375 individuals and 379 companies.

The report of the Secretary, on motion duly made, seconded and carried, was accepted and is made a part of these minutes. It was suggested that the Secretary's recommendations should be brought to the attention of the new Board more specifically. In response to question, the Secretary stated that publicity was given to other foundry groups in American Foundryman when received, and it was suggested that the question of publicity be discussed with other nearby Chapters before broadening such publicity. It was suggested that clippings of articles published in American Foundryman be sent to the presidents of those companies from whose employees articles may emanate.

Report of the Treasurer

The Treasurer reported that the non-exhibit year of 1950-51 showed a total Income of \$382,923 and total Expense of \$431,353, or an excess of \$48,429 Expense over Income. This, however, was well below the budgeted \$72,230 Excess Expense for the year. Income per member was \$41.00 against \$32.90 in 1948-49, and Expense per member was \$46.20 against \$39.90 in 1948-49.

The report indicated net worth of the Society (Fund Principals) as of June 30, 1951, was \$406,811, including all Award and Building Funds, against \$320,701 in 1950, \$199,930 in 1949, and \$275,122 in 1948. The Treasurer reported the Society's finances in satisfactory shape with Expense under close control of the Finance Committee and Treasurer.

The official audited report of finances for the fiscal year as prepared by George V. Rountree & Co., Chicago, was presented and discussed in detail. It was the concensus of the meeting that this was the most comprehensive audit ever presented before the present Board of Directors, and the Treasurer was instructed to commend the auditors. The Treasurer also went over in detail the various items of Income & Expense for the year, pointing out major variations from the budget.

The Treasurer also reported that the bank balance for the Cupola Research Fund as of June 30, 1951, was \$8,525.96, including \$3,525.96 in cash and \$5,000 in investments. The Treasurer's report is made a part of these Minutes.

The report of the Treasurer on motion duly made, seconded and carried, was accepted and the Staff commended for maintaining a low differential between Income and Expense in a nonexhibit year.

Report of the Technical Director

The Technical Director presented a detailed report covering technical committees, A.F.S. sponsored research projects, 1951 Convention Program, publications of the Society, the 1951 Apprentice Contests and other technical activities.

Recommendations for continuation of research projects were tabled for consideration by the 1951-52 Board.

It was reported that 50 different functions were scheduled during the 1951 Convention, with 90 authors or discussion leaders participating, and that 43 formal papers were published as separate preprints in advance of the meeting. The Technical Director specifically mentioned the Symposium on Principles of Gating session, an all-day function, with a total of 14 authors participating and a remarkable attendance of 1200. He stated that the Symposium would be published as a separate book covering the entire proceedings.

During the fiscal year 1950-51, it was reported that four new publications had been produced and that the foundry high school textbook had been completed and would be published during 1951-52 by John Wiley & Sons. He reported that the Sand Testing Handbook was in the last stages of revision for definite publication during 1951-52; that the Charles Edgar Hoyt Annual Lecture of 1951 by J. C. Zeder, of Chrysler Corp., was available in pamphlet form; that 1500 copies of Transactions, vol. 58 (1950) were published, of which 1122 copies were sold; and that pre-publication order forms for Transactions, vol. 59 (1951) were mailed to the membership in July.

The Technical Director reported a total of 246 entries in the 1951 Apprentice Contests from 76 companies and 8 Chapters, against 267 entries from 98 companies and 10 Chapters the previous year. In the way of Chapter program aids, the Technical Director pointed out that a new Chapter Speakers' List had been distributed, that 81 showings of the color-sound film on "The Principles of Gating" were made during 1950-51, and that a new film was now in the making for first presentation at the International Foundry Congress in 1952. He also stated that a program had been initiated involving the recording of technical subjects on magnetic tape for use by the Chapters, and it was hoped to develop a library of such subjects in the future.

The report of the Technical Director, on motion duly made, seconded and carried, was accepted with commendation, and is

prothe

IONS

950were s in MAN agasuch lists

es as come nent

prerged rmation the

vs of dent June Co.,

and well tenthe here ttee.

mit-

pter

end

ving rnaical

in net of were

con-

made a part of these Minutes.

The Technical Director suggested that the services of stenotype reporters at the Annual Convention might be dispensed with and that each Division might be asked to appoint a technical reporter to prepare a summary of the oral discussion and to obtain additional written discussion for Transactions. Considerable discussion of this suggestion followed, and the Technical Director was requested to contact the Technical Divisions and General Interest Committees of the Society to determine their attitude toward the matter.

Report on International Foundry Congress

Vice-President Seelbach, Vice-President-elect Wagner and Secretary Maloney all reported on aspects of the International Foundry Congress, including a preliminary organization meeting held in Cleveland during April. The Secretary was directed to prepare a short presentation on the International, which could be placed in the hands of all A.F.S. Directors to use in their Chapter contacts work. Vice-President Seelbach stated that he and the Secretary would be contacting ECA in August to determine the extent of cooperation that might be rendered by them.

National Castings Council

President Woody reported that one meeting of the National Castings Council had been held during 1950-51, but that little had been accomplished which could be reported to the Board.

Publications Committee

The report of the Publications Committee meeting, held June 29, not being available, no report was presented.

Finance Committee

Recommendations of the Finance Committee, being embodied in the Annual Report of Society finances for the fiscal year ended June 30, 1951, as brought out in the report of the Treasurer, no special report was presented.

The President pointed out that \$125,000 of the General Reserve Fund and \$65,000 of the Building Fund had been invested in Government securities during the month of May. As Chairman of the Finance Committee, he recommended approval by the Board of Directors, and on motion duly made, seconded and carried, the investments made by the Finance Committee were approved.

Committee on 1952 Exhibits

The Secretary reported that a meeting with the Committee on Exhibits had been held June 15 with Vice-President Seelbach and Vice-President-elect Wagner present. He reported that the rules and regulations for the 1952 Exhibit were read in detail and approved at that time, but that certain questions had been raised by a few exhibitors, necessitating a further meeting in Atlantic City to be held August 13, a which Mr. Seelbach would be present. He reported that the exhibitors had approved practically all recommendations made by the Staff for the rules and regulations and that excellent cooperation was anticipated.

Retirement Plan Trustees

In lieu of a report from the trustees, the Secretary presented a financial report on the Retirement Trust Account, showing that as of July 31, 1951, all funds in the Trust Account were being disbursed to apply against 1951-52 premiums and to reimburse A.F.S. for prepaid employee contributions during 1950-51. He stated that the trustees had approved balancing the Trust account in this manner each year hereafter.

President Woody expressed the suggestion that the trustees

President Woody expressed the suggestion that the trustees might find it desirable to meet once a year and that a report of the trustees should be presented annually to the Annual Board Meeting. The Secretary was directed to bring this suggestion to the attention of the Chairman of the Trustees.

Board Policy Steering Committee

The Secretary reported that a number of Board Policy Steering Committees had been organized, but that no meetings had been held to formulate Board policy. He stated that this matter would be carried forward shortly to the end of developing specific Board Policies for approval by the Board of Directors at an early date.

Housing Committee

President Woody reported for the Housing Committee and stated that a meeting of the Action Sub-Committee had been held in Chicago, July 11, at which time preliminary sketches and floor plans were presented, as well as several architect estimates. The Sub-Committee visited a number of locations around Chicago and authorized the Staff to pursue investigations for the selection of an architect and toward selection of properties for purchase.

Recommendations of the Housing Committee meeting held in Buffalo, April 26, were read in brief, minutes of this meeting having been previously approved by letter ballot to the Board. On motion by Director Eagan, seconded by Director Farquhar, and carried, the Board of Directors approved the specific recommendations of the Housing Committee, including action and recommendations of the Action Sub-Committee.

Research Committee of the Board

A report of the Research Committee of the Board was presented by Chairman N. J. Dunbeck and is made a part of these Minutes. Action on the recommendations for continuation of research projects was deferred to the meeting of the 1951-52 Board, since they involved budgetary procedure.

Safety & Hygiene Committee of the Board

Vice-President Seelbach requested that the discussion on Safety and Hygiene be tabled until the meeting of the new Board, at which time the new Staff Director of Safety & Hygiene, Kenneth M. Morse, would be present.

Chapter Contacts Committee

A list of the Chapter Contacts Committee, assigning each Director to contact one or more of the Chapters, was presented by Vice-President-elect Wagner as Chairman of the Committee. The Directors were requested to make their contacts with the Chapters as early as possible, communicating to the Secretary their intention to do so, so that specific points may be brought to their attention.

Recommendations of Executive Committee

No action taken by the Executive Committee in meeting April 24, 1951, required approval by the Board. President Woody pointed out that by action of the Board, July 1950, minor expenditures below \$1,000 can be made by the Finance Committee without requiring special approval by the Board.

Recommendations of Educational Division

Director Sefing, on behalf of the Educational Division, presented certain recommendations toward increased activity along educational lines. These recommendations included (a) a request for a separate convocation with Chapter Educational Committee Chairmen, (b) a full half-day for educational activities at the next Chapter Officers Conference, (c) that a speaker on educational matters address the next Chapter Officers and Directors dinner at the Annual Convention and that the speaker be selected from a Chapter doing an outstanding job, and (d) that the Chapters be rated on educational activities on some basis that would make a small Chapter competitive with a large one.

The Secretary suggested that the Educational Division be asked to sponsor a session at the 1952 Convention, whereby the Educational Division could discuss with the Educational Committee Chairmen of the Chapters greater participation by the Chapters in educational matters. He stated that the National Office would make every effort to get a maximum number of Educational Committee Chairmen in attendance. Director Sefing indicated approval of this idea.

No action was taken on other recommendations presented. The Secretary indicated that the Educational Division might well concentrate its discussions along secondary school lines. It was the consensus of the Board that this should be done.

FEF Request for Convention Meeting

The Secretary presented a request from the Foundry Educational Foundation for an all-day session on education, under sponsorship of FEF, during the 1952 International Foundry Congress, at some place other than where the technical sessions may 10

d ti fe quit P w g

o co B el oi

11: 12: 1: 1:

3:

in dire

son and tert Pip

Pip Cla R. (Na pres he held

ONS

and

een

and

ates

Chi-

the

for

d in

nav-

On

and

om-

and

of 1-52

fety , at

eth

ted tee. the ary ght

ody ex-

reomties on

be hat asis ne.

be

the

m-

the

nal

of

ing

es.

ca-

ler

ay

It was the consensus of the Board that activities of the nature described should be sponsored by the Educational Division of the Society and the Secretary was directed to express to FEF the feeling of the Board that a precedent might be set if the request were approved, whereby other societies might make similar requests at some future date. It was felt that all activities during the A.F.S. Convention should be sponsored by the Program and Paper Committees of the Divisions and the Committees of A.F.S., with the possible exception of small meetings of other foundry groups whose directors might be attending the Convention.

In closing the meeting, President Woody expressed his gratitude for the constant and fine support given him by the Board of Directors of the Society and emphasized the need for greater concentration on membership. He expressed the hope that the Board of Directors would give their utmost support to Presidentelect Seelbach to the end of working as a team toward progress of the Society and the industry.

President-elect Seelbach responded by commending President Woody for his aggressive and forward-looking administration and expressed the thanks of the Society for his energy and effort given to the job. He stated that he would endeavor to advance the Society to the best of his ability with the cooperaion of the Board.

The 4th Annual Dinner in honor of retiring Directors was served the evening of July 26 at The Homestead. Following the dinner, Secretary Maloney presented color slides of his European trip. Guests of honor included retiring Directors T. H. Benners, Jr., N. J. Dunbeck, Robert Gregg, and V. E. Zang. Retiring Director and Past-President E. W. Horlebein was unable to be present.

There being no further business to come before the meeting, President Woody declared the 1950-51 Board of Directors meeting adjourned, and turned over the gavel of the Chairman to Mr. Seelbach.

> Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:
WALTON L. WOODY, President

Eighth Annual Chapter Officers Conference June 25 - 26, 1951 - LaSalle Hotel - Chicago

PROGRAM

	Monday, June 25	4:35 PM	A.F.S. and Chapter Finances. Secy. Wm. W. Maloney
10:00 AM	Welcome to Chapter Officers, Chairman W. L. Seelbach	6:30 PM	Dinner
10.00 AM	Introductions	7:30 PM	"America Tomorrow"Guest Speaker Carl Taylor
	A.F.S. BackgroundA.F.S. Pres. W. L. Woody		
10:25 AM	Functions of the National Office		Tuesday, June 26
	Secy. Wm. W. Maloney		
10-40 AM	Technical Activities of the A.F.S	9:15 AM	Membership WorkSecy. Wm. W. Maloney
10.10 AM	Tech. Dir. S. C. Massari	10:50 AM	
	Editor H. F. Scobie		Secy. Wm. W. Maloney
11:10 AM	AMERICAN FOUNDRYMAN Adv. Mgr. T. B. Koeller	11:10 AM	A.F.S. Building ProjectPres. W. L. Woody
	Drog W I Woods		Management and Public Relations
11:40 AM	Regional Conferences Secv. Wm. W. Maloney	11.30 AM	
12:00 рм			Pres. W. L. Woody
			Secy. Wm. W. Maloney
1:00 PM	1	10.00	V.PElect I. R. Wagner
1:30 PM	Program BuildingTech. Dir. S. C. Massari		Luncheon
3:00 PM	Chapter Educational Activities	1:30 PM	International Foundry Congress
	Prof. G. J. Barker		Secy. Wm. W. Maloney
	Tech. Asst. Jos. E. Foster		Exh. Mgr. A. A. Hilbron
9.50 ===	Chapter Administration Chairman W. L. Seelbach	1:50 PM	The Shakeout
3:30 PM	Chapter Administration Secv. Wm. W. Malonev	3:00 PM	Adjournment of Conference

ATTENDANCE

Directors and Guests

Participating in the 8th Annual Chapter Officers Conference in addition to those mentioned, were the following national directors: T. E. Eagan, Cooper-Bessemer Corp., Grove City, Pa.; L. C. Farquhar, Sr., American Steel Foundries, East St. Louis, Ill.: J. J. McFadyen, Galt Malleable Iron Co., Galt, Ont., Canada; Martin J. O'Brien, Jr. Symington-Gould Corp., Depew, N. Y.; J. O. Ostergren, Lakey Foundry & Machine Co., Muskegon, Mich.: F. G. Sefing, International Nickel Co., New York; Frank W. Shipley, Caterpillar Tractor Co., Peoria, Ill.: James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.; and L. D. Wright, U. S. Radiator Corp., Geneva, N. Y.

Directors-elect present were Harry W. Dietert, Harry W. Dietert Co., Detroit, and J. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala. Guests included FEF Chairman Claude B. Schneible, Claude B. Schneible Co., Detroit, Prof. R. E. Kennedy and Roy W. Schroeder, University of Illinois (Navy Pier), Chicago; C. E. Hoyt, retired A.F.S. executive vice-president; and Wm. G. Gude, Penton Publishing Co., Cleveland.

Chapter Representation

Chapter officers present were chiefly chairmen and program chairmen (usually also vice-chairman). They were:

BIRMINGHAM—Program Chairman Fred K. Brown, Adams Rowe & Norman, (also V. C.), and Secretary-Treasurer John F. Drenning, Kerchner, Marshall & Co., Birmingham, Ala.

British Columbia, Chairman W. M. Armstrong, University of British Columbia, Vancouver, and Program Chairman Herbert Heaton, Heap's Engr. Ltd., New Westminster, B. C., Canada (also V. C.).

CANTON DISTRICT—Chairman C. B. Williams, and Secretary Wendell W. Snodgrass, Massillon Steel Castings Co., Massillon, Ohio. CENTRAL ILLINOIS—Chairman R. E. Dickison, Brass Foundry Co.. Program Chairman G. H. Rockwell (also V. C.) and Secretary.

Treasurer B. L. Bevis, Caterpillar Tractor Co., Peoria, Ill. CENTRAL INDIANA—Chairman Robert Spurgin, Swayne, Robinson & Co., Richmond, Ind., and Vice-Chairman Carl O. Schopp, Link-Belt Co., Indianapolis, Ind.

CENTRAL MICHIGAN—Chairman Thomas T. Lloyd, Albion Malle-

- able Iron Co., Albion, Mich., and Secretary-Treasurer R. K. Moore, Foundries Materials Co, Coldwater, Mich.
- CENTRAL NEW YORK-Chairman Wm. D. Dunn, Oberdorfer Foundries, Inc., Syracuse, N. Y., and Vice Chairman Don J. Merwin, Oriskany Malleable Iron Co., Oriskany, N. Y.
- CENTRAL OHIO—Chairman W. T. Bland, Commercial Steel Casting Co., Marion, Ohio, Vice-Chairman E. M. Durstine, Keener Sand & Clay Co., and Secretary N. H. Keyser, Battelle Memorial Institute, Columbus, Ohio.
- CHESAPEAKE—Chairman Allen S. Kittrell, Leach Pattern Shop, Baltimore, Md., Program Chairman C. W. Galloway, Chambersburg Engineering Co., Chambersburg, Pa. (also V.C.), and Secretary-Treasurer C. A. Robeck, Gibson & Kirk Co., Baltimore.
- CHICAGO—Chairman Walter W. Moore, Burnside Steel Foundry Co., and Vice-Chairman John Owen, Harbison-Walker Refractories Co., Chicago.
- CINCINNATI DISTRICT—Chairman A. W. Schneble, Advance Foundry Co., Dayton, Ohio and Program Chairman B. A. Genthe, S. Obermayer Co., Cincinnati (also V.C.).
- DETROIT—Chairman Vaughan C. Reid, City Pattern Foundry & Machine Co., and Vice-Chairman Michael Marchol, Atlas Foundry Co., Detroit.
- EASTERN CANADA-Chairman W. Turney Shute, Canadian Car & Foundry Co., Ltd., and Program Chairman A. J. Moore, Montreal Bronze Ltd., Montreal, Que., Canada (also V.C.).
- EASTERN New YORK—Chairman John E. Waugh, General Electric Co., Schenectady, N. Y., and Program Chairman Leigh M. Townley, Adirondack Foundries & Steel, Inc., Watervliet, N. Y. (also V.C.)
- METROPOLITAN Chairman D. Frank O'Connor, O'Connor's Foundry, Inc., Hackettstown, N. J., and Vice-Chairman J. S. Vanick, International Nickel Co., New York.
- Mexico Crry-Chairman N. S. Covacevich, La Consolidada S. A., and Program Chairman Juan Latapi Sarre, Fundiciones de Hierro Y Acero S. A., Mexico, D. F., Mexico (also V.C.).
- MICHIANA—Vice-Chairman Andrew Peterson, Oliver Corp., South Bend, Ind., and Secretary-Treasurer Vincent C. Bruce, Frederic B. Stevens, Inc., Elkhart, Ind.
- Mo-Kan-Chairman E. C. Austin, Jr., National Aluminum & Brass Foundry, Independence, Mo., and Vice-Chairman John Redman, Jr., Redman Pattern Works, Kansas City, Mo.
- NORTHEASTERN OHIO—Chairman Gilbert J. Nock, Nock Fire Brick Co., and Program Chairman Frank C. Cech, Cleveland Trade School, Cleveland, (also V.C.).
- NORTHERN CALIFORNIA—Chairman Philip C. Rodger, General Metals Corp., Oakland, Calif., and Program Chairman George W. Stewart, Fast Bay Brass Foundry, Richmond, (also V.C.).
- W. Stewart, Fast Bay Brass Foundry, Richmond, (also V.C.).
 NORTHERN ILLINOIS-SOUTHERN WISCONSIN—Chairman Carl L.
 Dahlquist, Greenlee Bros. & Co., Rockford, Ill.
- NORTHWESTERN PENNSYLVANIA—Chairman J. Douglas James, Cooper-Bessemer Corp., Grove City, Pa., and Program Chairman Fred J. Carlson, Weil-McLain Co., Erie, Pa., (also V.C.).
- Ontario-Chairman R. H. Williams, Canadian Westinghouse Co., Ltd., Hamilton, Ont., and Program Chairman Andrew

- Reyburn, Cockshutt Plow Co., Ltd., Brantford, Ont.,
- Orecon—Chairman E. J. Hyche, Rich Manufacturing Co., and Program Chairman Wm. M. Halverson, Electric Steel Foundry Co., Portland, Ore., (also V.C.).
- PHILADELPHIA—Chairman George Bradshaw, Philadelphia Naval Shipyard, and Vice-Chairman Arnold N. Kraft, Wilkening Mfg. Co., Philadelphia.
- QUAD CITY—Chairman Wm. C. Bell, Frank Foundries Corp., Moline, Ill., Director Eric A. Welander, Union Malleable Iron Works, East Moline, Ill., and Secretary-Treasurer R. E. Miller, John Deere Planter Works, Moline, Ill.
- ROCHESTER—Chairman Charles E. Vaughn, Ritter Co., Inc., and Program Chairman Leon C. Kimpal, Rochester Gas & Electric Corp., Rochester, N. Y.
- Saginaw Valley-Vice-Chairman Kenneth H. Priestley, Vassar Electroloy Products, Inc., Vassar, Mich.
- St. Louis District—Chairman Ralph M. Hill, East St. Louis Castings Co., East St. Louis, Ill., and Program Chairman Henry W. Meyer, General Steel Castings Corp., Granite City, Ill. (also V.C.).
- SOUTHERN CALIFORNIA—Chairman Henry W. Howell, Howell Foundry Co., Los Nietos, Calif., and Harold G. Pagenkopp, Angelus Pattern Works, Huntington Park, Calif. (also V.C.).
- TENNESSEE—Chairman Porter Warner, Jr., Porter Warner Industries, and Program Chairman W. M. Hamilton, Crane Co., Chattanooga, Tenn. (also V.C.).
- TEXAS-Chairman John M. Bird, American Brass Foundry, Ft. Worth, Texas and Program Chairman P. B. Croom, Houston Pattern Works, Houston, Tex. (also V.C.).
- TIMBERLINE—Chairman C. C. Drake, Griffin Wheel Co., and Vice-Chairman James E. Schmuck, Rotary Steel Casting Co., Denver,
- TOLEDO—Chairman John A. Mescher, Unitcast Corp., and Vice-Chairman John G. Blake, Alloy Foundries, Inc., Toledo.
- TRI-STATE—Chairman Dan A. Mitchell, Progressive Brass Mfg. Co., Tulsa, Okla., and Vice-Chairman C. C. Beagle, Webb Corp., Webb City, Mo.
- Twin City-Chairman C. F. Quest, J. F. Quest Foundry Co., Minneapolis, and Vice-Chairman Joseph W. Costello, American Hoist & Derrick Co., St. Paul.
- WASHINGTON-Program Chairman William L. Mackey, Washington Stove Works, Everett, Wash. (also V.C.), and Secretary Fred R. Young, E. A. Wilcox Co., Seattle, Wash.
- WESTERN MICHIGAN—Chairman Ross P. Shaffer, Lakey Foundry & Machine Co., and Vice-Chairman William J. Cannon, Nugent Sand Co., Muskegon.
- WESTERN NEW YORK—Chairman Erwin Deutschlander, Worthington Pump & Machinery Corp., and Program Chairman Carl A. Harmon, Hanna Corp., Buffalo, N. Y. (also V.C.)
- Harmon, Hanna Corp., Buffalo, N. Y. (also V.C.).
 WISCONSIN—Chairman George E. Tisdale, Zenith Foundry Co.,
 Milwaukee, Program Chairman Robert V. Osborne, Lakeside
 Malleable Castings Co., Racine, Wis., and Secretary Albert F.
 Pfeiffer, Allis-Chalmers Mfg. Co., West Allis, Wis.

N

mar

mer

was

I di

cast

lect

agei

51-7

MANAGEMENT OF INDUSTRIAL RESEARCH

By James C. Zeder*

THE HONOR OF BEING INVITED to present the Charles Edgar Hoyt annual lecture is one which I deeply appreciate. The specialization of our modern industry has produced a great multiplicity of technical organizations and societies, but there are very few indeed which can claim so vast and distinguished a membership as yours. I am proud, indeed, to have the privilege of addressing a group whom I respect so highly.

My earliest association with your profession was many years ago when I worked as a boy during summer vacations in a Bay City foundry. A foundry then was a miserable place in which to work, and frankly I disliked it heartily. But I have always been grateful for the experience which I obtained there.

Thanks to your efforts, there have been vast changes in foundries since that time. Working conditions are infinitely better and the quality of the product itself has been tremendously improved.

Since this early experience, I have always been interested in foundries, and not merely from an academic standpoint. The automotive industry uses more castings than any other branch of industry; and the particular corporation to which I belong not only has a large foundry operation of its own, but is also the largest single purchaser of ferrous materials of all kinds in the world today.

However, this discussion will be in a different direction from the technical treatment of foundry problems. Many of the past Charles Edgar Hoyt annual lectures report extensive observations of the materials, processes, and techniques of your great industry. I could not add to your technical knowledge. In contrast to that type of paper, I would like to report on observations I have been able to make in a province of industry that is a vital, basic part of the life blood of your company—of any company—if it is to survive in a highly competitive economy. This is the management of industrial research.

We know that man has been on earth for nearly 500,000 years. It took him 495,000 years to get out of the trees and caves and start living in more or less organized groups. And yet it is only within the last few years that we have had such amazing things as railroads, radios, telephones, automobiles, trucks, motion pictures, and indoor plumbing. And in the last few minutes, comparatively speaking, we have had television, nylon, atomic energy, and jet planes.

What could have happened to cram these things this whole technological advance—into a mere pocketful of years?

It was the advent of research on a large scale.

It was research that brought forth the manufacturing skills on which our present economy is founded. It was research that led the way to the incredible devices of modern living which we have come to accept as commonplace. It was research, and the realization of what could be accomplished through it, that produced the tremendous progress of our civilization in the last 100 years.

There are some who still think of research as a highly secretive process behind locked doors. But this kind of thing is far from the truth about modern research. The majority of information resulting from research is common knowledge when progress can be reported. The protection of patent and copyright laws allows the news of new and good things-the better ways-to be freely circulated to any who may be interested. Thus, the applications of the results of research have not been the coveted secret possession of the few, but the multiplied development of thousands of scientists who have been enabled to use and build upon the research of others, rather than laboriously climbing each step that leads upward to the result desired. This situation has been duplicated to a lesser extent in certain other countries and the end result has been the stupendous advance in technology during the last few years.

Research Defined

Let us refer to Mr. Webster's definition of research. You know, he was quite a researcher himself.

"Research," says Mr. Webster, "is a critical and exhaustive investigation or experimentation, having for

TIONS

Ont.

, and undry Naval

Corp.,
Iron
filler,
, and

ectric

assar Louis

lenry , Ill.

owell

opp,

/.C.).

Co.,

Ft.

ston

Vice-

iver,

lice-

Mfg.

ebb

Co.,

eri-

ing-

tary

y & ent

ng-

A.

Co.,

ide

F.

Orice President, Director of Engineering and Research, Chrysler Corporation, Detroit.

Presented by Mr. Zeder April 25, 1951 during the 55th Annual Convention of the American Foundrymen's Society, Buffalo.

its aim the discovery of new facts and their correct interpretation, the revision of accepted conclusions, theories or laws, in the light of newly discovered facts, or the practical applications of such new or revised conclusions." I want to repeat that. "The practical applications of such new or revised conclusions!"

Origin of Research

Research had its start in two ways; the natural curiosity of man, and the equally natural desire for perfection and better living. This latter desire has been the great motivator. He was cold, so he played with fire. He wanted better food, so he cooked it. Now he freezes it. He was sick—there were plagues—malaria—he fought to be free of them. He wanted to transport himself and his goods more easily—so finally we have automobiles. All were problems—problems whose solutions have raised our mode of living to its present high standards of comfort and convenience.

A few short years ago the United States government did little research outside some of the public health problems-the epidemics and their causes. During World War I, the Chemical Warfare Service learned that many problems could be solved quickly with well-managed research. This type of thing and the great success of the research laboratories of some of the big companies and foundations gave both industry and government a new slant on research. They found it was not only possible but profitable to develop a research department within their own confines or to take their problems to a professional or college research organization. Our government today is responsible for the support of over 40 per cent of all research carried on in America. Governmental budgets for this purpose currently total well over one and a quarter billion dollars-some projects running for a period of years. Industry is spending an estimated 450 million dollars yearly and the number of men employed in industrial research in 1951 has increased nearly fifty times since World War I to an estimated total personnel of 150,000.

Endowed Research Foundations

The role of endowed foundations such as Rocke-feller and Carnegie has become increasingly significant in the development of research for the public welfare and the conservation of our natural resources. Research has often been solely responsible for the creation and success of an entire industry—or the failure through substitution of better things, of another. Look at what happened to the Chilean nitrate industry. Research produced the synthetic process, and the industry which had accounted for over 40 per cent of the Chilean government's income was up against the problem of existence. Look also at the silk industry which has been replaced to such a great extent by synthetic fiber manufacture.

Another great factor in the growth of research is the attitude of the people of this country. *Nowhere* on earth is there an entire country whose citizens are so receptive to change, so eager to pioneer and so competitively minded. Our ancestors were no reactionaries, bent on preserving the old ways of life at all costs. If they had been, they would not have dared the Atlantic Ocean and there would still be wigwams along the Potomac! In nearly every other country, the adherence to tradition has acted as a brake to progress. Here, the philosophy has been that "If father used it, there must be something better by now!"

It has been said by many that we have invented few things of importance. We suppose most of the really good things came from Russia! (Better label that as a joke or we'll be quoted in *Pravda*!) We are told that we merely made them more practical. Even if that were true—it's a wonderful role for us to play. We of the United States have made practical, and available to millions of people—the new ideas, the new products, the better ways of life. It all adds up to more happiness—for everyone. *That*—is praise indeed.

But with the tremendous growth of research there are problems. The different avenues of approach can lead into a maze of bewildering lanes unless research is properly planned. Millions of dollars and millions of man-hours can produce wonderful results—but they can also be utterly wasted. Research can make or break men and companies depending upon the way it is directed or the way it is handled.

I cannot tell this group anything new about research itself. But perhaps I can contribute something to your knowledge of the methodology of research, how the group is organized and how it coordinates with—and becomes a successful part of an industrial organization.

Research Activities Classified

For purposes of analysis, the sequence of those activities that are more or less integrated with research, from the uncovering of new knowledge in the engineering sciences to the final design of finished goods may be placed under four general headings. These may be classified according to the names we prefer, as, Academic Research, Engineering Research, Product Engineering and Development, and Production Testing. There are, of course, other names for these classifications, but we believe these categories will apply to most modern industries.

Academic research may be defined as a search for new facts without regard to their possible use or application. Again, academic research has been referred to as "fundamental" or "basic" and has been called

"research for research's sake."

Experience has definitely indicated that this type of research can be carried on more economically in the long view by universities or research foundations not directly associated with industry. Academic research is the fountainhead of knowledge, opening up new horizons for progress. But to carry on such research in an industrial corporation imposes too much of a strain on normal viewpoints. After all, industrial corporations are run for the express purpose of turning out a successful product at a profit. Setting up academic research groups within corporate walls, and asking it to ignore profit considerations is extremely difficult. In most cases when attempting the academic,

sup cip sio the cor sea haz

all

aca

wo

Ci

the

It

disc ina pro Wa abo

An

Co

sta

got whe like was pla of lish

duct pro
Th
div
wit

F
of t

ing in i

spe

All such tails rese the

orga pers mos ARCH

have

ll be

other

as a

been

bet-

l few

eally

at as

told

en if

play.

and

the

s up

e in-

here

can

arch

ions

they

e or

re-

ning

rch,

ates

rial

ac-

rch,

ngi-

ods

1ese

fer,

luct

est-

rese

will

for

ap-

red

lled

ype

in

ons

re-

up

re-

uch

rial

rn-

up

ınd

ely

iic,

the research group soon finds itself happily—and profitably—engaged in direct engineering research or product development.

Engineering research is a little closer to the product. It is sometimes called applied research; in other words, research with the objective of product improvement. What is meant by this term is the organized search to supply new materials, new processes, and new principles to the improvement of products by basic revisions in design, the adoption of new features, or even the addition of entirely new branch industries to the corporation. The borderline between academic research and applied or engineering research is a little hazy at times, each one overlapping the other. Basically, engineering research fills a large space between academic research and product engineering. We often wondered under what research category, Columbus' discovery of America would come.

Research A By-Product Value

Columbus wanted to prove a theory, he wanted to discover new facts that would aid in the proof. Ferdinand and Isabella, however, wanted new lands, new products, new riches. Strickland Gilliland, one of Washington's noted correspondents, had this to say about Columbus' voyage in a speech entitled "What is a fact?" "We always think of the discovery of America as one of the great facts of history. Yet Columbus didn't know where he was going when he started out, he didn't know where he was when he got there, and he didn't know where he had been when he got back." A great deal of research is just like that, often of by-product value, but it can be wasteful unless it is carefully watched and properly planned.

Product engineering and development is the activity of bringing out the new models—including the establishment of all specifications, the preparation of production drawings and the constant refining and improvement of the product after it is in production. This division is the natural customer of the research division and the two must maintain close contact with each other.

Production testing is the devising and performing of tests on production samples to insure that the product meets the high standards of performance and life specified. These last two classifications do not come under a strict definition of research, either engineering or academic, but they may have a little of both in them at times.

I have reviewed these classifications only as a matter of locating ourselves on the map. Probably all of you are familiar with these general divisions and the various kinds of work that come under each heading. All of you know the structure and organization of such departments, and I will not bother you with details. Rather, I will confine my paper to engineering research and some of the fundamental problems of the management of that department.

This study falls naturally into two sections, the organization and administration of the group, its personnel and facilities, and the work itself. The most important element in a successful research group

is the man who heads it. A research program in an engineering organization must be a carefully planned, integrated effort to be effective, so a research head must not only be research minded but be a good organizer and administrator as well. We have had the experience of taking a research department out of the hands of a highly skilled researcher with little administrative ability, turning it over to a man who was less research minded, but who knew better how to organize and administer the department. The effective research output increased appreciably with exactly the same facilities and the same staff.

Function of Rsearch Director

The research director's duties fall broadly into two categories, technical and administrative. Under the technical category comes, above all, the planning. The director should be able to plan, first, for the over-all long range results and, second, help guide the actual day-to-day work of the group with the use of the facilities available. He must be able to determine the broad objectives and analyze the specific objectives that, when accomplished, will add to the result. He must also know what man or groups of men to entrust with each phase of the work.

Sometimes a research engineer will ride a project he likes for as long as he is let alone. Give him a job that interests him and he is happy to continue regardless of objective. Self-determination causes him to continue until stopped. Another type of researcher is the one who quits or loses interest when the work becomes difficult. In this case, the man must be guided or switched to less important responsibility. The director must be alert and on his guard constantly against situations of either kind in his organization.

After the planning comes the direction. While a research head need not be the top scientist in the organization, he must certainly have imagination and a thorough knowledge of his field, a good understanding of what research is all about, and an overwhelming desire to get on with it! A director must help guide the procedure until it is shown that the work is accomplished or it is not economical to continue further.

The director must also be a good leader. He must know his men and machines, which group to encourage and which to hold in check. He must keep track of work progress and suggest changes necessary to continue new interest in the solution.

On the administrative side, the research director must be able to present to the management a fairly detailed picture of equipment, personnel, and machines for each project. He should set up tentative schedules for the use of equipment and purchase of new instrumentation, although we recognize that in research this is not too easy to predict. The cost of such things as electron microscopes, cyclotrons, wind tunnels, and even the smaller units of scientific research instrumentation makes the multiple use of each item imperative. The director must check progress rates, and keep his staff adequate both in number and quality.

A research head has one of the most unappreciated

and frustrating jobs in the world—selling new ideas to people who are operating successfully with old ones. He must have both the courage of his convictions and determination to sell them to the management. He should be a good public relations man and should have enough showmanship or sales ability to get an idea across to the executive group. By salesmanship we do not mean just aggressiveness, but rather the skillful use of all the legitimate methods developed by modern business to put across its ideas and products. If management consistently turns down its research department, the research director must take some share of the blame, because selling research is a substantial part of his job.

Finally, and above all, a good research head must be alert to company objectives. He must concentrate on what is best for the corporation, and not be lured away to something less important or ride a hobby of his own. This at times means having the courage to stop a project, even at the risk of losing face, when it becomes obvious that results will never justify the costs. One of the most important things in research

is to learn what not to do.

Operation of Research Group

In the realm of the work itself, the successful operation of a research group as a whole demands two things, a continuous flow of new projects, as well as an output of successful and valuable developments.

Just as we have stressed the qualities necessary for the director, we wish to emphasize the extreme care with which research projects should be chosen. To a large extent, the quality of research output depends upon the intelligent choice of projects or objectives to be accomplished. Research should not move in the line of least resistance or necessarily that of greatest scientific interest, and a project should be acknowledged from the beginning as either a constructive step in product or process improvement, or as the start of an entirely new product, process, or industry. In the selection of these projects, several pertinent questions should be asked:

1. If the work is successful, will the results be used in the present product or for the direct benefit of other corporation groups?

Will the probable cost of doing the research justify the results obtained, and be within reasonable

financial means of the corporation?

3. Will it be logical to use the results at the time they become available, or will they be so far in advance of manufacturing techniques and market requirements that their public acceptance will be questionable?

There are many other questions such as—is manpower available and capable of handling the job? are adequate facilities and instruments available? is its scope too broad or not broad enough?—will it interfere with the everyday business of the corporation?—availability of raw materials, etc.?

The entire burden of answering these questions must not rest solely on the research group. When a preliminary investigation by this group has led them to believe that a project has product value, a review

of it and its proposed scope should be held with the heads of the production engineering and development groups in whose field of activity the project may eventually lie. This is of tremendous importance to the research group not only in evaluating the project, but in establishing shoulder-to-shoulder team spirit between research and development from the very start of the project. The most productive results are obtained when the research department is integrated into the engineering organization in this way instead of separated from it. It is possible, of course, to tie it in too closely, and the proper balance must be found; but isolating a research facility from the rest of the corporation-either geographically or organizationally-has definite disadvantages. main outlet which research has for its products is the product engineering and development group, and it is important, when selling to a customer, to have good contact with him. Also, research needs reasonably close association with the rest of the engineering department in order to keep informed on the existing problems and the immediate directions of corporation trends. This is a most important point and one that aids greatly in making use of the results when the project is successful.

Without this method of working together, the development laboratories are often inclined to be "against" the projects sponsored by the research group. Each side may claim that they should have had the job and that they could have done better or that the work should never have been started at all. By putting both sides in complete possession of the facts at the start of a new objective, and allowing each to take part in the decision to go ahead, probabilities of friction are eliminated and when research is through with its part, production engineering and development groups will proceed with better spirit

and more cooperation.

Despite the care of the research and executive groups in layout of the original project and the thorough analysis of it by the research director and the engineering groups, the best laid plans often go astray. Careful preliminary work does not preclude the possibilities of engineering compromises or unforeseen difficulties making it necessary to stop the work before it is finished, or to completely change its direction. The aim is to keep the loss of work and effort due to such circumstances to a minimum. Frequent progress reviews should be held with the original groups, so that everyone concerned may learn of any change which might affect his work or the use of the results.

Research Project Approval

We have learned through experience that nothing is quite as disorganizing to a research group, or as unsettling to its relations with the organization as a whole, as a project which has been approved merely because it was someone's favored hobby or was oversold originally by an influential sponsoring group. The unwillingness to listen or give sufficient time to the careful analysis of the problem, or to spend enough money on its preliminary exploration, will oftentime result in a high percentage of rejected ef-

is to a li roc since pat ever the to c ject por con

CH

for

mo

pla

H

wol F the own men and rese Thi

ma

edu alre ter of cou inte ject grou

wheengs basis atte old the

T

idea and a re ther grou for or s the thin the seen for a

forg men is co oper not grou RCH

the

nen

may

e to

pro-

ean

the

re-

t is

this

, of

ince

rom

Ol

Γhe

the

d it

ave

on-

ing

ist-

ora-

one

nen

de-

be

rch

ave

or

all.

the

ing

ob-

rch

nd

rit

ive

or-

he

go

de

ın-

he

its

nd

re-

ig-

of

of

ng

as

ly

er-

p.

nd

11

f-

fort and the loss of the work of capable men when a more careful approach would have shown that the project should never have been started in the first place.

For the operation of a research department there is no royal road to success. To stretch the metaphor a little farther, we might say that it is an extremely rocky and tedious path, if there is any path at all, since, of course, the job of research is that of the pathfinder and the pioneer. Even the pioneer, however, finds a place to settle down and take root, and the researcher, too, must occasionally stop long enough to consolidate his gains into usable form. If the project is successful, it becomes the problem of the corporation management to decide when it is to be commercialized or, expressing it from the research man's point of view, torn from the sympathetic arms of its fond parents and thrown to the development wolves.

From the viewpoint of the research department, the new idea is seldom old enough to stand on its own feet when it is transferred over to the development group. It will probably always be that way, and in some cases it may be necessary to allow the research group to continue the development work. This may arise because it would take too long to reducate a development group to the point of progress already reached by research, and it is faster and better to allow research to complete the job to the point of actual use in the product. This condition, of course, is not usual, and it should not be allowed to interfere with the normal transfer of research projects to the development and production engineering groups.

We have learned one thing, however, and that is to allow the research idea to continue to the point where its acceptance by development or production engineering groups is assured on at least as sound a basis as their own developments. Only too often we attempt to pour the new wine of research into the old bottle of convention and lose both the wine and the bottle.

Release of Projects and Ideas

The direct transmission of successful projects and ideas from research to the production engineering and development group is not the only way in which a research department serves its purpose. Now and then a new research design or idea is rejected on the grounds that it is too far in advance of its time, or, for various reasons it could not be used immediately, or simply that it is not desired commercially. But the idea nevertheless has had its influence on the thinking of the organization. It may just help pave the way for another idea which by contrast may not seem quite so radical. Or it may simply lie dormant for a time, acting as an inspiration for future thought and commercial realization even though the fact be forgotten that it originated in the research department. In these ways, a good research organization is constantly having its impact on the development operations of the company, even though there does not seem to be direct linkage between the various groups. At the very least, research is an important

stimulating influence which any progressive and growing company cannot afford to be without.

It is an oversimplification, however, to assert that research is a good thing and that every industrial corporation should have it. It must also be properly organized for the purpose intended. The mere fact of having a research department is no justification for a feeling of smugness. No one will attest more vigorously to the value of research than the writer, but we do not mind admitting that we made false starts over a period of years until experience directed us to develop the type of research organization which was most logical for us. We learned that a research department must be tailored to the particular job it has to do. By its very nature, there is no general prescription that can be followed. What is logical for one corporation may be all wrong for another selling a different product in a different market. Each company must devise its own formula.

How much should be spent on research? If enthusiasm is allowed to overwhelm reason and good business judgment, it is possible to spend too much. Research itself has become more and more costly. The instruments and equipment required are much more extensive and complex than in the past. Also, with the product of today, research channels have multiplied many times, so now it takes a lot more research manpower and facilities to develop the desired improvement and refinement.

Financing of Research Activities

With some companies the idea prevails that a fixed percentage of each sales dollar should be set aside for research. The thought is fine, but the hypothesis is wrong. Research does not start or progress on a dollar basis. A good research program depends on an intelligent understanding of the job to be done. On the money hypothesis, the research director's incentive is to look around for additional ways to spend the money. You rob him of the stimulation of having to compete with other divisions of the corporation for his share of the budget. Also, during periods of low sales volume, the fixed percentage system may result in drastic reductions in the research program at the very time when real research development is needed most. It is far better for the corporation, as a whole, for the research department to sell the management on the aimed objective or process improvement to be realized from the projects undertaken and request the money to do the job.

But while research cannot be bought on the dollar basis, industrial corporations naturally have the right to expect it to pay its own way and much more. If it has not accomplished that objective then it has probably not been properly organized or directed to meet the specific needs of the company. In such cases, the trouble is not with the job but with the management policies under which they are operating. With intelligent management, good direction, and sound organization, industrial research is *sure* to pay off.

There are almost as many methods of measuring the results of research as there are companies using research. In the Fourth Annual Conference on the Administration of Research held at the University of Michigan in September of 1950, Allen Abrams, Vice President of the Marathon Corporation, gave his definition of the way to measure research results. Said Mr. Abrams, "There is only one real measure of the results of research, one which will stand the test of history: 'Will it produce a new industry, new goods, new happiness, and most of all, a better world than we had in the past?'"

Guy Suits, Vice President of the General Electric Company, had this to say about the difficulty of measuring the results of basic research, "Let us take a look at our work in silicons, where we began by trying to replace a carbon molecule by one of silicon. We thought this would be a good idea because the silicon bond gives us a higher temperature stability of certain materials, but we never thought that it would give us bounceless rubber and rubber which would be compatible with pyranol in transformers."

A new industry arrived at!

The Minnesota Mining and Manufacturing Co. was not looking for scotch tape. But they found it and another new industry resulted. The petroleum industry is making alcohol, the alcohol producers are fermenting penicillin. I have already mentioned the silk industry.

For the purposes of management accounting, some large companies use a system of "royalties" whereby the research department is credited with a certain amount based on the probable cost of obtaining the license to manufacture the product in question. Another method is to set off the appraised value of the product against research cost.

The end results of research are everywhere apparent. All you need to do is look around you and to realize that this advancement has almost entirely come in the last 100 years. That is only two ten thousandths of the time man has been on earth.

I have left to the last a discussion of why industrial organizations must carry on long expensive research programs and spend millions of dollars for permanent research facilities. There are undoubtedly a good many reasons that we have not set forth, but fundamentally, the fact exists that at no time in history was there any indication that man was satisfied with his lot. Man is a pioneer. He has experimented with almost every phase of his existence, and has continuously sought for improved transportation, housing, food production, distribution, social organization, defense and health. This desire to achieve a closer approach to what man considers perfection in all fields of activity is so fundamental that we can almost conclude that it is a law of human behavior.

Research Is Good Business

Viewed from this standpoint, research is good business, and by an organized effort in this direction, we are making use of a fundamental law of human nature in satisfying the desires of mankind for constant progress. Competition certainly spurs industrial organizations into this activity. As an example, no one could compete in today's automobile market with a 1930 automobile, but competition itself is merely an

exemplification of the way research works to gratify the human desire for better things.

Staying in business in a competitive economy means that research must be a successful division of the business, and that as such it has to be done with good business judgment. The good judgment demanded by research is the same that would be exercised with any other capital investment. Patience is required, because research cannot be turned on and off like a machine. Once having set your hand to the plow of research there is no turning back, because you have committed yourself to progress, and your customers will demand it. Tomorrow's customers will have grown to expect even more from all of us in industry and if we wish their approval, and I hope their orders, the results of our research efforts will have to be better and more numerous than before.

Research has sometimes been characterized as a gamble. In the sense that it is an investment in an enterprise with an unpredictable outcome, it is a gamble; but the fact that its success is dependent on the quality of its direction and execution makes it a game of skill rather than one of chance. The great industrial might of this country has been built on gambles of this kind, and it is becoming increasingly apparent that the biggest and riskiest gamble which industrial corporations can take is to do no research at all.

Doing no research does not guarantee your present level of earnings. Your greatest need for research might not be to increase your profits but to maintain them. Says Roger Williams, executive committee advisor on research at DuPont, "We do research because we have to. If we let up, our competitors would trim us. Research is our insurance that our investment will continue to be profitable." Without research, your product may become hopelessly obsolete, not by the same thing being offered at a lower price, but by a product entirely different which does the job in a new, and better—and cheaper—way.

The men of engineering, the men of science, and the men of business and industry have always before them a dual purpose, whether all of them realize it or not. This purpose is to so serve their company that its competitive position in the industry shall not be successfully challenged. In doing this, they also serve their country-so that the position of leadership of the United States shall not be successfully challenged by any power on earth. We all know this. All engineers and men of industry in the United States are aware of this. They serve the ideals of the company they work for, and plan for, and dream great dreams for. But behind every job well done, every plan well executed, and every dream of better things to come-is the working and the planning and the dreaming for the United States.

Our ancestors took the great gamble—the gamble for their lives and their ideals when they came to this land. Pioneers—men and women willing and eager to stake their all.

The men of industry in our country have inherited this great spirit—the willingness to pioneer. It has been said that to pioneer is to risk mistakes. Not to pioneer—is failure. The United States will not fail. Prodoi to abl ma this light

mig ing tion mu gro to I tho

way
in i
pet
to i
gree
the
in i
and
to v

can and to r

tory T pres

51-3

SOME THERMAL CONSIDERATIONS IN FOUNDRY WORK

By

Victor Paschkis*

Introduction

EVERY HUMAN ENDEAVOR which persists over long periods of history creates a pattern of tradition which becomes the more set in its ways—or may we say ruts?—the older it becomes, the longer it is practiced. Progress is achieved not so much by ruthlessly abandoning time-honored traditions, as by the willingness to hold up to critical examination by the best available gauges the traditional patterns of action; and to maintain only those which prove to be acceptable by this test, while discarding all methods which in the light of present insights are no longer the best suited for the ultimate goal.

In these times of continued danger of war, one might be tempted to consider the possibility of applying this method of progress to the field of human relations, particularly regarding the pattern—much too much ingrained by tradition—of settling differences of group opinion by force instead of reason. But we have to limit our considerations to the application of these thoughts to the art of casting, to the foundry trade.

As soon as man knew how to melt metals, the pouring into a desired shape-casting-became the natural way to obtain complex shapes of great strength. Only in modern times has foundry work found serious competition in the form of welding and brazing. In order to meet this competition, progress is needed; and progress involves, as stated before, the critical review of the situation. This paper presents such critical review in one aspect of the casting field: that of heat flow and heat transfer. The foundryman is so accustomed to work with heat, that he is hardly aware of the heat -or thermal-laws with which he is dealing. But only by becoming aware of them, and recognizing them, can he hope to improve his methods in a systematic and economical way. Lacking such knowledge he has to rely on guessing and intuition, both of which have their place in industrial life, but are frequently all too unreliable.

It is known, that frequently an outsider looking at a problem may find suitable gauges and remedies, because he is less hampered by tradition than the person who is in continuous touch with the problem. The author is aware, that he is an outsider in the field of foundrycraft-not having ever made a casting or even a mold; but for the past seven years he has been looking at foundry problems with the eyes of a heat transfer man and thus the views and impressions of a heat transfer man regarding foundry practice will form the main part of this paper. In modern industry results can be obtained only by teamwork and cooperation between many specialists, one of whom, in foundry work, should be a heat transfer expert. Thus it is hoped, that the paper may not only serve as a gauge, but have some usefulness in building a science of metal casting to gradually replace the well perfected foundry art.

I. Fundamentals of Heat Transfer

1. Analogy to Flow of Liquid Metal—The foundryman, who in the field of thermal relations finds it difficult to accept the difference between "heat" and "temperature" and to visualize the rather abstract happenings of heat flow, may be helped by thinking of some well accepted observations with regard to the liquid metal. Everyone will expect to see liquid metal flow from a ladle, when the latter is tilted above the shop floor, and will not expect to see metal flowing out, if the ladle were tilted within a bath of liquid metal. For example consider (Fig. 1) a small spoon ladle full of metal within a large transfer ladle, also full of metal: there is no difference of level between the metal in the spoon ladle and that in the transfer ladle.

Metal will flow, if unobstructed, as long as there is a difference of level. Similarly, heat flow will occur always, if there is a difference of temperature levels. But, still referring to Fig. 1, the level difference between the transfer ladle and the floor in the position shown does not lead to flow of metal, because ladle with its lining forms a complete barrier to the flow of metal (until a hole is burned in the lining and ladle). No such complete barriers to heat flow exist. Whereas

ratify
neans
f the
good
ed by
n any

, beike a

have mers have ustry

their ve to

as a

n an

is a

t on

it a

great

t on

ingly

hich

arch

esent

arch

nain-

ittee

be-

itors

our

hout

ob-

ower

does

and

fore

ze it

oany

not

also

ship

chal-

All

tates

com-

reat

very

ings

the

nble

this

r to

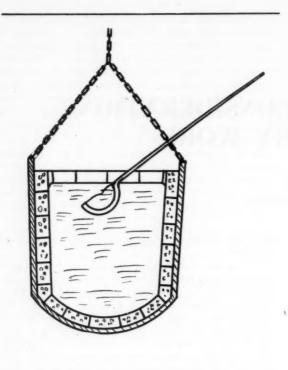
ited

has t to

ail.

^{*} Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York City.

Thirtieth annual official exchange paper from A.F.S. to I.B.F. presented at the 48th Annual Meeting of the Institute of British Foundrymen, Newcastle-on-Tyne, England, June, 1951.



Shop Level
Fig. 1-Fluid Flow in Ladles.

in the case of the ladle and the metal contained therein, the resistance to metal flow is infinite, there does not exist any material or arrangement which results in an infinite resistance to heat flow. Thermal resistance can be large or small, but can never be either zero or infinite.

The resistance to heat flow increases with length of path to be traversed by heat and decreases with the cross-sectional area of the heat path (similar to the way in which the resistance to metal flowing out of a melting furnace increases with the length of the spout, and decreases with the cross-section of the latter). Moreover, the resistance to heat flow depends on the material in which heat flow occurs. The property expressing this dependence is called the thermal conductivity. The thermal conductivity indicates how many Btu (British thermal units) flow through a body having a cross-section of "one" (say square foot), and a length of "one" (say foot), if the two faces are exposed to a temperature difference of "one" (degree F). The significance of this concept of thermal conductivity (usually denoted by "k") and the dimension of Btu/ft, hr, °F will be made clear in discussing the various thermal problems in the foundry.

Another concept of heat flow, which is important, is fortunately much more familiar to the foundryman: specific heat. The latter indicates the amount of heat absorbed by the unit weight (1 lb) of a material, if its temperature is raised one degree. It is expressed in Btu/lb, °F. More conveniently one should refer the specific heat to unit volume, instead of to unit weight. This figure is called "volumetric specific heat"

(Btu/cu ft, °F), and is found by multiplying the value of specific heat (Btu/lb, °F) by that of the density of the body (lb/ cu ft). We will denote the volumetric specific heat by c. The heat problems in which the foundryman is interested are almost all of the "transient type" which is better known as the "unsteady-state"; in such problems temperatures vary with time. In transient type phenomena both k and c enter; in most instances only their ratio (k/c) is important, and in some instances their product (k.c).

During solidification of metals the so called "latent heat of fusion" is given off. This "latent heat of fusion" acts in a similar way as if temporarily, that is over the range of solidification, the specific heat were increased. Heat of solidification is expressed in Btu/lb; but in heat transfer considerations the product of latent heat x density enters; this has the dimension of

Btu/cu ft.

There is still one concept which we have to introduce. It was stated above, that whenever there is a level (temperature) difference, heat flow is bound to occur. Our concepts of conductivity and specific heat take care of the heat flow within a body. But if the surface of the body is at a temperature elevated above that of the surrounding, the body will lose heat to the surroundings: a casting, after the shakeout may have a surface temperature of 1400 F, and face a shop atmosphere at 80 F. Heat flows from the casting to the air; the rate of heat flow depends on a number of items: the velocity of the air surrounding the casting, the nature of the surface (clean or scale) and the temperature level of the surface influence this rate. Moreover, the rate of heat flow is greater if the temperature difference between surface and atmosphere is higher. Thus it is customary to characterize the rate of heat flow from the surface to the surrounding by a proportionality factor, h, called in the heat transfer language the "boundary conductance"; it is represented by h, and expressed in Btu/sq ft, hr, °F.

Actually, as mentioned, the boundary conductance is not a constant, but depends on a number of variables, mainly on temperature. However, this relationship is taken care of by a number of mathematical operations, which need not concern us here.

2. Some Practical Consequences of Heat Transfer Theory—The properties introduced in the previous section are those which determine the nature of the thermal occurrences in foundry work, but only through complicated mathematical functions.

Without going into theoretical considerations, it may be of interest to state one result of the science of heat transfer, which has applications in many fields of

the foundry.

The time required for a thermal occurrence in bodies of the same material and the same shape, but different size varies with the size in a definite relationship; this relationship lies between linear and square proportionality. What does that mean?

Suppose two castings are made: both have exactly the same shape; the larger one has all dimensions exactly twice that of the smaller one. Similarly the mold dimensions for the larger casting are exactly twice as large as the corresponding ones for the smaller cast-

th

in

of

th

kn

ex

ni

di

ele and sid lar wil the a "

her two

tin

wh ma the par ma

spr peo san in sta

sta: nai ing ing. Then solidification of the larger casting will occur in a time which is between twice and four times the time required for solidification of the small casting. In most practical cases the figure "four" (square of the length ratio) is so close to the correct result that one can work with this figure. Similarly, if it is known that a core of a given shape and size takes, for example, 2 hr to dry, then a core with all dimension trebled as compared with the first one, will take six to nine times as long to dry, than the small one.

The usefulness of this approximate rule is further enhanced, if one introduces the concept of "critical dimension." Assume (Fig. 2) that a car bottom type

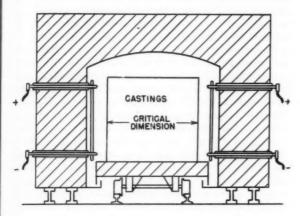


Fig. 2-Illustration of a Critical Dimension

electric annealing furnace resulted in a satisfactory anneal. The heating resistors were placed on the four sidewalls, but not in the car bottom or roof. Now a larger furnace is being considered. A change of height will not call for a change in heating cycle; with only the sidewalls serving as heat sources, the height is not a "critical dimension." But if length and width of the hearth are increased proportionately let us say to 1.5 times their original values, the situation is different. Length and width are the critical dimensions, and hence the heating time has to be increased to between 1.5 and $1.5^2 = 2.25$ times the original value.

The critical dimension, then, is that in which the heat flow occurs principally. By disregarding the noncritical dimensions we introduce a small inaccuracy, because there is practically never a dimension, in which there is "no heat flow." The closer we approximate the condition, i.e. the smaller, percentage wise, the heat flow in the non-critical dimension is, compared to that in critical ones, the closer our approximation holds.

II. Heat Problems in the Foundry

Flow Diagram—Thermal problems are more widespread in the foundry, than might be commonly expected. Figure 3 is a flow diagram of a foundry using sand molds; operations which should be based at least in part on thermal consideration are marked by a star. One important item is omitted from the diagram, namely, the design of the casting; in designing a casting, thermal considerations should very definitely enter, although in most instances the designer is not aware of this. Also, some of the parts of the production process are not customarily recognized as representing "heat flow problems." Therefore a brief review of the processes from a thermal angle may not be amiss. Inasmuch as much work has been done during the past several years on thermal aspects of solidification, the paper will deal primarily with this field.

1. Design—Thermal considerations here include the difference in solidification times of sections of different thickness and shape, limits of thickness which can be cast and stresses to be expected as results of differences in solidification and cooling pattern. Inasmuch as these problems reappear below in various other sections no further discussion here is necessary.

 Melting—With a few noticeable exceptions, one may say, that the two predominant types of melting equipment in this country are the cupola and the arc furnace. For producing malleable iron, a reverberatory-type furnace is widely used, known as "air furnace."

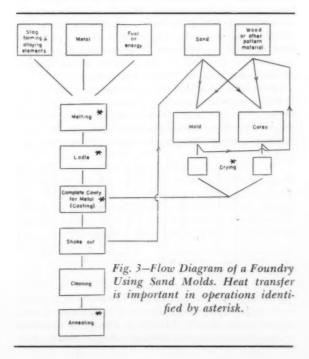
The cupola is considered, and rightly so, as the least expensive means of melting. But that does not mean, that for the quality obtained in the cupola, no lower melting costs can be obtained, than at present. Utilization of waste heat for preheating of air charge and fuel are already at least partly under consideration. Factors influencing the selection of wall thickness include:

Mechanical strength of the cupola;

Life of the refractories, governed in part by the inside surface temperature as well as the steepness of the temperature gradient in the wall;

Heat losses through the lining.

The latter two items are essentially thermal problems. Similarly, selection of size of the pieces of charge and coke is in part a thermal problem. Means of tem-



hat is were Btu/ act of on of intro-

WORK

value

sity of

metric

ch the

"tran-

teady-

time.

er; in

rtant,

latent

at of

nd to heat if the above the have of the of the pat-

er of sting, I the rate. tem-

rate by a nsfer epre-

varitiontical

nsfer vious the only s, it

ds of e in but

actly s exnold e as

cast-

perature measurement and elimination of errors incurred in such measurements are but a few examples

of points of possible further progress.

In arc furnaces the electrode size is selected (mostly by the furnace builder) based on experience. It is not known, if the sizes which are selected are the most economical ones. Changes in size might result in more than power savings and in increased output. Spacing of the electrodes, length of arc, insulation of the lining, thickness of the lining, and also the electric connected load (rate of power input), all have bearing on the furnace operation and economy and all are thermal problems. Core type induction furnaces, as used for melting in brass and aluminum foundries, operate frequently only during one or two shifts; but it is not practical to shut them down for the balance of the 24 hour day; hence they "idle" during one or two shifts, and thus the problem of heat losses becomes quite important. Yet next to nothing is known about them. A foundryman might argue, that this is the responsibility of the furnace builder. But this is a typical case of the responsibility being tossed back and forth between supplier (furnace builder) and user (foundry), with the problem remaining unsolved to the final detriment of both.

3. Ladle—Ladles serve to hold the metal after tapping, and of course also for transfer of the metal to the mold. Thickness of the lining, degree of preheating of the lining, time of holding and/or emptying, time between emptying and refilling, all influence the temperature at which the metal is withdrawn. With today's knowledge of heat transfer, this influence can be expressed in reasonably accurate terms. Generally speaking, one can say, that, the temperature drop will be smaller, the smaller the ratio of k/c (thermal conductivity/volumetric specific heat) of the lining.

Assuming that a ladle is filled in zero time, and also emptied in zero time, but held full or empty for finite intervals between the "filling" and "emptying," one can make the following statement regarding the influence of times on temperature drop: provided that the ratio of "full time"/"empty time" is constant, smaller heat loss will be encountered, as the lengths of "full time" and of "empty time" drop.

In order to illustrate this, reference is made to Fig. 4, which represents schematically two different modes of operation of a ladle. Times are plotted as abscissas and the metal content of the ladle as ordinates. Inasmuch as sudden filling and emptying is assumed, the

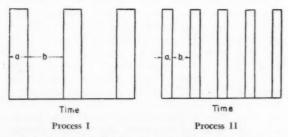


Fig. 4—Operating Schedule for Ladles. $a/b = a_1/b_1$ is the expression for equal ratio of "full times" (a and a_1) to "empty times" (b and b_1). Process II results in smaller heat losses in the ladle than in Process I.

metal content rises from zero to its full value immediately: in the diagram it is indicated by the vertical lines for each "full" period. The length of the "full" and "empty" periods are indicated by a and b. The two modes of operation are indicated by I and II, both having the ratio of a/b. They differ by the length of a versus al and b versus bl. In the case of operation II the heat losses and therefore the temperature drop of the metal in the ladle will be smaller than at operation I, although the ratios of "full time" ("empty time" are the same for operation I and II.

4. Casting—The process of casting is the central part of foundry work. It involves the coordination of flow of the liquid metal and of flow of heat. Only if the two are properly coordinated can sound castings be expected. Metal can flow only, as long as it is liquid; hence the solidification times of the metal as it flows through the gate, sprues, runners, filling the casting, and risers are of utmost importance. As metal freezes along these paths, it decreases the available feeding area for liquid metal, and thus interacts directly with the problems of liquid flow. Small wonder, therefore, that this part of the foundry work has received most attention from a thermal point of view. Some aspects of this work are discussed in part IV.

5. Drying of Molds and Cores—Drying of cores and molds consists in raising their temperature so that either (in case of oil binders), oxidation takes place, which increases greatly with the temperature obtained, or the vapor pressure (water) is increased, to drive the vapor through the interstices between the individual grains of the sand. It is known, that an attempt to dry too rapidly will result in a hardened shell, which prevents the moisture of the layers at some distance from the surface to escape; and the final result is a spoiled mold or core. Qualitatively the foundryman is aware of these conditions. But he probably does not know, that heat transfer is advanced enough, to give quantitative answers to problems involved.

Two remarks may be appropriate here. One refers to the apparently quite successful attempts to apply dielectric heating to the baking of cores. Dielectric heating results in a temperature distribution in the core, basically different from that obtained in conventional baking operations. Figure 5 shows schematically the temperature distribution in baking a core of thick-

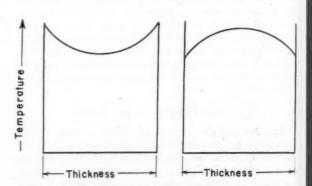


Fig. 5—Schematic Temperature Distribution in Core.
(a) (left) During Conventional Drying Operation.
(b) (right) When Dried by Dielectric Heating.

Н

nes

of

pre

cru

dry

bec

dry

ten

ger

wa

sur

cer

ity

rat

hea

po

5a)

site

mo

cai

pa

cre

sul

SIII

dri

flo

of

the

Fig

all wo var ma wh goa of

ent cat pre dif lish

ma the tio rtical full The d II, the se of pera-

ne"/ ntral n of ly if tings it is al as the

netal

aller

able s diider, has iew. V. and that ace. ned.

rive ndimpt nell, dissult dryibly

igh, fers ply tric the en-

ally ick-

re. on.

VORK

ness L; the left Fig. 5a shows the distribution in case of conventional heatings: the highest temperature prevails at the outside, the lowest at the center. Hence the outside will dry first, forming a relatively hard crust with low permeability. As the thickness of the dry layer increases, complete drying of the wet center becomes increasingly difficult. Remedy is possible, by drying in an atmosphere of controlled moisture content thus delaying the drying of the outside.

In case of dielectric heating (right Fig. 5b) heat is generated within the core; the temperature drop towards the surface is caused by the heat losses from the surface, and the highest temperature exists at the center. Because no hard dried crust of low permeability is formed on the outside, much higher heating

rates can be applied.

The second remark refers to a comparison of the heat flow in the sand while drying with that after pouring. When drying by conventional heating (Fig. 5a) the flow of moisture and that of heat are in opposite directions: heat flows towards the center, and moisture towards the outside. Hence, the moisture carries part of the heat which had reached the inner parts of the mold back towards the surface. A decreased apparent thermal conductivity of the sand results. In case of the metal in the mold, the inside surface temperature of the mold increases rapidly, and drives the moisture towards the outside. The heat flow is in the same direction; consequently the flow of moisture supports the heat flow; an apparent higher thermal conductivity of the sand results (Fig. 6).

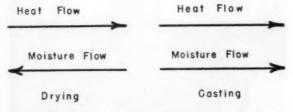


Fig. 6-Heat and Moisture Flow in a Mold During Drying and During Solidification of the Metal.

III. Methods of Making Thermal Studies

On Thermal Studies-It was stated above that all too little is known about heat flow in foundry work. This becomes understandable indeed, when the various methods are surveyed, which could be used in making thermal studies.

Before making such a survey, we should consider, what is meant by "thermal studies." The ultimate goal of thermal studies is, of course, the production of better and/or less expensive castings. In order to approach this goal, it is first necessary to know present conditions; for example with reference to solidification, how, thermally, the casting behaves under the prevailing operating conditions: how long does it take different parts of the casting to solidify. After establishing present conditions one has to find out which changes are required to obtain a more desirable thermal pattern. Obtaining of directional solidification, then, is a thermal problem. Knowledge and observation of temperatures at prevailing conditions is a step

in the right direction, but not sufficient; it might be followed up by an attempt to change conditions-an attempt, best based on heat transfer knowledge.

2. Direct Observation—Thermal studies, by direct observation imply therefore the possibility of measuring temperatures, and to change conditions at will. Both requirements are difficult to meet. Proper temperature measurements at very high temperatures and particularly of molten metal are difficult; most measurements are too inaccurate for any quantitative conclusion, to be drawn therefrom. Change of conditions -e.g. providing molds of various thermal conductivity -is, for economic reasons next to impossible. The same casting would have to be made over and over again. Because in shop practice it would be almost impossible to avoid secondary influences (e.g. minor changes in pouring temperature) which might becloud the results, a large number of tests would have to be run, that is, a large number of castings made, repeating tests with supposedly identical conditions. Thus a statistical average can be made, and then the various averages compared to detect the desired influence, e.g. the influence of mold conductivity on solidification rates. A list of these difficulties should include that of providing the desired variations (e.g. sands of varying conductivity).

To any practical foundryman it becomes obvious, that such systematic tests cannot be made in the shop; and even in specialized research laboratories they

would be extremely expensive.

Model Experiments—Another method is that of "model experiments"; they are essentially castings of the desired shape, scaled down as to size. The science of heat transfer permits, within certain limits, the correlation of the geometric scale factors with the time and temperature scales. Related to this technique is one in which the material in which heat flow occurs is replaced by another one which is preferable from some experimental aspect. An example for this latter method is the attempt to study solidification by means of wax models, replacing the solidifying metal by wax; the advantage here is, of course, the work with much lower temperatures.

Quite apart from the limitation of accuracy, there still remains the need to make a large number of experiments, and compare statistical averages, rather

than individual experiments.

Mathematical Methods—Besides these two methods there is only one left: that of mathematical analysis. However, there are a great number of different techniques, of greatly varying practicability.

What might be called the "classical mathematical approach" includes setting up a general equation, covering a problem, and solving this equation. Mathematical science does not as yet permit a general solution for the complete range of conditions prevailing

In case of solidification, approximate solutions have been developed, which either disregard the finite thickness of casting and/or mold1; the difference of thermal properties between liquid and solid state² (let alone the fact, that actually these properties vary continuously), but irregularly, with temperature, so that the assumption of "a value" for liquid and another value for solid metal already represents a simplification); formation of the air gap, etc. In fact, there are so many simplifications, as to make the results, practically, next to useless. But notwithstanding these simplifications the mathematics becomes in too many instances too involved for the foundryman, who has but little time to spare for such theoretical work.

Another group of mathematical techniques include graphical analysis, probably first introduced by E. Schmidt³ and then elaborated by Nessi and Nisolle⁴; and finite difference methods, worked out in detail by R. V. Southwell³ and applied to thermal problems particularly by H. Emmons⁶ and G. M. Dusinberre.⁷ All these techniques are rather lengthy, and require considerable familiarity on the part of the user so that they are hardly useful to the foundryman, who lacks daily touch with mathematical technique.

If then, no research methods are available, which the foundryman can apply directly he has to use the help of heat transfer specialists. Of the methods mentioned above, the "classical mathematical approach" is hardly available even to the mathematician. Graphical methods are lengthy. Finite difference methods may be eventually perfected for sufficiently accurate

application to solidification problems.

All foundry problems have a large number of variables; e.g. in case of the problem, as to how solidification progresses—even in as simple a shape as a cylinder, the following variables enter: diameter and length of casting; thickness of mold; thermal conductivity and volumetric specific heat of the cast metal, in the liquid and solid state; heat of fusion of metal; pouring temperature; liquidus and solidus temperatures; thermal conductivity and volumetric specific heat of mold; width of air gap between casting and mold; time when this air gap forms; boundary conductance between outside of mold and surrounding air. Such a list makes it understandable, that for most foundry problems the mathematical solution is not known.

All the methods mentioned before, even when developed would be cumbersome in solving for a sufficient number of values of all variables involved. Therefore a method should be described, which has proven successful in studying thermal problems in foundries: that of using an analog computer. Because of the importance which this technique has gained, it will be dealt with in a separate section, although it is a "mathematical method."

5. Analog Computer—Certain groups of phenomena in nature follow the same mathematical laws, although they seem to be physically quite unrelated. In all such cases it is possible to carry out measurements in one physical field, and apply the result to the other field, which follows the same laws. This, essentially, is the basis for the analog computer. For reasons not known, the flow of heat by conduction in bodies and the flow of electric current in bodies with evenly distributed resistance and capacitance follow the same mathematical laws. For complex conditions, such as in casting work, it is not possible to solve the mathematical equations of the equivalent electric circuit; but if the analogous electrical process is carried out, nature makes the solution for us, and by measur-

ing the electrical results, we can obtain the results of the thermal analogous problem. In actually building the electrical analogous circuit one "trick" is used, which is common in electrical engineering: that of sectioning or "lumping." One thinks of the body, in which heat flow occurs,—casting and mold—as being divided into a number of sections or lumps; and within each lump we consider the thermal properties to be concentrated in axis and center (the specific heat and heat of fusion in the center; the thermal resistance in the axis).

Some details of the technique are shown in the appendix, and can be omitted by the reader, not interested in them: he need no more understand the intricacies of this computing device in order to make good use of its results, than he need to know the details of the working of a bus, before boarding it to get to his destination.

Here several general aspects of this method should be considered.

a. The method allows us to apply a time scale. The length of the individual electric "computation" experiments is usually in the order of magnitude of between 5 and 15 minutes. Solidification of very thin castings, which freeze in much shorter times (e.g. die castings) can be studied, by "blowing up" the time by taking a "slow motion picture" of the happenings; by applying a time scale studies of very large castings, which may freeze only in hours are reduced to last a few minutes.

b. As in any computing technique, one can arbi-

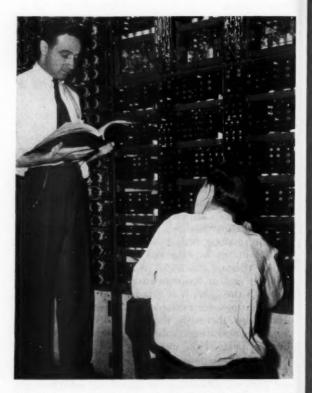


Fig. 7—Part of the Frames with Capacitances and Resistances.

tran put men if, e stud mol of c age. tests in at

"cor desi limi ing arbi one viou anal sand on wou

sand

d.

duc

occi

ORK

s of

ling

sed,

t of

. 111

eing

ith-

s to

neat

nce

the

ter-

tri-

ood

s of

his

uld

ale.

on'

of

hin

die

me

igs:

igs,

t a

bi-

id

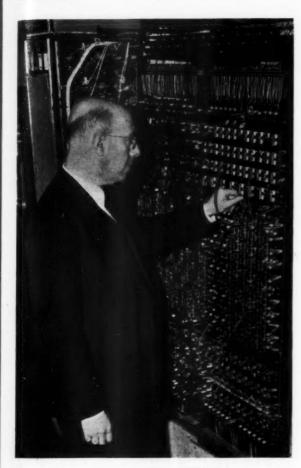


Fig. 8-Part of the Instrument Frame.

trarily limit variations of conditions in any one computation to one variable, whereas in direct experiments accidentally more than one variable changes; if, e.g. the influence of pouring temperature is to be studied, variations of composition of dryness of the mold of sand properties mar the picture; a number of castings have to be made to obtain a statistical average. In any computing technique this need of many tests, required to allow statistical averaging, is eliminated.

c. As in any computing technique, one can introduce values for the variables, independent of their occurrence in nature. For example, one can introduce "complete insulation" (no heat transfer) across any desired face; or "zero thermal resistance" as another limit. Thus one can establish what happens in limiting cases; moreover, this possibility of selecting any arbitrary value for the different properties enables one to explore desirable materials. Consider the previous example of sand conductivity, in case of the analog computer one need not attempt to find many sands with varying conductivities. One can determine on the computer first, what value of conductivity would be desirable, and then later limit the search for sand to the one with approximately the best value of conductivity.

d. One can measure temperatures and rates of heat

flow, accurately at any desired point in the body. There are no difficulties of inserting thermocouples, speed of the instrument etc.

The Heat and Mass Flow Analyzer at Columbia University is such an analog computer, it looks like a large telephone exchange⁸ (Fig. 7 and 8), consisting of racks with electrical equipment and measuring instruments. Provision is made for a rapid change of circuits in order to allow presentation of many different problems. Problems regarding the foundry industry have been carried out over the last seven years in a cooperative program with the American Foundrymen's Society.

In the following section some of the work carried out on foundry problems so far on the Heat and Mass Flow Analyzer is reviewed.

IV. Thermal Studies with the Heat and Mass Flow Analyzer

1. First Comparison Between Analyzer and Casting Work-In 1944 the American Foundrymen's Association set up a Heat Transfer Committee which supervised the work reported hereafter from the foundry aspect. The Committee decided first to investigate prediction of solidification rates found from "bleeding tests" ("pour out tests") made by K. L. Clark. On the Analyzer an "infinite" slab was investigated, having no end or corner effects. Clark used slabs of 2 in. x 6 in. x 2 in., 4 in. x 8 in. x 8 in. and 6 in. x 12 in. x 12 in., to obtain data on different thicknesses assuming that the size of the slabs is sufficient to preclude end effects. Castings were made in part in sand, in part against a chill which was backed by sand. Properties of steel and sand which had to be used in the electric computation experiments were selected as carefully as possible. For details, reference is made to the original publication.9 Here it is sufficient to deal with a few

a. All concerned expected to find correlation between the bleeding tests and the Analyzer curve for the "solidus." To understand this remark it should be remembered, that the Analyzer produces merely curves, relating position, time and temperature, but has no way of indicating in what state the metal is at any given point or temperature. The latter information is determined by metallurgists.

Thus one curve may be taken showing for each position the time necessary to reach the liquidus and another curve showing the time to reach the solidus. Contrary to expectation the curve obtained from the bleeding test correlating solidification thickness with time after pouring checked well with the Analyzer curve for the liquidus (see V-1).

b. Because of the high thermal resistance of the sand, the additional resistance of the air gap has relatively small influence on the rate of solidification; thus, even if such gap is assumed to exist, the time, when it is formed, does not appear to be significant.

In case of casting against a chill however, the added thermal resistance of the gap is large compared with that of the chill. Hence, the estimated time when the air gap forms is important. On the Analyzer this time can be selected arbitrarily; that is one can easily obtain results for different estimated times of

"air gap formation." It was found, that if the time was assumed to vary with the square of the thickness of the cast slab, satisfactory check with Clark's work was obtained.

c. The thermal resistance of the air gap between outside surface of the casting and the inside surface of the mold depends on the difference of the fourth powers of the temperatures of these two surfaces. It was found that this difference was quite constant and therefore the thermal resistance of the air gap does not change much during solidification.

2. Influence of Properties and Pouring Temperature—As the next step, computations on the Analyzer were carried out, in which all steel properties and the pouring temperature were varied arbitrarily over a wide range. For details, again reference is made to the original paper.¹⁰ The following observations may be of interest:

a. Times to reach both liquidus and solidus change markedly with pouring temperature (the latter being understood, as the temperature at which the steel enters the mold, the filling time of the latter being assumed to be zero).

b. As far as reaching the liquidus is concerned, heat of fusion has but little influence. Times change, very roughly, proportionally to specific heat and conductivity, for the latter slightly less than for the former. Solidification range has great influence.

c. As far as reaching the solidus is concerned, solidification range is of only small influence; heat of fusion changes the time to reach the solidus greatly. Conductivity has almost no influence, and specific heat has a smaller influence on the time to reach the solidus than on those to reach the liquidus.

What do these findings mean for the practical foundryman? Inasmuch as properties for different steels do not vary greatly, he can transfer his experience from one composition to the other with reasonable accuracy. But he must pay strict attention to the pouring temperature and in practice also to the pouring time. The latter influences the temperature with which the steel ultimately reaches the mold; and this temperature is the one, which decides solidification time.

3. Freezing of Steel Spheres—Proceeding with experiments with steel (chronologically other experiments were taken first), a series of tests on steel spheres was run.¹¹ These tests served to compare Analyzer work against a previous publication by Briggs and Gezelius.¹² Based on the recognition of the importance of the pouring temperature (see 2) experiments were run separately for different pouring temperatures.

a. The check with the values of Briggs is good, except in case of a pouring temperature of 2740 F. With so small a degree of superheat the minor changes in pouring temperature will result in great changes of solidification times. Hence, conceivably small errors in determining the "2740" by Briggs might account for the discrepancies between his and the Analyzer's findings.

5. By using the experiments for comparison, one can compare the time for the center to reach the

liquidus and the solidus in the case of a slab and in the case of a sphere. The original paper contains the comparison of values for reaching the liquidus; the comparison of those for the solidus have been computed in the meantime. Both for slab and for sphere the solidification times increase with pouring temperatures. But the increase, percentage wise, is less for the sphere, than for the slab. The difference between slab and sphere is less marked for the solidus (9.6/6.5 = 1.48), than for the liquidus (3.27/1.47 = 2.22), (see table).

Pouring Temp., °F	Time to reach the Liquidus, sec.			the Solidus,		Ratio of Times Slab/Sphere	
2740	316	215	1.47	2930	450	6.5	
3010	1105	338	3.27	5800	600	9.6	

4. Comparative Solidification Studies—Having data on spheres and slabs, comparative studies including cylinders were made, particularly in view of the paper by N. Chworinoff¹⁸ which is finding considerable interest in America.

Chworinoff states the following:

a. Any two castings having the same ratio of volume/area (V/A) should solidify at the same time.

b. The time for complete solidification of two castings varies proportional to the square of the ratio (V/A).

Accepting the liquidus as freezing point, as seemed

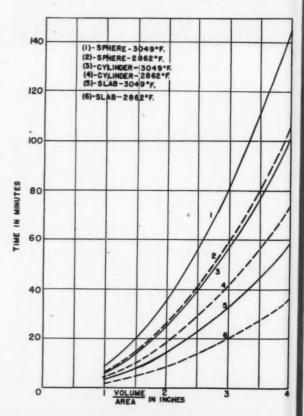


Fig. 9-Time to Reach the Liquidus for Different Shape Factors (volume/area).

FOR CENTER OF SLAB TO REACH

to

th

(C

fo

CO

re

all

th

sh

lo

qu

gra

sai

an

on

sla

fec

in

Fig.

the
a.
tion
liqu
the
for
k =

the Thu for b.

that

d in the the

ORK

con)here perfor

veen /6.5 .22),

les here data

of

ling aper iter-

voltwo the

med

ent

obvious from the bleeding tests, it was found that Chworinoff's relationship does not hold.

Figure 9 shows the summary of the experiments. Values of volume/area were plotted as abscissa, times to reach the liquidus were plotted as ordinates.

Six curves are shown, three of which hold for a pouring temperature of 3049 F (curve 1, 3, 5) and the other three for a pouring temperature of 2862 F (curve 2, 4, 6). Two curves (1 and 2) hold for spheres two for cylinders (curves 3 and 4) and the last two for slabs (curves 5 and 6).

No further discussion of this set of tests will be made here because of the recognition obtained since completion of the tests that bleeding tests are not

reliable in determining solidification.

5. Influence of Sand Conductivity (Dry Sand)—In all investigations up to that time the conductivity of the sand was considered to be constant, namely, k = 0.9 Btu/ft, hr, °F.

Investigations, mainly by Lucks and co-workers14 showed that conductivities of sand might be much lower than this value and that conductivity changes quite a bit with temperature. Therefore one program15 was carried out to investigate the influence of sand conductivity on solidification rates of steel. The analysis was made for a steel slab 4 in. thick, covered on either side by a layer of sand 5 in. thick. The slab was considered to be so long that no end effects occur. The results of the tests are summarized in Fig. 10. Figure 10 shows the times for the center

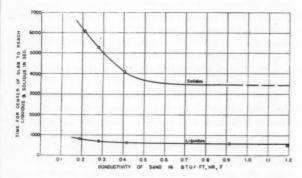


Fig. 10-Influence of Sand Conductivity (Dry Sand) on Solidification Times.

of the slab to reach the liquidus and the solidus (two curves) plotted against the average conductivity of the sand. Two points are of interest.

a. The influence of conductivity on the solidification time is greater for the solidus than for the liquidus: comparing the times for the center to reach the solidus we find a ratio of 6050 $\sec/3400 \sec = 1.78$ for the values of sand conductivities of k = 0.9 and k = 0.208 respectively; whereas for the times to reach the liquidus the ratio is only $780 \sec/570 \sec = 1.36$. Thus the sand properties are of considerable interest for complete solidification.

b. The influence of the conductivity becomes more pronounced at conductivity values below 0.65. Above that the changes are much less marked.

c. The conductivity of all sands increases markedly with temperatures. This is understandable because the term "conductivity" for sand is an approximation: the figure actually represents an apparent conductivity including the phenomena of conduction between individual sand particles and radiation across the interstices between the sand particles; the radiation effect obviously increases at elevated temperatures.

Therefore in computations the change of sand conductivity with temperature should be taken into account; or, where this is too cumbersome, great care must be taken in selecting the average conductivity.

6. Influence of Moisture in Sand on Conductivity-A later step in our investigation deals with the influence of moisture on the solidification. When casting in moist sand, the moisture is driven towards the cold surface. This occurs by evaporating the moisture in layers near the hot surface; the steam thus formed flows through the pores and condenses as soon as it reaches portions of the sand which are at temperatures below the boiling point of water. The heat of evaporation acts as a heat drain: more heat flows to a section of sand at the same temperature at which no evaporation takes place. The apparent conductivity of the sand is thus increased.

The representation of this process on the Analyzer is quite complex; therefore in the computation carried out on the Analyzer to date, a constant surface temperature on the casting-sand interface was assumed, whereas actually this temperature changes during solidification.

For this simplified condition (constant interface temperature) it is found, that the influence of moisture

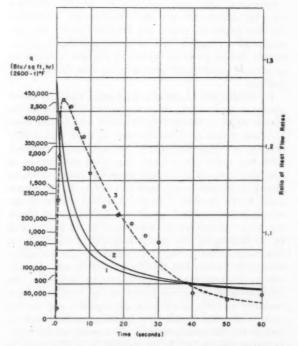


Fig. 11-Influence of Moist Sand Conductivity on Heat Extraction From a Face Held at Constant Temperature.

disappears after relatively short time and that the influence is more marked at lower temperatures than at high ones. Figure 11 expresses the ratio of heat extraction from the "ideal" casting ("ideal" because the interface between casting and mold is assumed to remain at constant temperature) in moist sand to that in dry sand of otherwise unchanged properties; this ratio is plotted against time.

It is obvious that moisture will exert an influence on the formation of a solidification skin but will hardly influence freezing to great depth.

Additional work on this problem is being carried out.

7. Experiments For Materials Other Than Steel—For the sake of completeness it should be mentioned that work has also been carried out on the rate of solidification of white cast iron¹⁷ and of aluminum.¹⁸ In case of white cast iron, check with bleeding tests is poor.^{17, 18} But it was found that the bleeding tests themselves were difficult to carry out and in view of the later research and observations regarding bleeding tests (V-1) the validity of the bleeding tests themselves may be questioned.

The check in case of aluminum was good. These tests are worth mentioning because aluminum, as pure metal, freezes at one temperature rather than over a temperature range as steel and other alloys do. Freezing at constant temperature requires a different technique on the computer; the fact that aluminum tests also match direct observations is an added proof for the correctness of the Analyzer.

V. Some Open Problems

Obviously the seven years work on the Analyzer may be considered only as a beginning. Therefore the two problems which are discussed below are not meant as a complete list of unfinished business but merely should point to regions of inquiry which seem at present particularly pertinent in view of the latest findings.

1. Solidus and Liquidus as Freezing Point-As stated above in Section IV, the comparison of temperature curves taken on the analog computer with results from bleeding (pour out) tests seemed to show that solidification is completed when the liquidus is reached. This was quite unexpected; but good correlation of Analyzer tests with observations of three different observers seemed to preclude any error. In the meantime, R. W. Ruddle,19 in his excellent review of the solidification problems questions the validity of bleeding tests. Almost at the same time W. S. Pellini and H. F. Bishop at the Naval Research Laboratory in Anacostia, Washington, D. C., raised similar questions. Moreover, Pellini and Bishop verified by actual temperature measurements (discussion to Ref. 15), the correctness of the curves found on the Analyzer. They included in their check both the liquidus and solidus curves; in other words they found that the times required for different points in the casting to reach the liquidus and to reach the solidus were the same as those predicted on the Analyzer.

In view of those two observations (Ruddle, and Pellini and Bishop) one seems justified to disregard bleeding tests as means to predict solidification rates and to rely only on temperature measurements and predictions.

Accepting this, two questions arise.

1. Comparative Solidification Times—Chworinoff¹⁸ makes two statements in this connection, both related to the ratio volume/area which may be called shape factor (S).

a. Bodies having the same shape factor solidify in the same time independent of their size or individual

b. Solidification times vary proportionally to the square of the shape factor.

In previous publications from the Analyzer laboratory, it was stated that his predictions do not check with the findings on the computer. This statement needs re-examination since it was based on the assumption that the freezing point coincides with the liquidus. Accepting the solidus as freezing point may influence the statement.

Chworinoff in his Fig. 19 presents a graph in which solidification times are plotted against the shape factor; he uses a log-log scale and obtains for a large number of different castings a straight line relationship between ordinate and abscissa, which in view of the scales used, indicates validity of his square law (b). The range of shape factors examined by Chworinoff is very wide: the lowest value is S=0.19 and the largest value which can be read on his graph definitely is S=5 near the upper end of the abscissa scale. There are a few points still beyond the S=5 but no dimensions are shown for these other readings. Based on his scale it should be S=7.5. The values given here (0.19, 5, and 7.5) are in inches whereas the original graph is in millimeters.

The ratio of the largest to the smallest S value on Chworinoff's chart is 5/0.19 = 26.

Tests on the Analyzer are available so far only for slab and sphere. The shape factor, S, has the dimension of length, and can conveniently be expressed as multiple of the smallest dimension, D, of the shape. In case of sphere or cylinder, D denotes the diameter; in case of a large slab (with no edge effects) it denotes the thickness and otherwise the smallest dimension.

In case of finite slabs, S approaches zero as D becomes smaller and smaller: the surface of a sheet may be quite large, but if the thickness is small the volume becomes small too.

For a large slab (no edge effects) one finds readily $S_P=D_P/2$. For a sphere the value is $S_8=D_8/6$. The subscripts P (plate) are used for the slab and S (sphere) for the sphere. Hence the ratio $S_P/S_8=3$.

Examining any pair of slabs or spheres of the same dimension D, will result in a ratio for values S of 3:1.

For the same shape—whatever the shape may be, sphere, cylinder, slab, etc.—the "square law" is correct provided that all dimensions including those of the mold changed proportionally. That means that if a sphere of 3-in. diameter is compared with one of 6 in., the solidification time of the latter will be four times as long as that of the former; provided that the dimensions of the mold have changed in the same

th

ce

re

ORE

and

ff1:

ted

ape

in

ual

the

ora-

eck

ent

the

nay

ich

fac-

rge

on-

of

law

vor-

and

aph

issa

= 5

ngs.

ues

eas

on

for

en-

as

ipe.

ter;

de-

di-

be-

nav

vol-

dily

4/6.

d S

3.

me 3:1.

be,

cor-

hat

e of

our

the

une

ratio and provided the same temperature and same metal is being used.

For different shapes however, so far only a comparison between slab and sphere is possible having as mentioned, a ratio of shape factors of 3. A comparison has been carried through of a slab of 4-in. thickness (no edge effects) with a sphere of 12-in. diameter. The two having the same shape factor should freeze in the same time. Their solidification times, taken as the times for the center to reach the solidus are shown in Table 1; the first column contains the pouring temperature, the second the solidification time of the sphere, the third that of the slab and the fourth the ratio of the two. As stated, according to Chworinoff the solidification times should be identical, or this ratio should be one. It will be seen that at high pouring temperatures this ratio as postulated by Chworinoff is closely maintained whereas at low temperatures the difference between the actual ratio and the postulated ratio is quite appreciable.

TABLE 1—COMPARISON OF SOLIDIFICATION TIMES OF A LARGE SLAB (NO EDGE EFFECTS) WITH A SPHERE, BOTH HAVING THE SAME VALUE OF VOLUME/AREA

Pouring Temp.,	Times to Reac	Ratio	
°F	Slab	Sphere	Slab/Sphere
3010	4600	4320	1.06
2900	3900	3920	0.99
2820	3280	3520	0.93
2800	3100	3400	0.91
2740	2375	3280	0.73

It must be remembered that the accuracy of tests at small degrees of superheat is reduced. But even discounting this possible source of error, and accepting also the ratio of 0.73 at 2740 F as correct, it is significant to compare this value with the one obtaining for the times to reach the liquidus. At a temperature of 2740 this ratio is only 0.16.

Thus it is possible that Chworinoff postulated his rate based on observations at relatively large degrees of superheat only.

2. Air gap—The formation of an air gap in case of casting against chill causes a considerable change in rate of solidification. So far it has been difficult to determine the time when such air gap forms. Since the technique of temperature measurements in liquid steel have been developed and are practical, as Pellini and Bishop show, it may be possible to use the Analyzer to determine the time for formation of the air gap. This time has considerable influence on the rate of solidification, although percentage wise it is less important than previously assumed. The air gap time influences the time to reach the liquidus more, than it does the time to reach the solidus.

For example, in case of a 2-in. casting backed by a 3-in. chill, which in turn is backed by 5 in. of sand, the time to reach the liquidus increases from 180 seconds to 270 seconds, if the air gap time is reduced from 36 to 9 seconds. The increase is roughly 50 per cent. The same decrease of air gap time changes however the time for complete solidification (time to reach the solidus) only from 420 to 540 seconds; a change of 28 per cent.

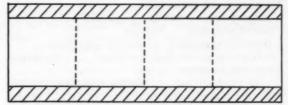


Fig.12 -Rod Subjected to Heat Flow.

APPENDIX

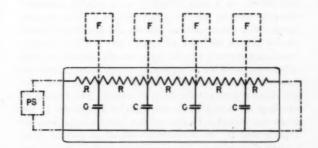
In order to explain the working of the Heat and Mass Flow Analyzer,20 consider a rod, Fig. 12, which is completely insulated on the sides, so that heat flow can occur only parallel to the axis. Let the rod be initially at constant temperature throughout and at time zero a high temperature be impressed on one end while the other is maintained at its original temperature. Then the rod will slowly increase in temperature until, in steady-state, a straight line temperature distribution occurs with the highest temperature at the hot surface and the lowest at the cold surface. The differential equations for the temperature rise with time at any point in the rod are the same as the differential equations for the increase in electric potential if an electric conductor with evenly distributed resistivity and capacity is suddenly exposed to a voltage difference.

Therefore one could study the time-temperature history by observing the potential increase in a suitable electric conductor.

Instead of attempting to find such a conductor with the appropriate properties, the "lumping technique" is used which has been mentioned in the text. This consists in considering the rod being divided into sections or lumps as indicated in Fig. 12. Within each section the resistance is considered to be concentrated at the axis, and the capacitance in the center.

Thus a simple electric network as shown by the solid lines in the "box" of Fig. 13 represents the rod shown in Fig. 12.

Between the electric computing circuit and the heat flow problem certain analogies exist, which are obvious from the identity of the equations.



PS-Power Supplies

R -Resistors representing resistance to heat flow

C -Capacitance representing heat storage capacitance

F -Devices to feed current in representing internal heat sources

Fig. 13-Equivalent Electric Circuit to Represent Rod Shown in Fig. 12.

These analogies are as follows:

temperature : voltage (potential)

rate of heat flow : current

thermal capacitance : electric capacitance

(volumetric specific heat x volume) : (condenser) thermal resistivity

(1/conductivity) : electric resistivity

To the circuit shown in the box of Fig. 13 various elements have to be added to represent different conditions. For example, if the rod is heated up as described above, connections are made as shown by dashdot lines in Fig. 13. If instead, the rod is heated by internal heat sources, such as dielectric heat and cooled to the outside, a circuit would be fed as shown by the broken lines in Fig. 13, and closed as shown by the dash-dot lines omitting the power supply PS.

If heat flow is in more than one direction, a network of resistors has to be used. One element of such network for two-dimensional heat flow is shown in Fig. 14a, for three-dimensional heat flow in Fig. 14b.

Fig. 14—(Left) Element of Network to Represent Two-Dimensional Heat Flow; (Right)—Three-Dimensional Heat Flow.

Further refinements refer to the changing of properties between liquid and solid. Instead of having in each section one capacitance and two resistors (one on either side of the capacitor) each section has several resistors and capacitors. Figure 15, for example, shows one section each of the steel and mold circuits for a one-dimensional casting experiment. (A onedimensional experiment is one for a slab, disregarding end effects, or a long cylinder, or a sphere. Actually the heat flow in the two latter shapes is three dimensional. For mathematical reasons the one-dimensional circuit can be used for all three cases.) Three capacitances are shown: one to represent the specific heat in solid state, another to represent the difference between heat storage capacities in liquid and solid state and the third represents the heat of fusion. The release of the heat of fusion while the metal temperature drops from the liquidus to the solidus temperature may be represented by an increase in apparent specific heat during that period. The resistances next to the capacitance represent the thermal resistance in solid state, and the outside resistors the difference between the thermal resistance in liquid and solid state.

As the temperature (voltage) of each section decreases from the initial "pouring temperature," the following changes have to be made.

When the voltage (temperature) reaches the liquidus the "heat of fusion capacitance" is inserted into the circuit.

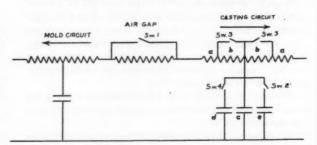
When the voltage has dropped to the mid-point between liquid and solid values, the properties (resistance and capacitance) are changed from their "liquid" values to their "solid" values. This implies shortening the two other resistances, because the thermal conductivity of liquid metal is higher than that of solid metals; it also implies the disconnecting of the difference capacitances which represents the difference between liquid and solid specific heat.

When the voltage drops to the solidus point, the "heat of fusion capacitance" is taken out of the circuit

In case of pure metal no "heat of fusion capacitance" is inserted. Instead, when the voltage (temperature) of any section has dropped from the pouring value (temperature) to the solidification point it is connected to a constant voltage supply which feeds current into the circuit at that point. The current represents the rate of heat generation by the latent heat of fusion. This power supply is held in contact until enough current has drained into the circuit to represent the entire heat of fusion.

An air gap between casting and mold is represented by a resistance between the end of the casting circuit and the beginning of the mold circuit.

As mentioned in the text a time scale can be introduced when using the Analyzer. This is done by appropriate choices of electric resistances and capacitances to represent given values of thermal properties.



Resistors a represent thermal resistance in solid state Resistors b represent difference in thermal resistance between liquid and solid state

Capacitor c represents heat storage capacity in solid state

Capacitor d represents the difference in heat storage capacity

between liquid and solid state

Capacitor e represents heat of fusion Switch (SW) 1 is closed initially and is opened as soon as the air gap forms

Switch (SW) 2 is initially opened and is closed when the voltage (temperature) has dropped to the value of the liquidus

At a voltage half way between that representing liquidus and solidus switches (SW) 3 are closed, thus representing an increase in conductivity upon solidification and switch (SW) 4 is opened according to a decrease in specific heat.

Capacitances c and d are precharged before the experiment to a voltage representing the pouring temperature. Capacitance e is precharged to a voltage corresponding to the liquidus; the capacitance in the mold circuit is either uncharged or, in case of a preheated mold, precharged to a voltage corresponding to the mold temperature.

Fig. 15-End Sections of "Casting Circuit" and "Mold Circuit."

ie

25

le

n

g

e

e

r-

r.

g

ls

ıt

t

t

0

d

t

ì

Physically the laboratory consists of a large number of resistors and capacitors so arranged that an unlimited number of different connections can be made conveniently and in short time. In addition, of course, the necessary power supply and measuring devices are provided.

Acknowledgments

The author is indebted to the members of the Heat Transfer Committee and particularly to Dr. H. A. Schwartz, Chairman of this Committee since its inception, who introduced the author to the field of foundry work and who showed continued interest and support. He and Mr. E. C. Troy also were kind enough to read the manuscript.

Thanks are also due to the American Foundrymen's Society which as first organization in this country took up a systematic study of heat transfer in its field of operation by means of the analog computer. The items 9, 10, 11, 15, 16 and 18 in the bibliography represent parts of reports issued by the Heat Transfer Committee as results of work sponsored cooperatively by the American Foundrymen's Society and Columbia University.

Bibliography

- C. Schwarz, Zeit. fur Ang. Math, u Mech., vol. 13, p. 202 (1933).
- N. M. H. Lightfoot, Proc. London Math. Soc. (2), vol. 31, p. 97 (1929).
- 3. E. Schmidt, (a) August Foeppl Festschrift, Berlin 1924, (b) Forschung auf dem Gebiete des Ingenieurwesen, 13 (5), (1942).
- 4. A. Nessi and L. Nissole, (a) Methodes graphiques pour l'etude des installations de chauffage, etc., Dunod, Paris, 1929

- (b) Resolution Pratique des Problemes de Discontinuite, etc., Dunod, Paris, (1933).
- 5. R. V. Southwell, "Relaxation Methods in Engineering Science," Oxford University Press, New York (1940).
- 6. H. W. Emmons, "The Numerical Methods for Transient Heat Flow," Trans., ASME, vol. 65, No. 6, (1943).
- 7. G. M. Dusinberre, "Numerical Analysis of Heat Flow," McGraw Hill, New York (1949).
- 8. V. Paschkis, "The Heat and Mass Flow Analyzer Laboratory" Metal Progress, vol. 52, p. 813, November (1947).
- V. Paschkis, "Heat Flow Problems in Foundry Work," TRANSACTIONS, A.F.S., vol. 52, pp. 649-670 (1944).
- 9. V. Paschkis, "Studies on Solidification of Castings," Transactions, A.F.S., vol. 53, pp. 90-101 (1945).
- 10. V. Paschkis, "Influence of Properties on Solidification of Metals," Transactions, A.F.S., vol. 55, pp. 54-61 (1947).
- 11. V. Paschkis, "Study on Solidification of Steel Spheres," Transactions, A.F.S., vol. 56, pp. 373-378 (1948).
- 12. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction in Steel Castings," Transactions, A.F.S., vol. 43, p. 274 (1935).
- 13. N. Chworinoff, "Theory of the Solidification of Castings" Die Giesseri, British Iron and Steel Institute, Translation No. 117, vol. 27, pp. 177, 201, 222, (1940).
- 14. C. F. Lucks, O. L. Linebrink and K. L. Johnson, "Thermal Conductivity of a Sand Mixture," Transactions, A.F.S., vol. 56, pp. 363-365 (1948).
- 15. V. Paschkis, "Influence of Dry Sand Conductivity on Rate of Freezing of Steel Slabs," Transactions, A.F.S., vol. 58, pp. 147-152 (1950)
- pp. 147-152 (1950). 16. V. Paschkis, "Heat Flow in Moist Sand," Transactions, A.F.S., vol. 59 (1951).
- V. Paschkis, "Studies of Solidification of White Iron Castings", Transactions, A.F.S., vol. 56, pp. 371-373 (1948).
- V. Paschkis, "Studies on the Solidification of Aluminum Castings," ibid, pp. 366-371.
- 19. R. W. Ruddle, "The Solidification of Castings," Institute of Metals, 1950.
- V. Paschkis and H. D. Baker, "A Method for Determining Unsteady State Heat Transfer by Means of Electrical Analogy," Trans., ASME, vol. 64, p. 105 (1942).

BASIC CUPOLA MELTING AND ITS POSSIBILITIES

Ву

E. S. Renshaw*

ABSTRACT

The use of a basic-lined cupola as part of a production unit is described and experience gained in operating with basic slags is discussed. Slag conditions which favor desulphurization to low limits also give high carbon pick up in low-carbon charges and the possibility of taking advantage of these factors is discussed with particular reference to nodular iron production. Additional refractory cost may prove to be justified if subsequent processing cost is lowered by reducing the quantity of alloy required for ladle treatment. Reference is made to water-cooling as a means of reducing refractory erosion and offering the possibility of operating with slags of higher basicity.

THE IMPORTANCE OF THE CUPOLA FURNACE to the foundry industry cannot be overstressed and the economic advantages of this type of furnace will ensure its continued use for many years to come. By virtue of its apparent simplicity it has, in general, fulfilled the requirements of the industry and being continuous in operation, has fitted into the pattern of molding continuity. The main limiting feature of the cupola and the factor which distinguishes it from the bath-type furnace is that chemical reactions cannot be invoked at will within the furnace itself. Consequently the required metal composition is obtained by simple calculation of charged elements, allowances for inevitable losses and gains of these elements being made on the basis of previous performance under a given set of standardized conditions.

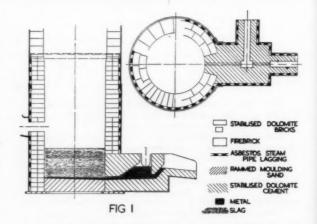
The established practice of acid cupola melting has resulted in little interest being taken in the effect of slag composition and the role of slags during the melting operation. Work carried out on the basic cupola on the other hand, has revealed the importance of slag chemistry and the possibility of promoting desirable reactions not hitherto obtainable. Published reports have shown that sulphur reduction by basic slag operation is substantial and controllable. Indications of considerable carbon absorption, well in excess of that obtainable in the acid cupola, have been

reported but control of the extent of this pickup has not been studied in detail. Dephosphorization has been attempted but with limited success and much work remains to be done on this aspect before a commercial process could be developed.

It is the purpose of this paper to give details of an example where the basic cupola has been successfully employed; to report observations as a result of production experience, assessing its value for future application and indicating the direction in which development might usefully proceed.

Production Cupola

The earlier work on basic melting conducted in the author's foundry¹ resulted in the method being adopted for production and a basic cupola has been in constant daily use for several years. Details of the cupola lining are shown in Fig. 1. Stabilized dolomite



bricks form the basic-lined portion which extends from the sole plate to 461/2 in. above a single row of tuyeres. Circle shapes are used except for a section above the tuyeres, that is in the melting zone where crown shapes are substituted, giving additional support as erosion takes place. The tapping hole and continuous slag separating box are rammed with stabilized dolomite cement suitably prepared in ram-

Official Exchange Paper from the Institute of British Foundrymen to the American Foundrymen's Society. It was sponsored by the Gray Iron Division and presented at a Gray Iron Session at the 55th A.F.S. Convention in Buffalo, April 26, 1951.

u

Ti Si

tı

h

th

in

^{*} Ford Motor Company, Limited, Dagenham, England.

h

1

ming form by mixing with 6 per cent water and 1 per cent sodium silicate.

Expansion characteristics of stabilized dolomite are such that suitable allowance must be made in the construction of the lining and this is provided for by placing a layer of asbestos steam pipe lagging, 1 in. thick, between the shell and basic refractory. Stabilized dolomite cement is used for daily repairs and this material when mixed with water has the patching consistency of ganister and is simple to apply. Further details of cupola dimensions are given in Table 1.

TABLE 1-PRODUCTION CUPOLA DIMENSIONS

Internal diameter, in.	42
Lining thickness, in.	9
Well depth from sand bed to tuyeres, in.	29
Tuyere level to charging door, ft.	165/12
No. of tuyeres	6
Tuyere dimensions, in.	6 x 6

The cupola in question is operated in conjunction with an electric furnace as a duplexing unit2 and desulphurization being the only objective, basic melting has enabled the equipment layout to be such that the cupola metal is able to flow continuously from the tapping spout direct into the electric furnace in the shortest distance with a minimum delay. Ladle treatment with soda ash between the cupola and electric furnace would result in loss of time and metal temperature, the heat loss having to be recovered by current and electrode in the electric furnace. The unit operates on a continuous pouring basis, the metal required being a low-carbon alloy of the graphitesteel type with a carbon content of 1.20 to 1.45 per cent. The carbon in the cupola metal is held between 2.5 and 3.0 per cent and the finished carbon is obtained by regular additions of steel scrap to the electric furnace bath. The total metal charged into the electric furnace consists of, approximately, 60 per cent solid steel scrap and 40 per cent molten cupola metal, shop returns, runners, risers, etc., being adequate to provide all the metal for the cupola charge, details of which are given in Table 2.

TABLE 2-PRODUCTION CUPOLA CHARGE DETAILS

12	200 Lim	estone		90
lb	8 Fluc	orspar		10
	Cok	e		135
elting Rate	9000 lb/hr			
T.C.	S	Si	P	Mn
% 1.40	0.050	2.40	0.050	0.50
2.60/2.90	0.040/0.070	1.90	0.050	0.40
	elting Rate T.C. % 1.40	8 Fluor Coke	8 Fluorspar Coke elting Rate 9000 lb/hr T.C. S Si % 1.40 0.050 2.40	8 Fluorspar Coke

As the sulphur content of the metal is not required to be below 0.060 per cent, the flux and coke charges have been standardized accordingly. The previous acid practice with this metallic charge and coke of the same quality gave a sulphur content of between 0.12 and 0.16 per cent in the molten cupola metal.

Production experience has enabled some interesting observations to be made on composition control and afforded a theoretical and practical basis on which control can be established.

Sulphur Removal

The transfer of sulphur from molten metal to slag depends on the presence of sufficient base in the slag to convert the sulphides in the metal to basic sulphide, insoluble in the metal and soluble in the slag.

In the basic cupola, a larger quantity of base can be introduced than is possible in the acid-lined furnace and the sulphur carrying capacity of the slag produced is thereby increased. The transference of sulphur from metal to slag also depends on the ratio of CaO to FeO and the presence of the latter in the slag phase limits the desulphurizing power thus:

Ca S + 2 FeO = CaO.FeO + Fe S.

Therefore, the desired reaction is one in which carbon plays an essential part represented by the following:

FeS + CaO + C = Fe + CaS + CO

It has been shown that slag in the melting zone of the cupola is rich in iron oxide,⁸ the proportion of which may be dependent on a number of well known operating factors. On collecting in the well, however, this oxide is reduced by the incandescent coke, the extent depending on the time of contact, volume of slag and no doubt, temperature and slag viscosity, thus creating the correct slag condition suitable for desulphurization.

Experience in operating with basic slag has shown this theory to apply in practice and the sulphur pickup may be controlled accordingly. The first metal tapped from the cupola is likely to be high in sulphur and this may be accounted for by low initial operating temperature with insufficient slag volume in the well. Higher initial coke bed and limestone added to the coke bed itself will reduce this condition to a minimum. Any circumstance, giving rise during melting to oxidizing conditions, such as low coke bed height results in a noticeable increase of the iron oxide content in the slag, accompanied by an increase in the sulphur content of the metal. As in cupola practice generally, the normal precautions are taken to ensure correct conditions of combustion, as over-blowing obviously results in excessive oxidation and increased sulphur pickup.

If for any reason a loss of slag volume in the furnace well occurs, thus reducing the time of contact between the slag and coke, there is an immediate increase in the iron oxide content of the slag and desulphurization is reduced in consequence. In the continuous tapping system shown in Fig. 1, the slag volume in the well is determined by the metal level at its exit from the box and should this level be too low, the slag depth in the well is reduced. Enlargement of the cupola tapping hole through erosion also results in loss of slag height in the well, so that refractory practice at this location is of importance.

The nature and color of a basic slag affords a useful and rapid check on its desulphurizing properties. The shade may vary from white, creamy white, pale brown, dark brown to black with chalky to vitreous fractures. Maximum desulphurization is certain with white chalky slags and an increase in iron oxide content may be immediately observed by a change to the darker shades, vitreous fractures being evidence of

shortage of base. For control purposes, samples of slag for visual examination are taken under standard conditions of cooling as the cooling rate effects the appearance and small oil sand cores have been found suitable for sampling from the slag stream.

In the practice described above, the normal slag analysis is as follows:

	Per Cent
CaO	42.0/45.0
MgO	9.0/11.0
FeO	Less than 1.50
MnO	Less than 1.50
Al ₂ O ₃	8.00/10.0
SiO ₂	30.0/35.0
S	up to 1.0
CaF ₂	2.0

Any slag within this range of composition will be white to pale brown in shade with a chalky fracture. An increase in iron oxide and manganese oxide above 1.50 per cent will result in a progressive color change from brown to black. This method of slag examination offers a much more positive guide to melting conditions than a similar examination of acid slags, as in the latter case changes in composition are not so readily discernible.

Silicon Loss

Whereas silicon loss in conventional acid practice is usually no greater than 5 to 10 per cent, allowance must be made in basic practice for a higher loss which in the above production cupola is 20 per cent. It has not been found possible to limit this silicon oxidation, the extent of which may be determined by the degree of slag basicity. Under conditions which exist in the cupola melting zone, silicon is first oxidized according to the following equation:

$$Si + 2 FeO = SiO_2 + 2 Fe$$

The SiO₂ thus formed may be neutralized immediately by CaO as follows:

$$SiO_2 + CaO = CaO. SiO_2$$

or, with an excess of FeO, a ferrous silicate may result according to the equation:

 $SiO_2 + FeO = FeO. SiO_2$

followed by the CaO reaction:

FeO. $SiO_2 + CaO = CaO$. $SiO_2 + FeO$

It is not unlikely that in the absence of an excess of base, the silica formed by oxidation could be reduced by carbon and re-enter the metal, whereas with ample base, the more stable calcium silicate is formed. As these reactions are an inevitable part of the melting process, silicon control must be based on standardization of melting conditions and charged materials with suitable allowance for loss.

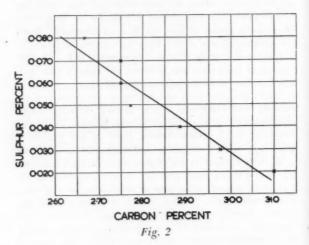
In the production cupola, no attempt is made to dephosphorize and under the operating conditions which favor desulphurization, no loss of phosphorus is obtained or anticipated. Manganese loss is calculated at 20 per cent.

Carbon Control

The question of carbon pickup will be dealt with in detail later. At this point it is sufficient to state that slag basicity and quantity of coke are the two main factors which determine the extent of the pickup so that an increase in the proportion of basic flux and higher coke quantity in a given charge will tend to increase the carbon absorption.

In the unit described above, high carbon pickup is not desired. The carbon content of the cupola metal is preferably held below 3.0 per cent so that the balance of steel scrap to be added to the electric furnace bath, to give a final figure of 1.40 per cent is within the melting capacity of the unit. In this practice, therefore, the coke quantity is standardized at 11 to 12 per cent and the combined fluxes, limestone and fluorspar at 8 to 9 per cent, giving an average figure of 2.8 per cent carbon at the cupola spout. This represents a pickup in carbon of 1.4 per cent from the charged carbon of 1.30 to 1.40 per cent.

It has been established in basic melting that carbon pickup bears a distinct relationship to sulphur reduction. In other words, conditions which favor carbon absorption intensify the desulphurization reaction as clearly shown in Fig. 2. The figures plotted



on this graph are averages taken from a large number of production melts under the standardized conditions described above. The coke quality without doubt has an important bearing on carbon pickup and for reference purposes, the properties of the coke used for this work are given in Table 3.

TABLE 3-COKE PROPERTIES

Fixed Carbon, %	87.5
Ash, %	11.0
Sulphur, %	1.0
Volatile, %	0.5
Shatter test	82% on 2 in. screen
	90% on 11/2 in. screen

Possibilities of Further Application

It will be realized that the use of a basic cupola on the specialized unit described, represents the application of basic melting in, perhaps, its simplest form with a single main metallic constituent in the charge. It is of interest to examine whether advantage can be taken of the effect of melting metals under the inIES

ckux

nd

tal

al-

ce

in

ce.

to

nd

re

p-

he

on

re-

ır-

ic-

ed

ip

n am e. fluence of basic slags and with conditions which are present in the cupola-type furnace. For example, in the manufacture of nodular cast iron, low sulphur is critically important in the development of the nodular structure. In the production of these irons by magnesium treatment, sufficient magnesium alloy must be added to reduce the sulphur to about 0.015 per cent before the requisite amount of magnesium is available for solution in the iron to produce the desired effect.

Carter³ has pointed out the advantage of having low initial sulphur content prior to the magnesium treatment in order to avoid the use of excessive amounts of expensive alloy and an unpredictable recovery of the essential element. Loss of metal temperature which accompanies excessive ladle treatment seriously affects the casting properties of metal and in good practice should be reduced to a minimum. It is correct to assume, therefore, that the nodular treatment would be facilitated if the initial sulphur content of the metal at the cupola spout could be held at, say, 0.030 per cent and with basic melting, under certain conditions, this is quite possible. The effect of such conditions on the control of other elements must however be considered, in particular carbon the absorption of which is intensified by basic conditions. Since the reduction of sulphur to low limits is normally accompanied by high carbon pickup, let us consider the latter.

The problem of increasing carbon pickup in the cupola has recently created considerable interest as it is realized that greater proportions of scrap could be utilized in furnace charges, if carbon absorption could be intensified. Attempts have been made to raise the carbon during the melting by the inclusion in the fuel charge of such materials as petroleum coke, pitch coke, graphite and carbon electrodes. Levi4 has reported an increase of 0.25 per cent carbon in the metal at the spout in certain charges by the use of these materials either singly or in combination. The availability of these products is limited and their use is consequently restricted. Coke quality undoubtedly has some bearing on the rate at which carbon can be absorbed and the pickup is reduced with coke of higher ash content.

Higher melting zone temperature has the effect of raising the carbon content and consideration must be given to the use of hot blast for this purpose. In the author's experience, operating with a blast temperature of 650 F and using the Griffin system for this purpose, it has been found possible to raise the carbon by 0.07 per cent in a cylinder iron mixture, thus

enabling the replacement of pig iron by steel scrap in the charge to the extent of 4 per cent.

The effect of slag composition and of the nature of the slag on carbon pickup has not been thoroughly investigated, possibly because in acid practice little opportunity is offered for the modification of slag properties. Fluorspar is frequently used for increasing the fluidity of acid slags and the author has found that in producing more active and fluid slags in the acid-lined cupola, carbon pickup is increased. For example, the carbon content of a cylinder iron has been raised 0.10 per cent by replacing 10 per cent of the limestone charge with fluorspar thus enabling pig iron to be replaced with steel scrap to the extent of 6 per cent.

Work carried out in Germany and first reported by Heikin⁵ revealed the remarkable effect of basic slag operation in accelerating carbon pickup when melting cupola charges consisting of 40/60 per cent hematite pig iron and 60/40 per cent steel scrap. In a cupola lined with dolomite, carbon contents in the region of 3.80 to 4.30 per cent were obtained whereas in the acid cupola, similar mixtures gave 2.85 per cent carbon. Confirmation of this has since been reported by the Institute of British Foundrymen. In the latter work, a stabilized dolomite lining was used and all-steel melts gave carbon figures of 3.0 to 3.5 per cent, being up to 1.0 per cent higher than in the acidlined cupola. The same effect was noted by the author when, during the war, a basic-lined cupola was used on a cupola/converter plant for steel casting production and Table 4 gives details of the charges used with typical analysis obtained.

Table 4-Data on Basic Cupola used in Cupola/Converter Practice

			Lb			Lb
Pig Iro	n	8	100	Coke		270
Steel So		12	200	Spar		10
	ilicon (4	5%)	25	Limestone		130
		LY ANALYSI	is		CALCULA	TED
T.C.	Si	S	P	Mn	Si	T.C.
3.44%	1.24%	0.038%	0.050%	0.50%	1.50%	1.78%
3.33	1.0	0.039				
3.15	1.10	0.042				
3.35	1.13	0.040				
3.52	1.18	0.033				
3.34	1.00	0.042				

In acid practice, it is recognized that low-carbon charges will absorb a greater amount of carbon than high-carbon charges, depending to a certain extent on

TABLE 5-EXPERIMENTAL MELTS TO SHOW THE EFFECT OF BASIC MELTING ON CARBON PICKUP

	M	etal Cha	arge, lb			Flux,	lb			Cha	rged	Analysi	s. %		Actual	Analy	sis, %	C	arbon, 9
Melt	Pig Iron,	Steel Scrap,	Iron Scrap,	Ferro Sili- con (75%)	Coke	Lime- stone		TC	Si	S	Mn	P	TC	Sî	S	Mn	P	Loss	Pick up
A	1200				170	110	20	3.75	3.50	0.04	1.00	0.040	3.50	2.75	0.023	.90	0.040	0.025	
В			1200	- 8	135	120	10	3.35	2.30	.12	0.70	.130	3.56	2.08	.046	.62	.125		0.21
C		750	250	30	180	120	10	1.02	3.00	.07	0.70	.075	3.61	2.17	.030	.56	.075		2.59
D		1000		40	180	120	20	0.25	3.00	.05	0.60	.050	3.10	2.35	.033	.50	.040		2.85

the sinicon and phosphorus present and the nearness to the eutectic composition. High-carbon charges tend to lose carbon in melting.

A similar effect is observed in melting under basic conditions but the gain in carbon in low-carbon charges is greater by a considerable degree. Table 5 gives the results obtained in a series of experimental melts made in the basic cupola in order to observe the effect of melting metal charges of varying charged carbon content. Melt (A) consisting of an all-pig iron charge showed a carbon loss from 3.75 per cent to 3.50 per cent with a sulphur reduction from 0.044 per cent to 0.023 per cent. Melt (B), an all-cast iron scrap charge approximating eutectic composition, showed a slight gain in carbon from 3.35 per cent to 3.56 per cent and a sulphur reduction from 0.120 per cent to 0.046 per cent. Melt (C) consisting of 75 per cent steel scrap and 25 per cent cast iron scrap gave a gain in carbon from 1.0 per cent to an average of 3.61 per cent, the sulphur being reduced from 0.070 per cent charged to 0.030 per cent. Melt (D), an all-steel scrap charge with silicon added to approximate the previous charges, showed a carbon pickup from 0.25 per cent charged to 3.10 per cent, the average sulphur being 0.033 per cent. In Melts (C) and (D) the coke percentage was increased to allow for the carbon ab-

sorption and, no doubt, facilitated both carbon pick-

up and desulphurization by producing the necessary

reducing conditions. The limestone charge was also

increased in order to raise the slag basicity and the

fluorspar to assist the slag fluidity.

It may be inferred from these results that when operating the basic cupola with high flux charges capable of reducing sulphur to a low level, cognizance must be taken of the high carbon pickup, particularly in low-carbon charges. Metal of normal gray iron composition can be produced from high-steel charges and, at the same time, the sulphur content can be reduced to a low figure. The question of control of the reactions which produce the carburization, desulphurization and silicon oxidation must be considered, realizing that such reactions are not likely to proceed to equilibrium in the cupola furnace. For this reason greater fluctuations in carbon and silicon than normal could be anticipated, particularly in high-steel charges with the silicon concentrated in a small proportion of the charge and a receiving unit of adequate capacity would be required to smooth out variations.

Alternatively, when it is not required to take fullest advantage of the carburizing reaction, it would appear necessary to use charged materials with a composition near to eutectic value in which case, silicon control would be simplified, this element being distributed more evenly throughout the components of the charge. When the price of steel scrap is favorable, high steel charges would be a more attractive proposition, particularly for large scale production and when the phosphorus needs to be considered.

Such results cannot be obtained in the acid-lined cupola and it is evident from the foregoing that cupola charges may be computed for basic melting which would produce a low-sulphur iron of a base composition suitable for the nodular treatment.

Refractory Technique

Good refractory technique is essential in any type of melting furnace and the loss of refractory in cupola melting is undoubtedly high. This loss, occurring mainly in the melting zone, is due to the high temperature prevailing in this region and to attack by slag rich in iron oxide. Basic refractories, in general, have a higher fusion point than the acid refractories normally used for cupola linings but the presence of

high iron oxide slag must be considered.

The most economical basic material available is calcined dolomite which may be prepared in ramming form by mixing with dehydrated tar. The time available for daily cupola repair is normally limited and would seldom allow the re-ramming and sintering of dolomite to be thoroughly carried out. While the fusion point of this material is high, it is subject to attack by iron oxide slags and Heiken reporting on its use, found it necessary to re-ram the cupola after each melt of 40 tons of metal. Carter³ has reported the use of magnesite bricks for the initial lining construction and a ramming mixture of 90 per cent magnesite and 10 per cent bentonite for patching purposes. Lining erosion in this case was stated to be less than would be expected from acid refractory, the indication being that the magnesite lining could be operated two to three times as long as an acid lining. The resistance of magnesite to slags rich in iron oxide is known to be high and this may account for the improved results.

Stabilized dolomite consists essentially of tricalcium silicate and magnesia, the manufacture of which has been described elsewhere. The thermal shock resistance of this material in brick form is less than magnesite, as also in its resistance to iron oxide slag attack, and for these reasons its performance could not be expected to equal that of magnesite. However, it has been used consistently by the author because of its availability in England as a home produced material. The consumption of refractory on the production unit is as follows:—

Patching material—42 lb per short ton of metal melted.

Basic brick —13.6 lb per short ton of metal melted.

This consumption is comparable to that of acid refractory with a similar metallic charge and for a cupola of 42 in. diameter, but it should be noted that the coke percentage on this unit is 11 to 12 per cent and the melting rate is within the rated capacity. When melting with higher coke quantities, lining life could be expected to be shorter and consumption of refractory greater than that normally obtained in acid practice.

As cost of basic refractories is several times greater than acid refractories, their use can only be justified when the additional cost is offset in other directions, as in the production unit described above. In contemplating the use of a basic lining for the manufacture of nodular iron, therefore, the higher refractory cost must be balanced against the cost of the alloy required to desulphurize to the necessary low limit of

TES

ola

ng

ıp-

by

al,

ies

of

is

ng

il-

nd

of

he

to

on

er

ed

n-

ıg-

ır-

be

he

be

E.

de

he

m

as

st-

g.

it-

ot

of

IC-

e-

a

at

nt

fe

of

d

21

d

n-

C-

y

sulphur. The evidence is that a case is presented where the use of a basic lining is worth consideration and in reducing ladle processing to a minimum, the final alloy treatment will be under a greater degree of control.

Water Cooling

The heavy erosion of cupola linings, whether acid or basic presents a constant problem in cupola melting and recent work on the water cooling of critical portions of the lining appears to offer a solution. It has been shown that replacing refractories in the melting zone by water jackets and lining the cupola well only with basic refractory, it is possible to operate with a basic slag.⁸ The erosion products from a refractory wall constitute an appreciable proportion of the slag bulk formed in the cupola and when erosion is prevented by water cooling, larger quantities of basic flux can be used to increase the slag basicity to the extent of that obtained in the basic-lined furnace.

An interesting application of this idea is evidenced in the so-called 'metallurgical blast cupola' developed and described by Doat.⁹ Although not described as such, this furnace is essentially a water-cooled, hot-blast cupola. Whereas the development of the hot-blast cupola has had for its main intention economy in fuel, the 'metallurgical blast cupola' is operated with normal or higher than normal coke quantities, producing a top gas richer in CO and of higher calorific value, so that only a portion (35 per cent) of this gas is required for pre-heating the incoming air. The remainder of the gas, it is stated, is available for heating purposes external to the furnace, as in blast furnace practice.

Recuperation is such that the blast temperature is in the region of 1000 F and by employing water cooled tuyeres, protruding some distance into the furnace, combustion appears to be centralized. In this manner the melting zone itself is restricted in area and is reduced in height. This effect, combined with water cooling of the refractory for some distance above the tuyeres, reduces the lining erosion to negligible proportions and the slag suffers no dilution from this source.

Under these conditions, it would be expected that any basic flux in excess of that required to absorb the coke ash and oxidation products, would be available to increase the slag basicity, the effect on metal composition being similar to that obtained in the basic-lined cupola. The results quoted by Doat appear to confirm this as it claimed that in melting a charge of 100 per cent steel scrap, carbon can be absorbed up to 4.0 per cent and sulphur reduced to 0.050 per cent. One may speculate as to how much the increased carbon pickup and sulphur reduction may be attributed to the higher temperature blast but as yet, there is no evidence that this factor alone is capable of producing such results and the inference is that slag reactions play a most important part.

Up to the time of preparing this paper, no intimate details of operation of this type of furnace have been published but the principles are undoubtedly sound and it is in this direction that further work could usefully proceed.

Summary and Further Comments

It will be appreciated that the basic cupola should be considered as a special purpose furnace. In normal gray iron production sulphur does not present a problem when coke of reasonable sulphur content is available. When high steel scrap charges are to be melted in the acid cupola, sulphur pickup is prone to be high and normally the soda ash ladle treatment will suffice for desulphurization. The use of a basic-lined cupola is not necessarily economically competitive with the soda ash treatment but the latter has certain disadvantages and this paper gives an example where the basic cupola has been used to advantage.

Because of current interest in the subject, particular reference has been made to the possible use of the basic cupola in the manufacture of nodular iron. Operating factors which determine desulphurization to the low limits required for nodular treatment have been discussed and it is clear that desulphurizing slags can be worked in the cupola, such slags being sufficiently basic and low in iron oxide content. The phenomenon of high carbon pickup obtained under these operating conditions confirms previous observations and is a factor which must be taken into account, particularly where low sulphur metal is the objective. Providing that irons melted in this manner prove to be consistently responsive to the nodulizing treatment, it is suggested that the added cost of basic melting will be offset by economy in the amount of alloy required for the treatment.

Development in the technique of melting in the cupola furnace has been relatively slow, much of the published literature being devoted to the theory and practice of combustion and principles of routine control. As is frequently the case, technical development is created by necessity and in recent years, with changes in the raw material situation and the introduction of new irons, it has been clearly shown that some advantage would be gained if the versatility of the cupola could be increased. Recent work on basic melting, water cooling and the use of high temperature hot blast has indicated a new approach to cupola development, and there is ample evidence available that their possibilities should be thoroughly explored.

Acknowledgment

The author acknowledges his indebtedness to the Ford Motor Co., Ltd., Dagenham, England for permission to publish this work and to his foundry colleagues for their invaluable support.

References

- E. S. Renshaw, "Basic Cupola Process for Desulphurisation," Foundry Trade Journal, vol. LXX, June 24/43.
 E. S. Renshaw and T. Foley, "Duplexing Low Carbon
- 2. E. S. Renshaw and T. Foley, "Duplexing Low Carbon Alloys," *The Iron Age*, May 22/47.
- 3. S. F. Carter, "Basic-Lined Cupola for Iron Melting," A.F.S. Transactions, vol. 58, pp. 376-392 (1950).
- 4. W. Levi, "Operation of the Cupola," A.F.S. Transactions, vol. 58, pp. 1-19 (1950).
- Heiken, "Production Results with a Basic-Lined Cupola," Foundry Trade Journal, vol. 72, Mar. 2/44.
- 6. Second Report on the Basic Cupola, Proc. Inst. of British Foundrymen, 1945/6.
- 7. T. Swinden and J. H. Chesters, "Dolomite Bricks for use in Steel Works," Proc. Iron and Steel Inst (1941).

8. E. S. Renshaw and S. J. Sargood, "Some Modification in Cupola Design," Foundry Trade Journal, Oct. 13/49.
9. R. Doat, "The Metallurgical Blast Cupola," Foundry Trade

 R. Doat, "The Metallurgical Blast Cupola," Foundry Trade Journal, vol. 89, no. 1774, Aug. 31/50.

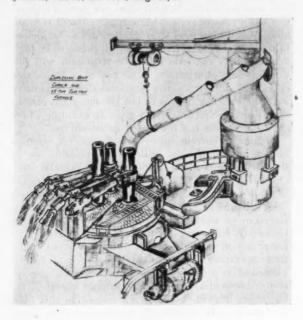


Fig. 3

DISCUSSION

Chairman: MAX KUNIANSKY, Lynchburg Foundry Co., Lynchburg, Va.

Co-Chairman: GOSTA VENNERHOLM, Ford Motor Co., Dearborn, Mich.

R. V. RILEY (Written Discussion): ¹ Mr. E. S. Renshaw's valuable paper draws attention to the relationship between the sulphur and the carbon content of cupola melted iron. He has shown that when the carbon is high, the sulphur is low, and vice versa. In cupola melting this relationship is now regarded as almost axiomatic.

It must be understood, however, that the relationship is possibly merely coincidental and that the mechanism of carbon absorption is probably not governed by the laws which relate to sulphur removal. Renshaw's paper shows the importance of slag reactions in the control of sulphur. These slag reactions probably do not influence the carbon. In a paper presented to the 1951 Annual Conference of the Institute of British Foundrymen* it has been shown that carbon solubility in molten iron in the cupola is an example of simple solution of a solid in a liquid.

The rate of solution depends upon:-

1. The temperature of the molten iron.

2. Concentration of the carbon already present in the iron.

3. Intimacy of mixing of the carbon with the molten iron.

4. Nature of the carbonaceous material.

The presence of agents which may interfere with or assist carbon solution rates.

The effect of the slag upon the carbon absorption in cupola melting is, I consider, indirect, insofar as its action is simply to remove from the surface of the furnace coke, the ash it would otherwise accumulate. The accumulation of ash on the coke surface certainly prevents carbon solution and any manipulation of the physical or chemical properties of the cupola slag so as

to secure the more rapid removal of the ash, will have the effect of increasing the carbon pick-up of a low-carbon charge. This is, I think, the explanation of the action of fluorspar in an acidlined cupola when a slightly greater pick-up of carbon occurs.

MR. RENSHAW (Reply to Dr. Riley): Dr. Riley's contribution to the discussion is particularly welcome in view of his very comprehensive investigation and findings on the solubility of carbon in molten iron which in themselves have been verified in

practice by the author.

In reporting the relationship between carbon pick-up and sulphur reduction, it was intended to show that the conditions created in the furnace to promote desulphurization directly effected carbon absorption. In other words, slags of higher basicity may react more rapidly with the acidic oxides of the coke ash leaving the incandescent carbon more exposed to the molten iron. This effect has already been referred to by the author in a previous paper.

There is also evidence that a higher melting zone temperature is obtained in basic melting due possibly to the more rapid disposal or fluxing of the coke ash, a condition which may in

itself promote carbon absorption.

Furthermore, under these conditions the incandescent coke may prove to be more actively reducing, the effect being to reduce the iron oxides which could react with the carbon in the metal.

The effect of slag, therefore, on carbon absorption is, perhaps, indirect as suggested by Dr. Riley and it is not inferred that because sulphur is reduced the solubility of carbon in the molten iron is increased.

At this stage of the enquiry, however, it is probably wise to exercise some restraint in drawing conclusions. Levi has reported, for example, an instance where an unusually low carbon pick-up was obtained when melting a high-sulphur iron.

S. F. CARTER: 2 I would like to compliment Mr. Renshaw for his excellent paper and fine presentation. I would like to compliment him further for the alert and progressive contributions he has made over the last several years in the field of cupola

design and cupola operation.

We have experienced a similar relationship between carbon pick-up, sulphur elimination and slag basicity. We have had most of our experience in the higher and lower sulphur ranges because of nodular iron experimentation. As the slag basicity increases the sulphur content of the iron decreases and the carbon content increases. I think there is both a direct and an indirect cause. Due to differences possibly in coke and raw materials, our data might not fall perfectly on Mr. Renshaw's curve but certainly they would be parallel to it. Before we can properly control the basic cupola we need to understand this relationship better.

the bh

thol

I would like to ask two questions.

1. Whereas Mr. Renshaw's experience has been mainly with lower carbon irons, our experience has been on higher carbon irons where carbon pick-up has been encouraged. Has Mr. Renshaw found basic operations to have the same control possibilities as the acid cupola from the standpoint of carbon consist-

ency using comparable charge materials?

2. Has Mr. Renshaw found the basic cupola more efficient from a thermal standpoint? We have seen some evidence that the basic cupola melts faster, will melt with a lower proportion of coke at the same temperature or will melt hotter with the same ratio of coke. Frequently when the two were compared, variations in charging materials added too many variables for a conclusive comparison. Has Mr. Renshaw any production experiences that would indicate that the basic cupola is more efficient from the coke consumption or temperature-production standpoint?

Mr. Renshaw: Mr. Carter's remarks are of particular value in view of the excellent paper he produced in 1950 on the same subject.* The question of control is of considerable importance and presents more difficulties than in acid melting, undoubtedly because of the reactions purposely created. This particularly applies to silicon. In the acid cupola operating to give silicon

^{9 &}quot;Factors Affecting the Solubility of Carbon in Iron," by R. V. Riley. Paper No. 997, Institute of British Foundrymen's Annual Meeting, 1951.

Research Manager, The Stavely Iron & Chemical Company Limited, near Chesterfield, England.

² Assistant Melting Superintendent, American Cast Iron Pipe Co., Birmingham, Ala.

^{*} S. F. Carter, "Basic-Lined Cupola," A.F.S. Transactions, vol. 58, pp. 376-392 (1950).

IES

eci

his

id

ion

erv

of

in

nd

ons

tly

er

he

he

he

oid

in

ke

to

in

DS.

he

to

6-

on

n

ns

la

on

ad

es

tv

he

ın

34

r's

ve

id

th

m

n

i-

t-

nt

m

16

d,

or

m

re

11

ce

control to 0.20 per cent, a similar change in the basic cupola would, in our experience, give a silicon fluctuation within 0.40 per cent. When operating with slags of high basicity in order to intensify carbon pick-up and sulphur reduction, a silicon loss of a higher order may be anticipated and control factors will require investigation. Variations may, of course, be smoothed out by using a receiving unit of suitable capacity.

With regards to thermal considerations, evidence is available to indicate that higher tapping temperatures are obtainable in basic melting for a given quantity of coke. For example, in the production cupola described, a lean coke ratio is employed in order to limit carbon pick-up. The coke used is a blast furnace grade of small size with 12 per cent ash content. Even so, with a 9:1 coke ratio tapping temperatures of 2700 to 2750 F are regularly obtained, whereas, with the same proportion of fuel used in acid melting, we should not expect to exceed a temperature of 2650 F.

Assuming that the basic slag will attack the acid coke ash with greater rapidity, this effect may result in a more rapid combustion of the coke and account for higher operating tem-

E. A. PIPER: ^a I am very much interested in the paper presented by Mr. Renshaw. As far as our own experience in operating a basic cupola is concerned, it is in close agreement with the statements made in the paper. We are melting at higher temperatures than in the previous acid practice, and are saving about 5 to 10 per cent on coke. Since we are producing a variety of jobbing castings to engineering specifications, we are not interested in very low sulphurs. We hold sulphur from 0.08 to 0.09 per cent and therefore, operate with a slag of a relatively low basicity.

Refractory cost per ton of metal is definitely higher with the basic cupola. The savings in coke and in labor for the daily patching operation are offsetting factors.

In normal times, when pig iron is considerably higher in price than scrap, an overall saving in melting high scrap mixtures in the basic cupola may be realized. Under the present local metal market conditions, the operation of a basic cupola can not be justified financially.

W. W. Levi: A I would like to compliment Mr. Renshaw on the excellent presentation he has made. It was his preliminary work which actually stimulated our interest in the subject of basic cupola operation.

Mr. Kuniansky and I visited in Mr. Renshaw's plant in February, 1950, and upon our return to this country decided that we would install a basic-lined cupola. Our basic-lined cupola is water cooled and is provided with hot blast. We have been operating our basic cupola since February 15, 1951 and have had some good results. As expected, some difficulties have been encountered.

We are trying to operate 12 hr per day and have succeeded in doing this for the last few heats. Our charge consists of 35 per cent steel, 20 per cent pig iron, 40 per cent cast scrap, and 5 per cent silvery. With this charge we are using a good grade of beehive coke containing 92 to 94 per cent fixed carbon and very low percentages of SiO_2 and Al_2O_3 .

Our coke ratio is 6.7 to 1, which is considerably more than Mr. Renshaw is using. We are operating our air blast heater at 850 F and under these conditions we are able to produce iron at the cupola spout containing between 0.025 and 0.035 sulphur with a total carbon content of about 3.8 per cent. We are auxious to obtain high carbon and low sulphur, which is quite different from what Mr. Renshaw is trying to do.

There is one question I would like to ask Mr. Renshaw. I was interested in his comment on the depth of slag that we would like to maintain in the well. His comment, I believe, was that if this depth was not maintained to the proper height, the desulphurizing would be considerably less than would be obtained with a deeper layer of slag. I would like to ask Mr. Renshaw what he considers the optimum depth of the slag and if he has any way of determining this depth.

Mr. Renshaw: It is very interesting to learn of Mr. Levi's experience with his work on the basic cupola. We have found that the depth of slag in the furnace well has a distinct bearing on the desulphurizing reaction. In the continuous tapping system described, the slag height in the cupola well is determined by the metal level at the front hole or metal exit. It is necessary in determining the height of the metal level to consider the fluid pressure of the slag depth in the well. With a depth of slag of about 2 ft, we find that a metal level of 7 in. to 8 in. is quite satisfactory and some indication of the slag height may be obtained by observation of slag erosion line after a melt. Increasing the depth of the well to carry a larger volume of slag would, no doubt, give a greater carbon pick-up but would involve some difficulty in initial heating. Carter's valuable work on the use of calcium carbide as an addition to the cupola bed has shown it possible to obtain high initial tapping temperatures. This may have some application when it is desirable to operate with a deep well.

MEMBER: Is there any difficulty in maintaining the diameter of the tap hole in a front-slagging cupola? If so, what are the best refractories to use for this application? What happens when the tapping hole becomes too large? What are some of the difficulties that one might look for in consideration of the basic front-slagging cupola operation?

MR. RENSHAW: Tapping hole erosion does present a problem, particularly for long operations. The cupola described is operated for 10-hr periods and using the dolomite monolith a tapping hole enlargement from 1½ in. to 4 in. is obtained. Somewhat less erosion has been found by using a mixture of ganister with 30 per cent graphite. The effect of an enlarged tapping hole can be offset by raising the metal level thus maintaining the depth of the slag in the well.

W. W. Austin: ⁵ I realize that the conditions for sulphur removal in a basic cupola are not generally compatable for conditions of phosphorus removal. However I would like to ask Mr. Renshaw if he considers there is any hope for dephosphorization in the basic cupola assuming a low-silicon charge?

Mr. Renshaw: Some attempt was made to dephosphorize in the basic cupola, when during the war, there was a danger of a shortage of low-phosphorus ores. A series of experimental melts were reported by the Institute of British Foundrymen and it was found essential to operate with a low silicon content in order to obtain any degree of dephosphorization. When operating such charges, certain difficulties arose in the manipulation of the slag which became viscous and froze around the tuyeres. This caused skulling and resulted in high metallic losses up to 30 or 40 per cent. Considerable work will be required before dephosphorization could be made a proposition.

MEMBER: We recently began basic cupola operation and encountered two difficulties, namely, viscous slags and excessive erosion of the refractories. Has Mr. Renshaw found an effective fluidizer for the slag and how has he reconciled that fluidizer with refractory wear, particularly ladle wear?

Mr. Renshaw: Viscosity of the slag increases with the lime content or basicity ratio and calcium fluoride assists in maintaining fluidity. It is preferable to use fluorspar in lump form to avoid mechanical losses in the stack and we have found that using a 2-in. to 3-in. grade of spar it is possible to maintain a calcium fluoride content in the slag of 2 to 3 per cent, thus assisting fluidity. To assist the flow of slag over the slag notch, a layer of charcoal may be placed on the slag surface and a gas flame directed on to the surface is a further help. There is no evidence that calcium fluoride slags give excessive erosion of the refractory.

When experimenting with slags of high basicity, i.e., with a lime content of 52 to 55 per cent, on cooling, the slag will tend to "fall" being evidence of the presence of dicalcium silicate.

^a Pohlman Foundry Co., Inc., Buffalo.

⁴ Metallurgist, Lynchburg Foundry Co., Redford, Va.

⁵ Senior Metallurgist, Southern Research Institute, Birmingham, Ala.

MODIFICATION OF ALUMINUM - SILICON ALLOYS

By R. H. Dyke*

Introduction

This paper gives the results of a comparison of the various modification techniques in use for aluminumsilicon alloys.

The effectiveness of several rapid tests which could be used by foundrymen to determine whether or not a melt had been successfully modified was also investigated.

Historical

Aluminum-silicon alloys were the first alloys of aluminum to be produced; a published account of their properties appearing in 1874.¹ However, the later development of other aluminum-base alloys having better mechanical properties resulted in the comparative neglect of the aluminum-silicon alloys until about 1923. Early work had shown that the addition of silicon to aluminum gave a progressive increase in tensile strength, accompanied by a reduction in elongation. The maximum tensile strength in a sand-cast alloy was found to occur with a silicon content of 10 to 12 per cent,^{3, 4, 5} the mechanical properties being:

Ultimate Tensile Strength 9-10 ton per sq in. Elongation 2 per cent

Up to 1923, the cast alloys were characterised by poor machinability and a coarse fracture attributed to the presence of massive plates of silicon.

In 1921, Pacz, in a patent⁶ claimed that the mechanical properties of the aluminum-silicon alloys, principally those containing between 8 and 13 per cent silicon, could be materially improved by treatment of the molten metal before casting with alkali fluorides, the most effective being sodium fluoride. This process was termed modification. This patent was followed by several others covering the modification of these alloys by the use of other agents including alkali metals,⁷ alkaline earth metals,⁸ antimony and bismuth⁹ and alkaline oxides.¹⁰

Of these modifying agents, the two most widely used have been sodium metal and the alkali fluorides. Modification using alkali fluorides is more popular in Europe, whilst the use of sodium metal, generally contained in a thin aluminum capsule, is preferred in America and Great Britain.

The exact mechanism of modification has been studied by many workers and several facts regarding the phenomenon have been established. These may be summarized as follows:

1. In an unmodified alloy containing 12 per cent or more of silicon, the silicon occurs as a primary phase in the form of coarse needles and plates. In a modified alloy the silicon occurs as a fine secondary phase of globular types. The transition from unmodified to modified structure is gradual, and one sample may contain both modified and unmodified areas.⁵

2. The process of modification is accompanied by an increase in both tensile strength and elongation.⁵ The finer state of distribution of the silicon results in improved machinability.

3. Remelting of a modified alloy, or holding the molten alloy for prolonged periods in the molten state results in a reversion to the unmodified condition.

4. The silicon content of the eutectic apparently increases from 11.7 per cent to 14 per cent during modification; the temperature of the eutectic arrest on cooling is also lowered, 12 the amount of lowering increasing with the cooling rate. Some work has shown a eutectic arrest as low as 564 C. 11. 13

5. The addition of excessive amounts of sodium to the alloys can produce "over modification." In this condition, the alloy becomes susceptible to gas porosity. 12, 14, 15 It appears that the aluminum-silicon alloys become more susceptible to gas porosity the higher the sodium content and alloys with a sodium content only sufficient to give modification will absorb hydrogen more rapidly than alloys in the unmodified condition.

Investigation of Modification Techniques

Experimental Detail—A comparison was made of three techniques commonly used to modify aluminumsilicon alloys containing approximately 12 per cent

^{*} Defence Research Laboratories, Department of Supply, Commonwealth of Australia, Victoria, Australia.

This is an Official Exchange Paper to the American Foundrymen's Society from the Australian Institute of Foundrymen.

ly

in

n.

in

d-

1e

e

nt

a

le

n

le

te

g

silicon using both flux and sodium metal methods. The modifying agents used were as follows:—

1. 66 per cent sodium chloride-34 per cent sodium fluoride flux.

2. A low melting point ternary flux containing 50 per cent sodium chloride, 35 per cent sodium fluoride and 15 per cent potassium chloride.

3. Sodium metal.

The aluminum-silicon alloy was prepared from commercial aluminum and silicon metals. The compositions of these materials and typical compositions of the final melts are shown in Table 1. The requirements of British Standard 2L33 are given as a basis for comparison.

TABLE 1-COMPOSITION OF ALLOYS

Sample	Element, %								
•	Al	Si	Fe	Mn	Zr	Others			
Aluminum-	99.7	0.1	0.2						
Silicon		98.	1.5	0.4		0.1			
BSS2L33	Rem	10-13	0.6Max	0.5Max	0.1Max	0.6Max			
AH1	Rem	12	0.23	**	0.01	0.26			
BL3	Rem	12	0.21	**		0.21			
SH2	Rem	12	0.20			0.25			

Charges of 10 lb each were melted in Salamander crucibles in a coal-gas fired furnace. After melting, any residual sodium was removed by adding hexachlorethane in successive amounts each of 0.5 per cent of the charge weight until the alloy was in the unmodified state.

Samples were taken in a small hand ladle and cast in green sand molds into bars approximately 1/2 in. in diameter and 61/2 in. long. One sample was taken before the addition of the modifying agent and subsequent samples 5, 10, 15 and 20 min after the addition. The temperature of the melt was measured by an alundum covered thermocouple and was held at the two selected modification temperatures, 700 ± 20 C and 800 ± 20 C. The melts were stirred with a graphite rod before each sample was taken.



Fig. 1-Stage 1, unmodified alloy. Mag. 200x.

The experimental results were expressed in terms of five arbitrary states of modification based on the microstructures of sections taken ½ in. below the top of each bar. The microstructures typical of these states were:

Stage 1. The unmodified alloy containing plates and coarse needles of silicon randomly distributed in a matrix of aluminum-rich solid solution (Fig. 1).

Stage 2. Coarse needles of silicon between dendrites of aluminum-rich solid solution (Fig. 2).

Stage 3. Coarse rounded particles of silicon between dendrites of aluminum-rich solid solution (Fig. 3).

Stage 4. A satisfactorily modified structure with fine particles of eutectic silicon and rounded dendrites of aluminum-rich solid solution (Fig. 4).

Stage 5. An over-modified structure with coarse particles of silicon and bands of aluminum-rich solid solution containing needles of a ternary aluminum-silicon-sodium phase (Fig. 5). A more detailed discussion of these structures is given later in the paper.

Modification by Sodium Chloride-Fluoride Flux

The original patent application of Pacz⁶ covered the use of sodium and potassium fluorides as modifying agents. Fluxes of this type generally consist of mixtures of sodium fluoride and chloride. Typical modification techniques using these fluxes, as given in the literature, are shown in Table 2.

TABLE 2-BINARY FLUX LITERATURE REFERENCES

Reference	Comp	osition	Amount	Temperature	Time
	NaCl	NaF	1% of Charge	C	Min
Polyak ¹⁵	2	1	3	760-800	
Baikov16	2	1	1-3	780-800	5-20
Botchvar ¹⁷	2	1		740-790	
Duport ¹⁹	2	1	1.5	900	5-10
Welter ¹⁸	1	2	0.4	700	
Archer ¹¹	1	2	3	927	

A study of the above references suggested that the

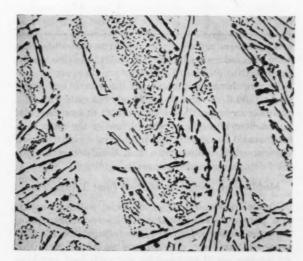


Fig. 2-Stage 2, large silicon needles. Mag. 200x.

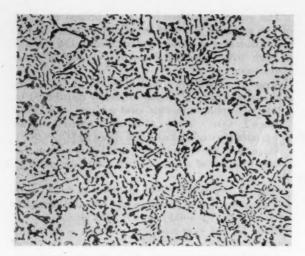


Fig. 3-Stage 3, fine silicon needles. Mag. 200x.

most satisfactory flux would be a 2:1 sodium chloridesodium fluoride mixture; an amount equal to 3 per cent of the charge weight being used at 800 C. The references gave no conclusive information on the time of treatment for satisfactory modification.

A flux containing 66 per cent sodium chloride and 34 per cent sodium fluoride was used, the amounts of flux used being 1, 2, 3 and 5 per cent of the charge weight. The stage of modification of the test bars, as assessed by metallographic examination, are given in Table 3:

TABLE 3-BINARY FLUX MODIFICATION RESULTS

Modification	Amount	State	of modif	ication af	ter
Temperature,	of flux, %	0 Min	5 Min	10 Min	15 Min
C					
700	1	1	1	2	3-4
	2	1	1-2	2-3	2
	3	1	2-3	3-4	4
	5	1	3-4	4	4
800	1	1	3-4	3-4	3
	2	1	3	3-4	4
	3	1	3-4	4	4
	5	1	4	4	4

Techniques which gave satisfactory modification were checked using fresh 10-lb melts, and the results confirmed to those in Table 3.

During the experiments with this flux it was found that at 700 C the flux did not melt but only formed a granular layer on the metal. At 800 C, however, the flux melted and covered the surface of the melt. German work²¹ has shown that modifying fluxes are most efficient when molten; this flux should therefore be used at a temperature approaching 800 C.

Modification using Low Melting Point Ternary Flux

Fluxes based on sodium chloride, sodium fluoride and potassium chloride have a lower melting point than the binary sodium chloride-sodium fluoride mixtures and their use allows lower metal temperatures, decreasing gas absorption by the melt and reducing fuel consumption and wear of furnace refractories.

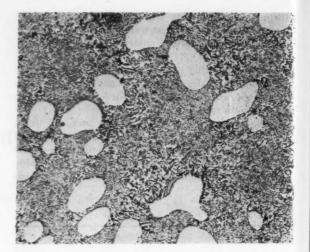


Fig. 4-Stage 4, fully modified. Mag. 200x.

Welter¹⁸ obtained modification using a flux containing 22 per cent sodium fluoride, 15 per cent sodium chloride, 44 per cent potassium fluoride and 19 per cent potassium chloride, 2 per cent of the charge weight being used at 900 C. A report on German work on the development of a ternary modifying flux21 described a flux containing 12 per cent sodium fluoride, 46 per cent potassium chloride and 42 per cent sodium chloride which was stated to melt at 650 C and was claimed to be satisfactory for modification at 720 to 760 C. A Belgian patent covering the use of a flux similar to this was taken out in 1943.20 The flux addition was 1 to 1.4 per cent of the charge weight, the modifying temperature being 650 to 800 C. No definite time of treatment was quoted. Duport and Caminade19 described the use of this flux: the conditions used being 2 to 4 per cent of the charge weight, 680 to 820 C modification temperature and 5 to 10 minutes treatment time.

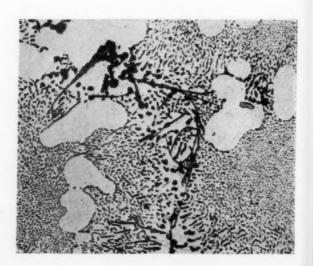


Fig. 5-Overmodified. Mag. 200x.

m-

di-

19

ge

an

ng

m

er

50

110

se

he

ge

00

rt

ne

ge

The flux selected for investigation contained 50 per cent sodium chloride, 35 per cent sodium fluoride and 15 per cent potassium chloride, its melting point being approximately 650 C. The amount of flux used was 1, 2, 3 and 5 per cent of the charge weight. The stage of modification of the test bars as assessed by metallographic examination are given in Table 4.

TABLE 4-TERNARY FLUX MODIFICATION RESULTS

Modification	Amount	De	gree of	Modifie	cation a	fter
Temperature, C	of Flux; %	0 Min	5 Min	10 Min	15 Min	20 Min
700	1	1	1	1-2	2	
	2	1	1	1-2	2	
	3	1	1	1-2	2-3	
	5	1	1-2	2	3	4
800	1	1-2	2	2	2	
	2	- 1	2	3	3	4
	3	1	2-3	3	4	
	5	1	3	4	4	

Techniques which gave satisfactorily modified structures were repeated on fresh 10-lb melts; the results confirming those given in Table 4.

Modification using Sodium Metal

The method of modification using sodium metal was first described by Archer.¹¹ The sodium was placed in a preheated ladle and the molten metal at approximately 770 C poured over it and then allowed to stand for 10 to 20 minutes before casting.

Later workers^{13,22} plunged the sodium metal, usually wrapped in aluminum foil, under the surface of the melt and held it there with an inverted cup type of plunger until dissolved. The amount of sodium used was about 0.06 per cent of the charge weight, the metal temperature was 700 to 750 C, and the holding time before casting varied between 5 and 20 minutes.

Additions of 0.02, 0.05 and 0.1 per cent of the charge weight of sodium wrapped in thin aluminum foil were made. The stage of modification of the test bars, as assessed by metallographic examination are given in Table 5.

TABLE 5-SODIUM METAL MODIFICATION RESULTS

Modification Temperature, C	Amount of Sodium,	Degree of		Modification After		
		0 Min	5 Min	10 Min	15 Min	20 Min
700	0.02	1	1	3-4	2-3	2
	0.05	1	2-3	3	2	1-2
	0.1	1	1	4	5	4
800	0.02	1	1-2	1	1	1
	0.05	1	5	4-5	4	4

The techniques which gave satisfactorily modified structures were repeated and the results confirmed those given in Table 5.

Conclusions

1. Metallographic examination showed that modification affected the structures in a progressive manner. The first change in the structure of an unmodified alloy is that the random distribution of the sili-

con particles changes to a dendritic form in which the silicon-aluminum eutectic is present between well defined dendrites of primary aluminum-rich solid solution (Fig. 2). This change is followed by a second phase in which the silicon particles decrease in size more rapidly with increasing degree of modification until, in a fully-modified alloy, the particles are too small to be completely resolved at a magnification of 200 (Fig. 4). An increase in the sodium content above that required for modification gives an "overmodified" structure in which the eutectic silicon particles become coarser and the aluminum-rich solid solution areas are banded, instead of being dendritic in shape. A new, dark gray aluminum-silicon-sodium phase appears as needles in the bands of aluminum-rich solid solution.

2. It was found that satisfactory modification could be obtained at temperatures in the range 700 C to 800 C with all of the three modifying techniques investigated. Assuming that for economic reasons 15 minutes is the longest practical modification time and that the amount of flux added and the modification temperature should both be as low as possible; the following flux modification techniques are considered to be the most satisfactory.

(a). Binary flux.—2 per cent of the charge weight of a 66 per cent sodium chloride—34 per cent sodium fluoride mixture allowed to remain on the metal for 15 minutes at 800 C.

(b). Ternary flux.—3 per cent of the charge weight of flux containing 50 per cent sodium chloride, 35 per cent sodium fluoride and 15 per cent potassium chloride allowed to remain on the metal for 15 minutes at 800 C.

(c). Sodium metal.—0.05 per cent of the charge weight of sodium metal plunged beneath the metal surface at 800 C and the metal allowed to stand for 10 minutes before casting.

3. The experimental work has shown that there is a difference in the progress of the modification between the flux and sodium metal techniques. With flux techniques the sodium content of the melt will increase progressively during the progress of modification whereas with the sodium metal technique the sodium content of the melt will reach a maximum when solution of the sodium metal is complete, and then, because of oxidation and volatilization, decrease during the holding period required for the sodium to diffuse throughout the melt. The sodium content of the melt is therefore always higher with the sodium metal technique than with the flux technique. As it is well established that the rate of absorption of hydrogen by the melt increases with sodium content,12,15 there is consequently more risk of excessive gas absorption when sodium metal is used than when modifying fluxes are used. Because of this and also because the modification occurs more gradually with fluxes, it is considered that the use of fluxes for modifying would give better results than the use of sodium metal.

4. Degassing with chlorine-containing agents results in a loss of sodium; it therefore being necessary to degas with these agents before modifying. This will also ensure that the residual sodium content of

R.

sei sta mo

po wa ou ure spe

are

to

un

ma

am

5-n

har

mo

sig

pro

bot

onl

sca

ed.

tio

un

san

han the I sev me me thicklish me

fou at :

fica me of

carrigati J. H

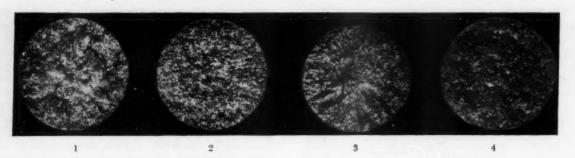


Fig. 6-Fracture test specimens. Mag. 2×.

the melt is low, enabling the foundryman to adopt a definite modification technique without the need to consider the effects of varying residual sodium contents of the melt. If degassing after modification is considered necessary, dry nitrogen should be used.

Investigation of Rapid Tests for Determination of Stage of Modification

Introduction—For the control of modification, it would be a great help to the foundryman if a rapid test to determine the degree of modification of a melt was available. To be of practical use, the test should require only equipment normally used in a light alloy foundry and the correlation between test results and stage of modification should be reliable.

For the purpose of this investigation, three tests conforming to the first of these requirements were studied to determine whether or not they gave a satisfactory indication of the degree of modification of the melt.

Fracture Tests—Fracture specimens were prepared by machining a small groove around each test bar, approximately 1 in. below the position of the metallographic sample and fracturing the bar. Typical fractures of bars having structures corresponding to the stages of modification described earlier are shown in Fig. 6.

It was found that a completely unmodified alloy has a rough fracture showing bright areas, the latter being due to the large plates and needles of silicon in the alloy. As modification proceeds and the size of the silicon particles decreases, the fracture becomes less rough and the size of the bright areas decreases. When the silicon particles become very small (stage 3), the fracture becomes dendritic in character and, in general, a small area in the center of the section has an even, grayish appearance characteristic of a fully modified alloy. With complete modification, the fracture is even and gray in color with some small bright areas.

In the case of a fine-grained alloy, difficulty was experienced in distinguishing between stages 2 and 4. The most satisfactory distinguishing feature was the presence in the fracture elongated bright areas repre-

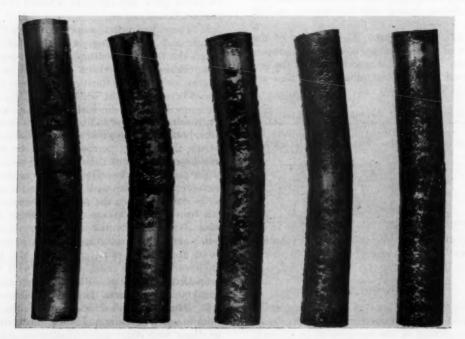


Fig. 7-Bend test specimens, Stages 1 to 5 from left to right. Approximately 1/2 actual size.

C.

16

in

er

n

ze

es

re

d,

n

ie

11

16

senting the silicon needles in alloys in the second stage of modification. The bright areas in a fully modified fracture (stage 4) were rounded in shape.

Bend Tests-A 5-in. length was taken from the bottom of the test bars and bent slowly about its midpoint over a mandrel 1 in. in diameter. The bending was continued until the first sign of cracking on the outer edge of the bar. The angle of bend was measured in degrees. The average angle of bend corresponding to each stage of modification was as follows:

Bend test samples of stages of modification 1 to 5 are shown in Fig. 7.

From these results the ductility of the alloy appears to reach a maximum at a stage corresponding to slight undermodification.

Hardness Tests - Hardness determinations were made on the sections used for the metallographic examination. The determinations were made using a 5-mm diameter indicator and 125-kg load on a Vickers hardness machine. The Brinell hardness numbers of most of the samples were in the range 55 to 65. No significant changes in hardness were found during the progress of modification, the average hardnesses of both unmodified and modified samples being 57. The only change which was found was a narrowing of the scatter of hardness readings as modification proceeded. The average scatter of four separate determinations on each sample decreased from 4 points in the unmodified samples to 1 point in the modified

It was considered that the change of hardness which occurs during modification is too small to permit hardness determinations to satisfactorily determine the stage of modification of any particular melt.

Practical Value of Tests-1. The investigation of several rapid tests which could be used by foundrymen to determine the degree of modification of a melt showed that fracture tests of bars of equivalent thickness to the casting being made would give a reliable indication of the stage of modification of the melt.

2. Bend tests are not considered satisfactory for foundry control as the ductility reaches a maximum at a stage of slight undermodification but may be of use to supplement fracture tests.

3. The variation of hardness with degree of modification is not considered sufficiently large for this method to be used for the following of the progress of modification.

Acknowledgments

The author wishes to thank the Chief Superintendent, Defence Research Laboratories, for permission to carry out and publish the results of the above investigation. The comments and encouragements of Mr. I. J. Ferris, Head of Section, and the practical assistance of Mr. W. L. Crofts are greatly appreciated.

Bibliography

- 1. Troost and Hautefeuille, "Aluminum-Silicon Metal," Ann.
- Chim. Phys., vol. 2, pp. 274-80 (1874).

 2. H. Schirmeister, "The Binary Aluminum Alloys," Stahl ü Eisen, vol. 35, pp. 649-52, 873-7, 996-1000 (1915).
- 3. Z. Jeffries, "Aluminum-Silicon Alloys," Chem. & Met. Eng., vol. 26, no. 16, pp. 750-4 (1922).
- 4. J. D. Edwards, "Properties and Manufacture of Aluminum-Silicon Alloys," Chem & Met. Eng., vol. 31, no. 13, pp. 504-8
- 5. J. D. Edwards and R. S. Archer, "The New Aluminum-Silicon Alloys," Chem. & Met. Eng., vol. 31, no. 13, pp. 504-8 (1924)
- 6. D. Pacz, U.S. Patent 1,387,900 of Aug. 16, 1921.
- 7. J. D. Edwards, Frary and Churchill, U.S. Patent 1,410,461 of March 21, 1922.
 - 8. J. D. Edwards, U.S. Patent 195,048 of Feb. 14, 1923.
- 9. British Aluminum Company, Brit. Patent 210,517 of Oct. 31, 1922
- 10. British Aluminum Company, Brit. Patent 219,346 of Jan.
- 11. R. S. Archer and L. W. Kempf, "Modification and Physical Properties of Sand Cast Aluminum-Silicon Alloys," Trans., A.I.M.M.E., vol. 76, pp. 581-627 (1926).
- 12. B. Otani, "Silumin and its Structure," J. Inst. Metals, vol. 36, pp. 243-67 (1926).
- 13. Anon., "Characteristics and Applications of High Silicon Aluminum Alloys," *Metallurgia*, vol. 26, pp. 67-8 (July, 1941).
 14. V. Zednik, "The Influence of Chemical Composition and
- Quality of Raw Materials on the Physical Properties of Silumin Gamma," Metallurgia, vol. 37, pp. 195-200, 253-8 (1948).
- E. V. Polyak and S. V. Sergeev., "The Viscosity of Aluminum-Silicon Alloys," Compt. Rend. Acad. Soc. U.S.S.R., vol 33, pp. 244-7 (1941)
- 16. A. I. Baikov, "The Modification of Silumin by Salts," Liteynoe, Delo, vol. 8, pp. 23-5 (1939).
- 17. A. A. Botchvar and A. G. Spassky, "Problem of the Modification of Silumin by Salts," Zvetnye, Metally, vol. 10, pp. 81-3
- 18. G. Welter, "Silumin-An Improvement of its Dynamic Elastic Qualities and Endurance Limit by the Addition of Copper," J. Inst. Metals., vol. 36, pp. 325-39 (1936).
- 19. J. Duport and G. Caminade, "The Modification of Alpax," Fonderie, vol. 54, 20947 (1950).
- 20. Silumin G.m.b.H, "Improvement of Aluminum Silicon Alloys," Belgian Patent, No. 449,021 (3.1943).
- 21. "Development of a Ternary Modifying Salt for Aluminum-Silicon Alloys," B.I.O.S. Final Report No. 358, Appendix A. 22. Anon., "A new Process for Silumin," Giesserei Praxis, vol.
- 57, p. 442 (1936).28. E. Schultz, "Microscopic Studies of the Modification of Silumin," Leitschrift gür Metallkunde, vol. 39, no. 4, pp. 123-8
- 24. R. T. Parker, "The Present Position of the Problems of Gases in Aluminum Alloys," Light Metals, vol. 3, no. 27, pp. 107-8 (1940).

DISCUSSION

- Chairman: M. E. BROOKS, The Dow Chemical Co., Bay City, Mich.
- Co-Chairman: W. D. DANKS, Howard Foundry Co., Chicago.
- L. W. EASTWOOD; 1 When sodium is added to an aluminum melt, hydrogen is apt to be added along with it. When the sodium is taken out of a bath of kerosene, for example, and cleaned, it always contains an appreciable amount of sodium hydroxide on the surface. The more sodium you add, the more hydrogen you would expect to add with it. One might conclude that the solubility of hydrogen in the melt was increased because of the sodium addition. I suspect that the increased amounts of sodium merely increased the amount of hydrogen added, making greater difficulties with pinhole porosity. I would like to ask Mr. Sicha which method of sodium addition he prefers?
 W. E. Sicha:

 1 It was my impression from Mr. Dyke's state-
- ment that he felt that it was not a larger sodium addition that was responsible for the increased gas content, but that an increased amount of gas probably would be associated with a larger amount of sodium present in the melt. The implication

was that the sodium content influenced the hydrogen solubility. Whether that is a fact or can be disproved has not been demonstrated conclusively. There is evidence from work that has been done in this country which would lead me to concur with Dr. Eastwood's opinion. This whole question of sodium modification is still important in spite of the fact that there is almost no production today in the United States of the eutectic aluminum-silicon alloys in modified form for the production of castings. However it is my understanding that these alloys are still used to substantial extent in Great Britain and certainly this paper indicates that they are important commercially in Australia. My belief is that we still have a great deal to learn, about effective modification of aluminum-silicon alloys. As long as we are depending on an element such as sodium for modification, we are working with a fugitive material and control and maintenance of a prescribed or specified degree of modification is very difficult.

Our greatest hope for the future, in considering modification of aluminum-silicon alloys, lies in the direction of discovering some other addition element to replace sodium or to supplement sodium that will maintain a high degree of modification.

As far as current practice in the United States is concerned, it is my opinion that addition of metallic sodium is the generally accepted modifying procedure.

O. H. Heil: ⁹ I believe that the second flux used contained 15 per cent of potassium chloride and reference was made to another element for the modification of silicon. Is there any evidence available as to the effect of potassium or calcium?

MR. SICHA: In the paper, there was no indication as to the specific effect of potassium or calcium. However, in earlier work done in this country, the indications were that sodium was the most effective modifying agent and that potassium and calcium were less effective modifying agents. It is conceivable that

potassium may have been introduced into the experimental melts by inclusion of the small percentage of potassium chloride in the salt. However, it is probable that the principal objective of adding the potassium chloride was to reduce the melting point of the mixed salt and thereby obtain greater reactivity.

HIRAM BROWN: 4 Dr. Eastwood made it plain in a previous discussion that calcium affects the ability of the material to be degassed, at least in the excessive use of chlorine. I have noted from my own experience, in many cases, I have been completely unable to degas heats of aluminum alloys with silicon which also contain calcium. Will sodium act the same way to retain gas as well as picking it up, so that sodium might be retaining the actual moisture rather than picking it up?

MR. SICHA: There is evidence indicating that a high calcium content may result in a higher hydrogen content in the melt. Mr. Dyke contended that sodium exerted a similar influence although my observations have not provided conclusive evidence on this subject.

MEMBER: Have you noticed more dross because of the addition of sodium to 356 alloy?

Mr. Sigha: My experience indicates that there is a somewhat greater tendency for oxide particles to be retained in aluminum alloys that are ordinarily subjected to modification treatment. However this characteristic should not present a serious problem unless exceedingly high sodium contents are encountered as a result of repeated and large sodium additions. It should be remembered that modification is attained with only a few thousandths of a per cent of sodium.

T

with

freq

anal

pari

rega

beni

slag

sista

temi

rate

of cau sion pen the pea cast iror S tion stuc occi fect the wise near surf The beir T capa ture

iska Offi Ciety holm

MEMBER: Has anyone attained perfect modification in relation to ductility? I would like to cite a case of 43 alloy where we reached 19 per cent ductility. This I might say is accidental, but do you think that can be consistent?

Mr. Sicha: In view of the fact that the sodium content can change so readily in a melt, it is improbable that you can establish control procedures that would provide 19 per cent elongation consistently.

¹ Assistant Research Director, Kaiser Aluminum, Seattle, Wash.

⁹ Chief, Cleveland Research Division, Aluminum Company of America, Cleveland. Mr. Sicha presented this paper in the absence of the author.

² Metallurgist, The Permold Company, Medina, Ohio.

⁴ Chief Metallurgist, Solar Aircraft Co., Des Moines, Iowa.

AN INVESTIGATION OF THE PENETRATION OF STEEL INTO MOLDING SAND

By Holger Pettersson*

ABSTRACT

The main investigation is based on immersion experiments with cylindrical sand "cores" in molten carbon steel in a high-frequency furnace. It is shown how the rate of penetration varies with time of immersion, metal pressure, temperature and analysis of the steel, grain size of the sand, various additions to the sand, etc. Standard castings have been employed for comparison. The two methods have given contradictory results as regards the effect of certain additions to the sand upon steel penetration. A study of slag formation shows that very little slag is formed at the sand/metal interface before the surface of a casting has solidified definitely. The metal pressure, the capillary forces, the original size of the pores in the sand, the resistance of the sand to enlargement of the pores and the sand temperature are believed to be the factors which determine the rate of penetration.

It is now fully realized that the chief cause of adhering sand is that steel penetrates the sand, causing it to be attached to the casting. Some adhesion may occur in some cases without any appreciable penetration, however. This type of adhesion, which the author has also studied to some extent, would appear to be of no great importance as regards steel castings, but is of great interest in respect of cast iron.

Some information as to the reasons for the penetra tion of the steel into the sand can be obtained by studying how this arises in practice. The penetration occurs as a rule at characteristic spots. The worst affected are those parts of the mold which project into the steel and are more or less surrounded by it, likewise the gates and those parts of the mold that lie nearest a gate orifice. Large flat or cylindrical vertical surfaces, however, are often free from adhering sand. The parts susceptible to adhesion are characterized by being most exposed to heat and/or erosion.

The properties of the steel generally regarded as capable of affecting sand adherence are the temperature and the fluidity. Fluidity, however, is a com-

posite property, involving among other things the temperature, the viscosity and surface tension. With alloy steel there is often in addition a strong disposition to reaction with the material of the mold.

Molding sands have exceedingly complicated properties. Grain size, grain distribution, permeability and sintering point have been found to have significance as regards adhering sand. Other properties, such as high-temperature strength, resistance to thermal shock and expansion and contraction may also be thought to have an effect, as also moisture content and the nature of the binder, e.g., whether organic or inorganic, gas producing or not gas producing. Several of the properties stated are dependent on each other, making it difficult to distinguish their specific effects. Permeability as also sintering point are dependent on the size and distribution of the grains. Both these properties increase with increasing grain size. Strength of sand and other properties of the mold are also dependent on how effectively the ramming has been done and how the mold has dried, etc.

As may be seen, there is a very large number of factors that can be supposed to affect the steel penetration and it is a question in the first place to study which of these are of essential importance. The present investigation is an attempt to survey the action of some of these factors and at the same time an attempt to discover the mechanism of the penetration.

Survey of Previous Work

The oldest opinion was that the adhesion was caused by too low refractoriness in the sand, so that it fused and in some way became welded to the steel. Thus arose the old term "burn on."

A later theory was that the steel was oxidized on the surface, after which iron oxide reacted with the quartz to form low-melting silicates, which were drawn into the pores and bound the sand fast.^{1, 2} After it was found that steel was present in the pores, the theory was modified by the assumption that metal was reduced out of the silicate at a later stage.^{2, 3}

One of the first to abandon the old idea that sand adherence was due to fusion was Vaska.⁴ He considered that actual adherence did not occur before

YC

elts in tive

ted

ain ing

um elt. nce

hat uni

nt.

em

a

be

ew

la

ere

al.

ga-

Research Mining and Metallurgical Engineer, Metallografiska Institutet, Stockholm, Sweden.

Official Exchange Paper to the American Foundrymen's Society from the Swedish Institute for Metal Research, Stockholm, Sweden.

the steel had penetrated into the sand. The condition for the penetration was that the steel remained fluid long enough and that the sand became sufficiently heated for the steel to be able to penetrate into pores and cracks. Moreover, he considered that penetration could be prevented if a dense slag layer were formed by reaction between the sand and oxides formed on the surface of contact. Not too high a sintering temperature in the sand would therefore be an advantage.

Caine^{5, 6} carried out extensive experiments with cylindrical test specimens of different molding sands placed in a mold. He did the casting at different temperatures and measured the depth of penetration. Caine regarded the penetration as a mechanical process, with the depth of penetration dependent on the size of the pores present in the sand to begin with or arising on contact with the steel by partial fusion. Further, he considers that the temperature of the sand must at least be above the solidus temperature of the steel for the penetration to be possible. He found that the penetration depth increased rapidly with increasing grain size in the sand and with increasing casting temperature.

Effect of Sintering Point

The effect of the sintering point was that as a rule a low-sintering point was favorable, but at the same time he considers the most important cause of penetration was the formation of large pores in the sand through fusing, particularly of the smaller sand grains. Caine does not attach any special significance to the chemical reactions. He goes so far as to state that the presence of silicate of iron in the sand after casting could never be demonstrated. He found that an admixture of organic substance, e.g., flour or dextrin, appreciably diminished the tendency to adherence, which he considers to be a result either of the increased gas pressure or of the steel becoming "over reduced" and consequently sluggish. From experiments with impurities in the sand, Caine draws the conclusion that if refractoriness is too low this may be an important cause of penetration-an opinion that appears to have been accepted in other quarters also.7

Woodliff⁸ found that penetration due to crack formation in the sand (veining) occurred especially when the sand was uniform, with the bulk on three adjacent screens.

A theory that the steel penetrates into the sand in the form of vapor has been put forward by Jones.⁹ He observed (with green-sand casting) that if the steel is tapped off or through some accident runs out of the mold immediately after pouring there is found a shell of solidified steel adjoining the mold walls and the sand is fairly easily stripped from this shell. From this he draws the conclusion that the penetration takes place after solidification has proceeded further and assumes that it can therefore only take place via gas phase.

Beeley and Protheroe¹⁰ have studied the effect of the casting temperature on the sand adherence by means of casting experiments. They found that for each kind of sand there was a critical temperature, at which penetration started. The type of casting used was of small section, so that extremely high temperature was required for penetration to occur.

The reactions in the mold-metal interface have lately been considered with great interest. Dietert and co-workers¹¹ demonstrated as early as 1944 how it was possible by heating a steel pin embedded in sand to obtain the same reaction products as in steel casting. Nevertheless all investigations have been devoted to the final products after casting and therefore comprise the reactions in the solid state.^{10, 12, 13, 15, 17} Nor has it been possible to draw any conclusions of significance from the results of the experiments. In all cases fayalite was demonstrated in the sand nearest the steel.

Portevin and Bastien¹⁴ demonstrated as long ago as 1936 on a purely theoretical basis that the capillary forces must be of great significance for the penetration of the steel into the sand. This appears, however, to have been neglected and not until quite recently were a couple of papers published in which the capillary forces were considered.

An interesting work is that carried out at Cambridge by Hoar and Atterton. They have gotten away from casting conditions and made use of ingenious apparatus to force molten steel into the sand by means of gas pressure. In this way they have shown that the critical pressure required for the metal to penetrate is dependent on the surface tension of the metal. They assume, besides, that the grain size of the sand has great effect on the sand's thermal diffusivity and therefore the depth of penetration. The penetration is assumed to take place instantaneously and immediately a sufficient pressure is reached. The requisite pressure is stated as being as high as 35 to 110 cm steel column.

Effect of Static Pressures

In Canada, Gertsman and Murton¹⁶ have carried out casting experiments with different static pressures. They do not seem, however, to have reached the conclusion that the capillary forces were responsible for their results. Time-temperature curves taken from the cores show that the temperature rises more rapidly in coarse sand than in fine sand.

In Sweden, the problem of adhering sand was studied in conjunction with an extensive investigation into the question of sand for the casting of steel, which was carried out by the research organization of Jernkontoret (The Swedish Ironmasters' Association) some years ago and which was published by Ljunggren.¹⁷ The adhesion was found to increase with the gas permeability, increased grain size and increased sintering point for the sand. On the other hand, no difference was found with the employment of various clays. From a microscopic investigation carried out by Prof. P. J. Holmquist it was ascertained that the adhesion was due to the steel penetrating into the sand and that chemical reactions took place which gave rise to slag formation. It was found that the slag formation was strongest in those layers of the mold which were easily detached from

ine dir (ir ora me los

otl

H

th

wi

by

su

th

uc

ob tio

lay era for clu wa tio

ma mo eve pro abl

witinv cyliproten of

per

wit me

me. by SAND

ture,

sting emp-

have

how

d in

steel de-

iere-

2, 13,

clu-

peri-

the

ago

lary

etra-

re-

the

am-

ten

in-

and

wn

to

the

the

vity

tra-

im-

re-

110

ied

res-

ned

on-

cen

ore

ud-

ion

eel,

ion

ia-

by

ase

nd

ner

ent

on

er-

ne-

ons

vas

ose

m

the casting. The pores there were entirely filled up with slag products, mostly glass, fayalite (ferromanganese silicate), oxide of iron and tridymite. It was therefore assumed that the slag formation might counteract the penetration.

As with the present work, most of the earlier investigations deal with dried sands. In an investigation by the Swedish Ironmasters' Association Research Organization, recently published by Berglund, 18 concerning sands for green-sand casting, a number of results of general interest were obtained. Among other things, it was found that an addition of cereal products in contents exceeding 1 to 2 per cent gave an increased disposition to penetration, whereas an addition of sludge containing iron oxide compounds (initial product for a red paint) had a markedly favorable effect. Moreover the relation between permeability or grain size and penetration seemed to be lost with fine sands owing to the superimposition of other factors.

Present Investigation

Scope of work. Experimental procedure—The first object of this work was to study the chemical reactions in the zone of contact between steel and sand.

To begin with the author had to visit a number of steel foundries to study how the problem looked in practise and to obtain suitable samples for closer investigation.

Samples of both adhering and non-adhering sand layers were then studied in metal microscopes, mineral microscopes and by chemical analysis. It was found difficult, however, to draw any fruitful conclusions from the observations. The reason for this was apparent when it was learned that all the reaction products observed might be formed in contact between *solid* steel and sand. Attempts were then made to study the slag formation when the steel was molten and when it was solid. It was realized, however, that it was necessary to try to attack the whole problem of adhering sand right from the beginning.

Since 1948 a high-frequency furnace has been available for this investigation. This has enabled the experiments to be carried out under good control and with fairly large possibilities of variation. The main investigations have been based on the immersion of cylindrical sand test specimens in molten steel. The properties of the sands, depth below surface, steel temperature and composition as well as the duration of contact between steel and sand have varied.

A number of casting experiments were made as supplement to the immersion tests.

Microscopic examination has been made parallel with the other investigations throughout the time.

The immersion experiments and the casting experiments are described below.

Immersion Tests

The action of a circulating, but not very eroding, melt on molding sand enclosed by it has been studied by means of cylindrical test specimens of molding sand attached to mild steel tubes being maintained

immersed in molten steel in a 150-kg (320-lb) high-frequency furnace. During the time the sand was immersed in the molten metal the temperature was kept constant.

The sand molds were mixed in a laboratory mixer. Normally 5 per cent bentonite (Western bentonite) was used as bond.

Cylindrical test specimens of 50-mm diameter (approx. 2 in.) and 50 mm high were rammed in a test piece rammer of the type usually employed in foundry laboratories and which gives constant ramming action for an equal number of blows.

The method first used for the attachment of the test specimens was later replaced by one more suitable. Only the latter will be described here.

After a modification of the usual standard rammer had been constructed, test specimens with a hole right through were produced by the method shown in Fig. 1. The fitting was done on perforated tubes of mild

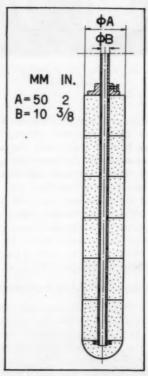


Fig. 1-Arrangement employed for preparing sand specimens.

steel. As a rule a set of identical test specimens was mounted on one tube, enabling the effect of the hydrostatic pressure conveniently to be studied. The joints between the specimens were tightened by a mold wash. The ready fitted rods were as a rule dried at about 390 F (200 C) for one night.

The dried rods were taken out of the drying furnace about half an hour before the immersion test, the depth of immersion being marked with a string. When only the penetration in the lowest test speci-

men was to be studied the immersion depth was measured from the middle of the test specimen.

To facilitate the experiments a holder device provided with weights was made, with which four tubes could be kept immersed in the furnace at one time (Fig. 2).

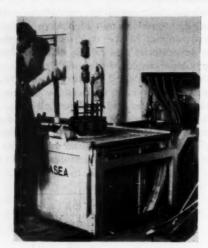


Fig. 2-Arrangement for holding rods during immersion.

The temperature of the steel was checked by means of immersion thermocouples. Screen curves were plotted for each sand tested and the average grain size was computed. In addition the green permeability for air, green strength and density were determined for the prepared molds.

Penetration depth in mm was used as measure of penetration. When penetration takes place in the form of very scattered veins this is not sufficient. The measurement of penetration depth can only be approximate (about \pm 5 per cent) as individual metal tips may shoot in for different distances.

Laboratory Casting Experiments

The object with the casting experiments was not to study systematically by casting the problem of adhering sand, but only to complement the results obtained with the immersion method.

Efforts were made to obtain information about the effect of: varying grain size and grain distribution, different hardnesses of ramming, different binders and binder contents, admixture of reducing or oxidizing agents, internal gas pressure in the metal on solidification, etc.

If adhering sand (metal penetration) is to be studied by means of casting experiments without taking abnormal measures, heavy pieces must be cast, preferably with a weight exceeding 500 kg (1000 lb). That, of course, is not possible in a laboratory, where only small test pieces can be cast. Because of this it was in many cases difficult to arrive at definite conclusions from the results of tests.

Three cast test specimens, A, B and C, were employed: Test specimen A (Fig. 3) with a cast weight

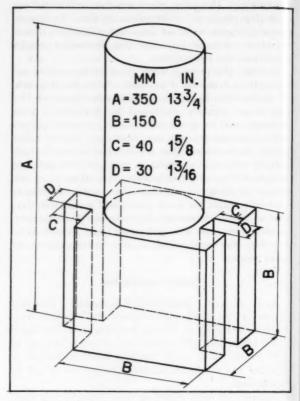


Fig. 3-Cast specimen A. Weight of casting about 80 lb.

of about 36 kg (approx. 80 lb) was that usually employed. It has in relation to its size a great tendency to penetration.

Test specimen B (Fig. 4) with a cast weight of about 20 kg (approx. 44 lb) was employed for the study of penetration in internal cores.

Test specimen C (Fig. 5) with a cast weight of about 7 kg (approx. 15 lb) was employed in some experiments with raised gas pressure in the metal.

Before pouring, the temperature was checked by immersion thermocouples in furnace and ladle. The pouring temperature was usually around 1575 C (approx. 2870 F).

Observations with Adhering and Non-Adhering Sand Layers

In favorable cases the mold material is easily detached from the casting and the sand adjacent to the steel can then usually be removed in the form of coherent layers or cakes. Such a cake is made up of different zones. A thin, more or less cohesive surface zone which has lain close up against the steel is blueblack and consists for the most part of iron oxides (Fe₂O₃ and Fe₃O₄). This zone corresponds to a similar layer on the steel, which in this case has a relatively smooth dark blue surface. Inside the surface layer comes a dark gray-black zone, the thickness of which is dependent on the size, etc., of the casting. With heavy castings the zone is usually 3 to 5 mm.

MM IN.
A=200 73/4
B=150 6
C= 30 13/16

B

B

Fig. 4-Cast specimen B. Weight of casting about 44 lb.

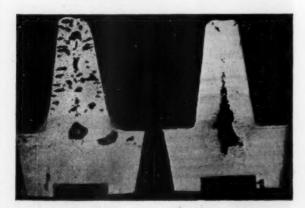


Fig. 5—Two cut specimens of type C weighing about 15 lb. The left section shows penetration caused by internal gas pressure.

Next follows usually a more or less thick zone, in which the sand is sintered and almost white in color. The light color is due to the organic substances having been burnt away. After the sintered zone there may come a less firmly coherent zone with a brownish color, which probably is derived from organic sub-

stances that have been scorched. Lastly comes the unaffected sand.

Only the dark zones that have been nearest the steel are of interest. The blue-black surface layer, as stated, consists of iron oxides and has obviously been pulled off the steel during cooling. Owing to the different shrinkage in the steel and the sand the oxide layer has been worn to pieces and partially accompanied the sand, to which it appears to adhere firmly and solidly. On the other hand, it is well known that in the oxide crust on steel an outer layer consisting of Fe₂O₃ + Fe₃O₄ is easily scaled from the steel. The fact that even the skin of the casting assumes a blue color from Fe₂O₃ probably means that the separation of the sand was done at a fairly high temperature and that the steel surface was again oxidized afterwards. From these observations it would seem that the sand detaches easily from the casting when the casting is smooth (no penetration) and the oxidation is sufficient for the formation of an annealed crust rich in Fe₂O₃. Some experiments that have been performed also confirm that the sand loosens less easily if the oxidation of the steel surface does not go further than to the approximate composition FeO. (If, as with cast iron, it is possible practicaly speaking to prevent oxidation entirely, then again the sand detaches easily).

The dark zone inside the outer layer is found on microscopic examination to consist of quartz grains, more or less cracked nearest the surface of contact and partially converted to cristobalite though never to any great extent. The interstices between the quartz grains are almost entirely filled with slag products (Fig. 6), chiefly fayalite, tridymite and glass of yellow or brown color and usually small amounts of iron oxide (Fe₃O₄).

When the sand adheres very hard to the casting it

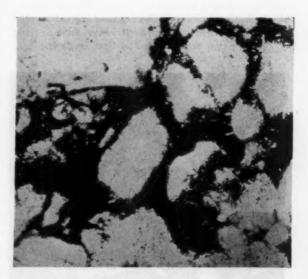


Fig. 6—Photomicrograph of non-adhering sand layer. Spaces between quartz grains filled with dark slag. Transmitted light. Mag. 75x.

1

SAND

em-

out

the of ex-

by The (ap-

dethe of of ace uedes

of ng.

mi-

ela-



Fig. 7A-Macrograph of adhering sand layers. Steel (light colored) has penetrated the sand. Mag. 10x.

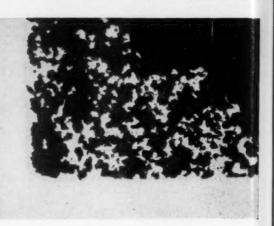


Fig. 7B-Same as Fig. 7A. Mag. 5x.



Fig. 8—Sectional view of adhering sand layer. Rounded iron beads in the quartz. Mag. 600x.



Fig. 9-Sectional view of adhering sand layer. Quartz (darkest), slag is fayalite and iron beads in eutectic arrangement. Mag. 500x.

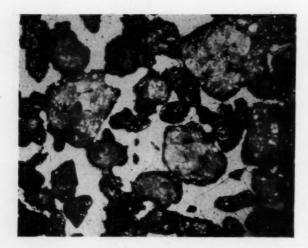


Fig. 10-"Dense penetration" in coarse sand. Mag. 50x.

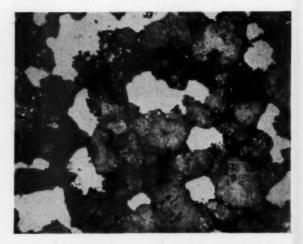


Fig. 11-"Sparse penetration" in fine sand. Mag. 50x.



Fig. 12-Sectional view of adhering sand layer. Strings of slag between sand grains. Mag. 300x.

is found that the steel has penetrated into the interstices between the grains (Fig. 7). In some cases, especially when the steel has eroded the mold there can be said to be an inter-effect, with the sand penetrating into the steel at the same time. In those parts where much steel has penetrated, the sand is not particularly dark in color and on microscopic examination is found to contain comparatively, little slag constituents. The quartz grains often have cracks and may be partially converted. Inside the quartz grains, and in slag formations present between them, small beads of iron are encountered, which were in solution at high temperature and have later separated during cooling (Fig. 8). Sometimes eutectic structures of fayalite, iron and tridymite or glass occur (Fig. 9).

The microstructure of the steel present in the interstices of the sand is largely the same as in the casting itself.* Existing departures from the normal struc-

ture may be explained as a result of oxidation. An observation of great interest is that the metal that has penetrated is distributed in the sand differently in different cases. In a sand with high average grain size (and high sintering point) the metal fills up the spaces between the grains fairly uniformly (Fig. 10). In fine sands, on the other hand, the intruding metal often forms rather coarse, more or less sparse, veins or branches, not to be confused with "veining." Between the metal veins the quartz grains lie in densely packed groups often sintered together (Fig. 11). This shows that mechanical changes have occurred in the sand in conjunction with the penetration. This is supported by another observation, namely that there can be seen in the microscope slag strings running through the steel from one sand grain to another (Fig. 12), which suggests that the grains are split apart by the penetrating metal. (This interpretation of the slag strings is not, however, the only possible one).

A special case is constituted by that where a smooth and almost dense surface has been produced on the mold by some kind of mold wash, but still there has been penetration. The sand lying behind may then be filled with steel, while the outer surface is permeated only here and there in pores and cracks. Such a mold layer is often rather easy to strip.

Distribution of Penetrated Steel in Sand

As stated above, it can be observed in the adhering mold layer that the penetrating steel is distributed in different ways in different cases. This condition can be studied more systematically in test specimens from the immersion tests.

If the sand is well graded, it is usually possible to distinguish an outer zone where the steel fairly completely fills the interstices between the grains, and an inner zone which extends 2 to 3 mm farther in the steel in the form of veins becoming more and more sparse inside (Fig. 13). This inner zone has often a rather uneven front. The type of penetration represented by the outer zone, i.e., well-filled pores, will hereafter be referred to as dense penetration while the other type will be called sparse penetration.

tic

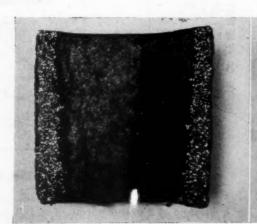


Fig. 13—"Dense penetration" in sand specimen; well graded sand. Approximately actual size.



Fig. 14-"Sparse penetration" in sand specimen; widely distributed sand. Approximately actual size.

^{*}This fact refutes the theory advanced by Jones^o that the steel penetrates in the form of gas.

In some cases, where the grain size (really the pore size) is small in relation to the pressure of the metal, no dense penetration occurs, but the penetration takes place solely in the form of separate or only partially connected veins, which are coarse in comparison with the average grain size. With widely distributed sands this is the usual type of penetration. Even in these cases it is possible to distinguish still sparser zones with uneven fronts (Fig. 14).

The presence of these zones may be explained in the following manner: To begin with, penetration takes place only in the largest pores existing in the sand. Even afterwards penetration takes place in these first. Gradually, however, a part of the smaller pores are also filled. How large a part is filled depends on the metal's pressure and the grain distribution. If the pressure is high, the greater part are filled; if the pressure is comparatively low, however, or if the size of pores varies greatly, as in widely distributed sands, then the steel can only penetrate into the larger pores and these are then enlarged at the expense of the smaller. Even with well graded sand and high pressure such displacements in the sand would appear to occur, but they will not be of any great extent, as there are then available for the steel several equivalent paths.

Effect of Factors on Penetration and Adherence

1. Effect of the time factor—In immersion tests it was possible to observe that, at least at low temperatures, a steel layer was formed around the sand mold. With longer immersion this layer melted. No penetration occurred as long as the steel crust was whole. The melting time for different temperatures was therefore studied by test specimens of a given sand

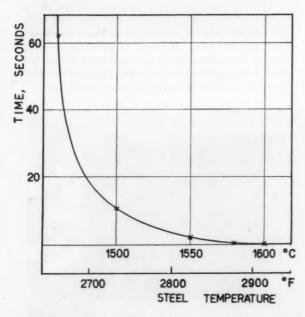


Fig. 15—Relation between melting time for the steel shell formed on immersion and the temperature of the molten steel.

being submerged somewhat under the surface of the steel and maintained there for a series of times of different lengths. The thickness of the steel crust, which first grew and then diminished, was studied and the time for complete melting was noted. Figure 15 shows the melting time as function of the temperature. As may be seen, the melting time decreases rapidly with rising temperature, approaching zero at high temperatures, i.e., no steel crust worth mentioning could be formed. At 1550 C (2820 F), used as standard temperature in the experiments, the time of melting is insignificant. The times measured apply, of course, only to the experimental conditions.

As it was of great interest for many reasons to see how the penetration rate varied with the penetration depth (the time) a number of test series were carried out with varying times of immersion in the molten metal. As may be seen from Fig. 16, the penetration depth increases almost linearly with the time up to approximately 12-mm penetration depth, and after that the rate of penetration increases more rapidly With low temperature in the steel the penetration does not begin until the steel layer which solidifies on the mold has re-melted. In the linear range the rate of penetration is evidently almost constant for a given time and the depth of penetration is therefore a measure of the rate of penetration. The fact that rate of penetration increases with greater depth of penetration may be due in part to thermic factors and in part to the ramming not being so effective farther in. That the thermic factors are the most important is shown by sintering experiments made by immersing test specimens composed of relatively easily sintered sands to a depth so small that no penetration was obtained. Figure 17 shows the results. The curves obtained for sintering depth versus time are of the same character as the curves in Fig. 16.

When casting, the time the sand remains in contact with molten steel is determined by a number of fac-

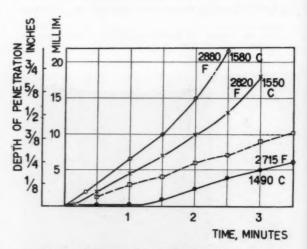


Fig. 16—Depth of steel penetration vs time in the melt a fine sand at a steel temperature of 1550 C (2820 F). at constant pressure (30 cm, 1 ft, steel column). Full curves are for a coarse sand at three different temperatures. The broken curve shows the relation for

in the sol

esp

H

to

in

tic

an

DEPTH OF SINTERING

lor sla sla sar me to

of SHOPPATION TO THE STATE OF THE SHOPPATION TO THE SHOPPATION TO

10 HI 1/2

Fig. an per fig. mr. GI

the

NI

he

of

st

ed

ire

ra-

ses

at

n.

as

of

ly,

ee

n

ed

en

to

er

m

n

te

n

te

e.

ì

n

1

e

t

tors, e.g., the size and shape of the mold, the casting temperature and the manner of casting. In practice, penetration is most likely to occur in those parts and those castings where steel remains fluid longest

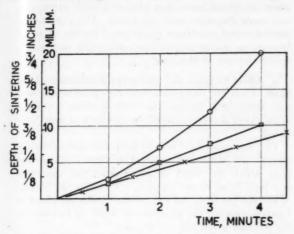


Fig. 17—Depth of sintering vs time of immersion in molten steel for three sand specimens with different sintering temperatures.

in contact with the sand. It appears to be usual that the penetration is preceded by a thin steel crust first solidifying on the sand and then being re-melted, especially in the vicinity of gates. Even after solidification adhesion is influenced by the time factor. The longer the casting is in cooling the more oxides and slag are formed in the mold-metal interface and this slag formation seems to favor the detachment of the sand from the casting.

2. Effect of static pressure of the metal—With the method employed by the author it has been possible to obtain a continually varying pressure on the same sand, from zero at the surface of the metal to a maximum of about 40 cm steel column at the bottom of the furnace. Figure 18 shows the relation between

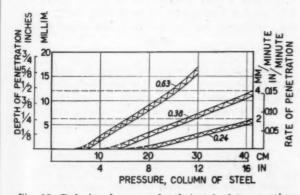


Fig. 18—Relation between depth (rate) of penetration and the pressure of the melt on the sand. Steel temperature 1550 C (2822 F). Immersion 3 min. The figures on the three curves give average grain size in mm for the corresponding sands, equivalent to A.F.S. GFN 27, 44 and 65 respectively. The irregularity of the penetrating steel front is indicated on the curves.

the static pressure and the penetration depth for three sands of different coarseness. The steel temperature was 1550 C (2820 F) and the time of immersion 3 minutes. Curves of the same type were obtained in other tests, but to simplify the graph only three curves have been included. As may be seen, a certain minimum pressure (minimum depth) is required for any penetration to take place. With rising pressure there is obtained to begin with a very sparse penetration and gradually the penetration becomes denser and the depth of penetration (rate of penetration) starts to rise in proportion to the increase of pressure. The coarser the sand is, the smaller the pressure required for penetration to start and the more rapidly the penetration takes place.

The significance of pressure for penetration has been confirmed by casting tests, partly by casting in molds of different heights and partly by producing gas pressure in the metal on solidification. By adding water to the melt or by incomplete deoxidation of the steel it has been possible to get penetration even in very small castings (Fig. 5) and with larger castings the result has sometimes been exceedingly deep penetration. It should be observed that the maximum static pressure of the metal is not determined solely by the height of the mold. The metal pressure can only be effective if the sand is in contact with molten metal. The maximum effective pressure in a mold at a given moment therefore is the difference in level between the upper surface of the molten metal and the lowest part of the mold where the steel has not yet solidified to a dense layer.

3. Effect of temperature of the metal—The effect of temperature on the rate of penetration was investigated by immersion tests in steel melts with approximately the same analysis but different temperatures. Figures 16 and 19 show some of the results. As anticipated the rate of penetration increased with the temperature.

A high casting temperature should therefore in itself lead to an increased disposition to penetration. At the same time a high metal temperature (in re-

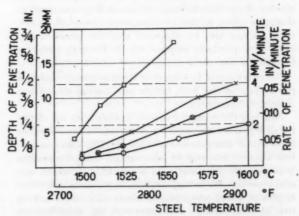


Fig. 19—Depth (rate) of penetration vs steel temperature. Immersion in the melt 3 min. Depth below surface 30 cm (1 ft). The four curves show the results with sands of different grain sizes.

ŀ

d

re

ti

0

m

V

tr

si

Iı

m

th

gı

u

h

A

it

to

aı

te

in

th

im

th

tic

th

m

fa

lation to the solidification temperature) means that the time for contact between the molten metal and the sand will be longer and the maximum effective pressure will be greater than normal, and this also favors penetration. The fact that the casting temperature is of decisive significance for the penetration is also confirmed by the casting tests.

4. Effect of composition of the molten metal—Only mild carbon steel and cast iron have been used in the tests. The contents of carbon, silica and manganese have varied within fairly wide limits: C=0.08 to 3.5 per cent, S=0.05 to 1.0 per cent, S=0.05 to 1.2

per cent.

A noticeable effect on penetration was observed only from the carbon content. From the tests made with varying carbon content it appears that the resistance to penetration increases continuously with increasing carbon content. On immersion in cast iron under normal test conditions (depth 30 cm, approx. 1 ft; temp. 1550 C, 2320 F) a penetration depth of 2 to 3 mm was obtained with the coarsest of the sands tested (average grain size = 0.63 mm - A.F.S. 27). With normal grain sizes no penetration was obtained. In the normal range of carbon content for steel castings, the effect of variation in carbon content was also quite noticeable, though the difference should not have any practical significance.

If the manganese content is so high in relation to the silicon content that the quartz is attacked, one might expect that the penetration would be strongly affected. As the lining of the furnace used for the tests is acid and is therefore severely attacked by high manganese contents, the contents for the experiments had to be limited to not more than about 1 per cent. No marked effect of the manganese content appeared in spite of the lining and the mold

being attacked.

No clear effect of different silicon contents were noticed. With high silicon contents a slight tendency to increased penetration could possibly be observed.

With plentiful addition of aluminum to the steel bath before and during immersion the penetration was less deep than normal. It is not, however, certain that this effect is due to aluminum dissolved in the steel. It can also be supposed that the pores in the sand were partially stopped up by the slag particles formed in the melt on addition of aluminum.

No effect could be observed from pouring water on the surface of the bath before and during immersion

(effect of hydrogen).

Nor was there observed with casting tests any definite difference in respect of penetration for different contents of manganese and silicon. The effect of carbon content was not so pronounced with the casting tests as with the immersion tests. This is due to several reasons. With high carbon content solidification begins at lower temperatures than with steels having low carbon content and moreover the solidification range is wider. The solidification time therefore will be longer with high carbon content than with low, particularly if the casting temperature, as in this case, is the same. During cooling the oxidation of the

iron in the surface will be less. Due to this fact the sand will adhere to cast iron and steel with high carbon content even if no penetration takes place. Nevertheless, penetration has often taken place in castings when no penetration was obtained with immersion tests using the same sand and metal. The penetration then occurred mainly in the form of fin-like or warty formations and appears to be connected with mechanical changes in the sand.

5. Effect of grain size and grain distribution of the Sand—Over 100 immersion tests were carried out with pure silica sand. Besides the original sands, a number of mixtures and fractions of various sorts of sand were tested.

The test procedure was to keep test specimens immersed for 3 minutes in steel at 1550 C (2820 F), after which the depth of penetration was measured at a level lying 30 cm (approx. 1 ft) below the surface of the molten steel. In Fig. 20 the results are given in a graph showing the penetration depth as function of the average grain size of the sand.*

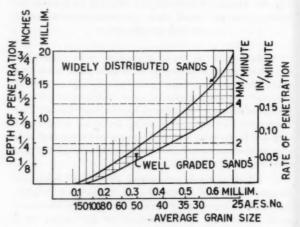


Fig. 20—Depth (rate) of penetration vs average grain size of sand for an immersion time in the melt of 3 min and a steel temperature of 1550 C (2820 F). The hatched area applies to sands with a wide grain size distribution and the checkered area to well-graded sands.

Practically all the individual values lie within the areas of the curves plotted. For sand sorts that were well graded most of the values lie well together, whereas the dispersal for widely distributed sands can be quite large. As may be seen, the penetration in-

This chart must be regarded against the background of Fig. 18. The curves of the type given there furnish a more complete picture of the properties of the sands.

^{**} In respect of penetration it is the pore size that is of the greatest interest. With graded sands this will be fairly well defined by the average grain size. This is not the case with a widely dispersed sand where the size of pores will vary considerably. Despite the denser packing such a sand usually contains a large number of pores that are larger than those in a graded sand of the same average grain size. The same applies, e.g., to a blend of a graded sand a very fine fraction.

ANI

the

car

ver-

ings

sion

tion

arty

ech

the

vith

ber

ere

im.

F),

red

ace

in

of

PENETRATION

OF

in

of

in

d

e

e

r,

n

of

a

creases rapidly with the average grain size. For the graded types of sands the relation between grain size and penetration depth could be interpreted as quadratic. It appears, however, more correct to assume that the relation is mainly linear but the penetration depth for the lowest grain sizes is greater than corresponds to the linear ratio. Moreover the penetration occurs more rapidly at greater depth on account of the shape of the test specimen.

The explanation of the branching off for small grain sizes is supported by the penetration changing in type-becoming sparser-at the same time as the curve begins to bend. As stated earlier, this probably is connected with displacements taking place in the sand which make possible penetration by veins of metal that are coarse in relation to the grain size. As a rule with widely distributed sands the depth of penetration is greater than with graded sands. With very scattered grain distribution the depth of penetration can be extremely large in relation to the grain size, but at the same time it will be exceedingly sparse. In such cases the depth of penetration is not a good measure of adhering tendency, as the sand with such very sparse penetration can very well strip easily from the casting, even though the penetration depth is great.

Drawing conclusions from the casting tests depends on subjective judgment and is therefore rather uncertain. The tendency with the casting tests has, however, been the same as with the immersion tests. A disposition to penetration—often within strictly limited areas—appears, however, in cases where according to the immersion tests one would not have expected any noticeable penetration.

6. Effect of the sand material—In the immersion tests other materials besides silica were tried, viz: Norwegian olivine, unpurified corundum (abrasive material), electrically sintered magnesia and crushed products of chromium magnesite-and fire bricks. The greatest deviation from conditions with silica sand occurred in testing the electrically sintered magnesia, as a particularly great penetration was obtained and in many cases the test specimens burst on immersion. This magnesia was obviously not dead roasted, but had absorbed moisture and carbon dioxide while in store. The same tendency was found in immersion tests with crushed fire brick. The other materials gave results which to a large extent agreed with those obtained with silica sand. This circumstance suggests that the transformations of the silica have no decisive importance for penetration. It should be remarked that none of the tested materials showed less disposition to sintering than the pure silica sands.

Less pure silica with lower sintering point than is usual with sand for casting steel was also tested. It was then found that with decreasing sintering point the depth of penetration is not appreciably affected but that the penetration is more and more sparse and may cease if the sand is not too coarse, due to the fact that a dense sintered surface is obtained.

Test specimens made of easily sintered silica sand or of fire brick showed considerable shrinkage after the tests and were often much deformed, which indicates that the material was plastic. Test specimens of olivine sand also shrank considerably, but showed no deformation whatever.

When casting on sand with a low sintering point the casting had an unsightly surface.

7. Effect of binders and admixtures in the sand—
(a) Immersion tests—The test procedure was the same as in the study of the effect of grain size (depth 30 cm, approx. 1 ft, time 3 min, temp. 1550 C, 2820 F).

Various types of clay did not lead to any difference as regards penetration. With high clay contents, 10 per cent or over, there appeared a change in the nature of the penetration. This was chiefly caused, however, by sintering starting to take place then.

Silica flour in contents less than 10 per cent had very little effect on the penetration. With higher contents the tendency to sintering was clearly increased.

Admixture of pulverized oxide of iron (Fe₂O₃) or manganese or (MnO₂) likewise resulted in increase of sintering tendency.

The addition of cereal substances, such as flour and dextrin, had as a rule a definite preventive effect on the penetration, but not with coarse sand. With the higher contents of dextrin, 2 per cent or more, the penetration could sometimes be increased. Lampblack and powdered graphite also decreased the rate of penetration.

The addition of powders of aluminum, silicon or silicon carbide hindered penetration. The effect of aluminum was decidedly the strongest and 2 per cent was sufficient to produce a noticeable effect. With 1 per cent aluminum (and over) penetration was entirely absent in sand molds with an average grain size in the sand of less than 0.33 mm. With higher contents of A1 (> 2 per cent) the mold became hard sintered.

b. Casting tests—(1). Test specimen A: Different types of, or different contents of clay binders had no manifest effect on penetration.

On admixture of 15 per cent kieselguhr (fossil meal) the sand was strongly sintered and the surface of the casting was unsightly and irregular. A certain amount of penetration occurred in the grooves (pores) of the sand.

Addition of small contents of iron oxide had no noticeable effect. With larger contents the casting surface became unsightly.

The addition of graphite (fairly coarse) resulted in a coarse unsightly surface and the adherence between the sand and the casting was increased.

Addition of aluminum, silicon, silicon carbide and dextrin had more or less the same effect. With small contents the effect was on the whole favorable, though the adherence caused by reactions in the interface was increased. With higher contents there set in, particularly with aluminum and dextrin, a tendency to veining in the groove and for a penetration which, especially in limited areas, could be very deep. Whereas dextrin counteracted sintering, Al, Si and SiC had the opposite effect.

(2). Test specimen B: The admixture of various

substances in the cylindrical core gave a more favorable effect than in the mold walls of test specimen C. Higher contents of dextrin and aluminum could give rise to deep veining, whereas lower contents had a good effect. A pitch emulsion that was tried had the same effect as dextrin. The organic substances made the core particularly easy to clean, as they counteracted sintering. Other substances had the opposite effect. Oxide of iron had a certain ability to counteract veining.

8. Effect of moisture content of the sand—Immersion tests with varying moisture content (including normal green sand) did not allow any definite conclusions. As a rule the effect of moisture content was surprisingly small or imperceptible, apart from explosions which might occur with high contents.

In casting tests the sand was more easily detached from the casting with green sand than with dried sand, owing to the oxidizing action of the moisture.

9. Effect of the ramming density—As there was reason to believe that loosely rammed sand might offer less resistance to penetration than well rammed sand, a number of tests were made of test specimens with different degrees of ramming. The following results were obtained for two different steel temperatures.

No of rammer blows		5	2	1	1/2	
Density, g/cm ³		1.49	1.43	1.40	1.35	
Permeability		308	347	398	512	
Penetration depth at 1550 C (2790	F)	3	3	3	max	7
" " 1560 C (2810	F	7	7	7-8	max	13

Definite effect of the degree of ramming was thus obtained only with very loose ramming. At the same time the character of the penetration changed and two zones could be distinguished: an outer zone with the same appearance as for hard-rammed test specimens and inner zone where the penetration was sparse and uneven and had obviously proceeded very rapidly. The cause of this is obviously that even with loose ramming of the test specimens a fairly tightly packed outer layer arises.

In casting there occurred a great disposition to veining with loose ramming.

10. Effect of permeability of the sand — It seems natural to assume that there would be a simple relationship between the rate of penetration of the steel in a sand and the ability of the sand to allow the passage of gas. Figure 21 gives graphically the relation between the permeability of the sand and the depth of penetration on immersion to 30 cm depth (approx. 1 ft) in steel at 1550 C (2820 F) and an immersion time of 3 min. The graph has approximately the same character as the graph showing the relation between average grain size and depth of penetration (Fig. 20). The correlation is better, however, in Fig. 20 than in Fig. 21.

11. Effect of mold washes—To begin with some attempts were made to produce a dense surface by painting the test specimens before drying with pastes of various refractory materials in water along with bentonite or fire clay as binding materials. Only a slight

decrease of penetration depth was obtained. Nor did the admixture of aluminum in the mold wash have any significant effect. A proprietary American mold wash, however, displayed good effects. On contact with the molten steel there was formed a dense glazed surface which effectively prevented penetration by the steel. Similar results could be obtained later with a mold wash with a silica flour base, of a composition more suitable than those previously tried. Nevertheless a certain variation in results could be obtained even with the best mold washes.

In casting tests the sand was easily detached if suitable mold wash was employed. This was due in the first place to a smooth surface being obtained, but the penetration and veining were also prevented to a great extent. When considering the favorable results with mold castings it must be remembered that the experimental molds employed were very small.

12. Effect of thermal diffusivity of the sand—The thermal diffusivity of the sand should have some influence on the rate of penetration. Judging from the different experiments with various materials in the sand, differences in the heat conductivity of these materials would have no great significance. On the other hand it may be supposed that grain size and grain distribution would appreciably affect the thermal diffusivity of the sand. 15, 16 In order to test this, some orientating immersion tests were made with thermocouples inserted in the test specimens, with the hot junction about 5 mm (0.2 in.) from the surface.

Only two grades of sand, with average grain size 0.32 and 0.64 mm (A.F.S. 50 and 26 respectively) were employed. Figure 22a shows the time-temperature curve for a depth of 30 cm below the surface of the steel. It is indicated on the curves when the penetrating steel reached the hot junction of the thermocouple. Figure 22b shows the corresponding curves for a depth of 5 cm (2 in.) below the surface of the steel, when no penetration occurred. In the

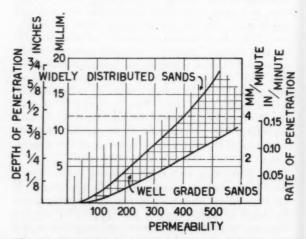


Fig. 21—Depth (rate) of penetration vs green permeability of the sand. Immersion time in the melt 3 min. Depth under surface of melt 30 cm (1 ft). For designations see Fig. 20.

SAND

did

ave

old

tact

azed

by

vith

tion

the-

ned

uit-

the

but

o a

ults

the

The

in

the

the

ese the

nd

m-

nis.

ith

ith

ur-

ize

ly)

ra-

ice

he

he

ng

ce

he

PENETRATION

PF

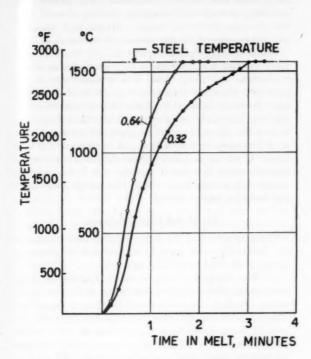


Fig. 22A-Variation of temperature with time of immersion at a point about 5 mm inside the surface of two sand specimens with average grain size 0.64 and 0.32 mm (A.F.S. GFN 26 and 50 resp.). In the course of measurement steel has gradually penetrated the sand. It has reached the thermocouple after about 11/2 min in the coarse sand and after 3 min in the fine sand.

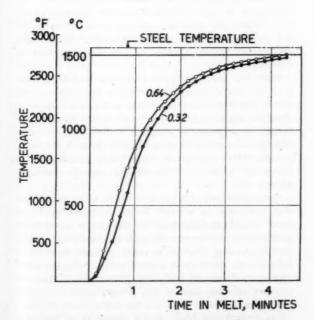


Fig. 22B-Time-temperature curves corresponding to those of Fig. 22A but with no penetration of steel.

latter case, no definite difference was seen between the two grades of sand. As it was difficult to insert the hot junction exactly 5 mm from the surface, the results for different tests showed fairly great deviation and the curves given must be regarded as only approximately correct.

13. Effect of pre-heating of the sand—In casting the sand is first subjected to heating by radiation from the rising melt and by hot gases. In the time taken for the steel to rise from a given level in the mold to so high a level that the pressure is sufficient for penetration to take place, the sand is also heated by direct contact with the molten metal. Some attempts were made to study more closely the effect of such preheating.

By holding test specimens just above the molten metal before immersion to 30 cm depth an exaggerated effect of the heating from a radiating molten metal and by hot gases could be studied.* With the comparatively fine-grained material used an appreciably greater penetration than normal resulted. At the same time there were generally obtained deep metal-filled cracks (veining). Sometimes parts of test specimens broke off during pre-heating. (Such spalling off may give rise to scabs when casting.)

In a similar way the effect of a strong pre-heating from contact with the molten steel was studied by holding test specimens a little under the surface of the steel bath for 1 or 2 minutes before they were lowered to 30 cm (1 ft) depth. The rather surprising result was that no penetration whatever took place unless the sand was very coarse (> 0.5 mm, A.F.S. No. 40). In the cases where no penetration occurred the test specimens were densely sintered on the surface and had a glazed appearance. As the temperature of the melt was usually not kept higher than 1550 C (2820 F) this strong sintering seemed rather surprising.

A strongly contributing reason for the occurrence of the dense surface layer would appear, however, to have been that fine slag particles eddied round in the molten metal and were deposited on the surface of the test specimen. Owing to the bath surface during the tests being kept almost bare, the oxidation and thus the slag formation was greater than usual. With very coarse sand the penetration was accelerated by pre-heating, but even in this case there occurred a considerable sintering effect with sparse penetration as

On trials with pre-heating for 2 minutes at a shallow depth, followed by immersion to 30 cm depth and immediately lifting there could be obtained a penetration depth of 3 to 4 mm (approx. 1/8 in.) with coarse sand. This shows that penetration can take place very rapidly if the sand is sufficiently heated in advance. With the same sand, however, twice as great penetration depth was obtained when a test piece was held for 2 minutes at a depth beneath the surface of 30 cm.

^{*} Experiments with pre-heating above the melt have been made on a large scale by Caine. 10,80

H

po

ly T

m

an

es

Ju

it

th

th

er

au

th

ha

th

or

in

ef

tie

10

ad

Sic

th

be

14. Effect of chemical reactions at the boundary between steel and sand—The chemical analysis of the sand which has been in contact with the steel during casting shows a rather high content of iron and manganese oxides. The following analysis for the dark zones of a nonadhering sand layer may be given as an example:

On computing the ratio Mn/Fe in the mold layer, it is found that this is about seven times as high as in the steel. The sand of a completely penetrated core was likewise found on analysis to have a rather high content of manganese. To enable the analysis to be made, the sand was leached out of a piece of core after treatment in a melt of sodium hydroxide and potassium carbonate. The penetrated steel was then left in the form of a spongy mass (Fig. 23).



Fig. 23—Spongy formation of steel obtained from a completely penetrated core from which the quartz has been removed.

As is shown by the microscopic examination the metallic oxides are comprised in the silicate in the form of fayalite. Caine's opinion that iron silicates are not present is without doubt incorrect.

The reason for the concentration of heavy metallic oxides in the contact zone must be that the steel is oxidized on the surface and the oxides then formed are absorbed by and react with the sand. This formation of slag on the boundary between the steel and the mold may, it is considered, substantially affect the penetration of the steel. This, however, can only apply to that part of the slag that is formed before the steel has definitely solidified on the surface and not to the slag formed while the casting is cooling.

Basing on the equilibrium conditions prevailing in the molten state in the system Fe-Mn-Si-O²¹ certain assumptions can be made concerning the slag which should be formed in a casting while the steel is molten. At the interface between the steel and the sand there is never any equilibrium, though the different components are striving all the time to attain one, and it is therefore possible to forecast qualita-

tively what will happen. Normally there prevails an oxidizing atmosphere during the casting of steel, and the alloying elements, mainly silicon and manganese, are consumed on the surface of the steel.

Diffusion takes place rapidly in the molten steel, however, and it can therefore be reckoned that where it contacts the mold there is a composition of the steel that does not depart too much from the average. A criterion for this is provided by the fact that the steel normally becomes free from blow holes. In fact, if the silicon content is appreciably lowered the steel will no longer be killed and bubbles of CO will form. If any appreciable difference from the average analysis arises in normal castings this must occur in a very thin surface layer where CO can go away without bubbles being formed.

Metal and Sand Conditions

With reference to the above equilibrium conditions the following cases may be considered as occurring between the melt and the sand:

a. The manganese content is not appreciably higher than the silicon content: Owing to oxidation a pure silicon dioxide slag is formed on the steel surface. The silicon content is then decreased in the steel.

b. The ratio Mn/Si is higher than in "a": Oxidation causes on the surface of the steel a fluid manganese silicate slag with low iron content. The silicon and manganese contents diminish in the steel.

c. Ratio Mn/Si relatively high as with "b": By reaction between Mn in the steel and SiO₂ in the sand there arises a fluid manganese silicate slag with low iron content. The manganese content in the steel decreases and the silicon content increases. If oxidizing conditions are present, case "b" should occur at the same time.

d. When the silicon content has decreased according to "a" a transition to case "b" can take place.

e. When the silicon and manganese contents in the bulk of the steel have gone down to abnormally low figures the Fe content of the slag may become considerable.

From the above it appears that one can hardly expect the FeO content of the slag to be large in relation to the MnO content as long as the steel is molten. Analysis of the mold layer lying closest to the steel surface during casting shows, it is true, that the ratio Mn/Fe is many times greater there than in the steel, but the MnO content nevertheless is insignificant as compared with the FeO content (cf. the example above).

Some attempts were made to study more closely the slag formation on contact between molten steel and sand. This was done both by allowing a fairly large quantity of steel to run through a cylindrical "sprue" and by having cylindrical cakes of sand float for 5 minutes on the top of a steel melt in the high frequency furnace. In both cases it was found that the amount of slag formed was insignificant compared with that formed after casting. Moreover the contents of FeO and MnO in the slag were of the same order of magnitude.

An analysis of the above-mentioned spongy mass of

AND

an

and

gan

eel.

ere

the

ver-

hat

In

the

will

age

in

ith-

ons

ing

gh-

ire

ce.

da-

an-

on

By

he

ith

he

If

ıld

ac-

ce.

in

lly

ne

X-

la-

n.

el

io

el.

as le

d

re

d

e

þ

steel from a core gave the following result. (For purposes of comparison, the standard analysis for this cast steel is given).

	C, %	Si, %	Mn, %
Steel from core:	0.05	0.41	0.07
Standard analysis:	0.20	0.45	0.65

One indication that these changes took place mainly in the solid state appears from the low C content. The cause of the comparatively high Si content is hard to determine. It may be due to a quantity of mechanically bound silica having come in with the analysis, but it may also be considered as a result of a certain reaction taking place between the manganese and the silicon while the steel was still molten. Judging from the structure of oxidized steel layers, it would also appear as if Si does not diffuse either to the surface to any great extent but is oxidized inside the steel forming slag particles.

Dietert¹¹ has shown that if, for example, a cylindrical pin of steel embedded in sand is heated to a temperature of about 900 C (1650 F) and fusing temperature there is obtained a reaction layer of the same type as that occurring in casting. Similar experiments, with the same result, have been made by the

author.

From the above observations it is clear that at least the main part of the slag products formed in casting have occurred by reactions in the solid state. That the manganese content in the slag products obtained is considerably higher in relation to the iron content than in the steel may be due to a certain amount of slag formation having occurred in the molten state or to Mn, at least at high temperatures, diffusing relatively rapidly even in the solid state.

The undoubtable conclusion that the greater part of the reaction products observed in the sand have been formed after the steel solidified on the surface indicates that the significance of slag formation has been exaggerated, at least in respect of its mechanical effects. The general opinion has been that slag formation might be capable of preventing penetration owing to the pores in the sand being closed.^{4, 15, 17}

Support for this was thought to be provided by the fact that slag formation is usually strongest in nonadhering sand layers. Instead, this may be explained by the air supply to the casting's surface being considerably reduced when penetration takes place. Owing to this it is almost only the protruding tips of the penetrated steel that are accessible to the oxidizing action of the air. When no penetration occurs, on the other hand, the air has access to the whole of the surface, especially if the sand is separated from the steel during cooling, because of shrinkage tension. In projecting parts of the mold and in internal cores, where the penetration most often takes place, the air supply is often hindered by the shape of the parts and the pressure arising with the shrinkage of the steel, which presses the sand tightly against the steel and at the same time facilitates sintering.

Notwithstanding the above, it should be remembered that penetration can occur after the surface has partially solidified and that a shell of steel formed momentarily may melt up later. In such cases enough slag formation may take place to act to some extent as mechanical hindrance to the penetration. No such action occurred in immersion tests. The slag so formed can only be a very small part of the final amount.

Nevertheless there still remains the possibility that the formation of a slag skin, even if it is thin, may strongly influence the course of penetration in other than purely mechanical ways. It is chiefly the surface tension and wetting tendency of the metal that may be expected to be influenced.

Discussion of Results

A. Comparison between immersion tests and conditions when casting—In the immersion tests the penetration was found to follow fairly simple laws and the results obtained could easily be reproduced with satisfactory accuracy. There are, however, certain exceptions, e.g., the experiments with electrically sintered magnesia, high content of dextrin, etc. In these cases it was possible to obtain unexpectedly deep penetration, but the results were capricious. The reason may have been splitting of sand due to gas evolution.

In casting tests it appeared that the mechanism of penetration normally prevailing in immersion tests is not the only one present in casting. This could also be ascertained in observations at foundries, especially in the casting of cast iron. Certain differences could, however, be recognized as a result of the eroding action of flowing metal, though in other cases this explanation is not sufficient. In the casting tests, unexpected results were obtained, chiefly with the sand mixtures containing reducing agents, especially aluminum, when instead of the favorable action obtained in immersion tests the opposite was most often the case. Even without the admixture of reducing substances a certain penetration would sometimes occur, which according to the immersion tests was not to be expected.

The immersion tests differ in many respects from conditions prevailing with casting. The author was not able to decide with certainty which difference is the most important but the more oxidizing conditions and the large mechanical stresses on the sand with casting and the absence of any appreciable erosion with immersion tests would seem to play a large part.

B. Factors which favor penetration—Of the factors having impelling effect on the steel in penetration the most important as shown by the tests is the force of pressure of the metal on the sand. At least after penetration has started, gas generation on contact between the sand and the steel and other processes which have a splitting effect on the sand may facilitate penetration. The penetration is favored besides by high temperature both in the molten metal and in the sand.

The pressure on the sand of the steel arises not only on account of the weight of the melt but also through the kinetic energy of the pouring metal. The kinetic energy of the metal flow is an important cause of adhering sand, mainly owing to erosion.

The pressure of the steel on the sand may also rise during solidification because of gas evolution in the melt. It is known that bubbles rarely occur solely because of high hydrogen content. If, however, a little bubble has been formed owing to steam formation around a detached sand grain or owing to CO evolution, then hydrogen gas can afterwards expand rapidly in it and cause an excess of pressure. It does not seem probable that gas pressure arising in this way can be any usual cause of adhering sand. Nevertheless it was found in the casting tests that the possibility exists. A condition is, however, that sufficient excess pressure can be produced before the steel has managed to form a dense shell along the whole surface of contact with the mold.

Shrinkage cavities, therefore, should not be dangerous, despite the fact that hydrogen gas may give rise to excess pressure in them. In steel foundries, hydrogen may be delivered to the steel from metallurgical processes, ladle linings and from moisture in the sand. If moisture enters the steel is also oxidized. If this occurs after the steel has been deoxidized one can suppose in extreme cases that CO bubbles occur on solidification, especially in the vicinity of the mold walls, where the steel is in contact with steam and air.

Splitting of the sand owing to intense gas evolution may be supposed to favor penetration of the steel. It would be particularly bad if the gas evolution takes place inside the grains so that these are burst. The unexpectedly large penetration obtained in immersion tests with sands of magnesia appears to be a result of this. On the other hand, gas evolution taking place on the surface of the grains need not be detrimental. A homogeneous gas pressure in the sand should rather be preventive of penetration. If, however, penetration has started, at least so far that the molten steel bulges in between some sand grains, then an intensive gas evolution, localized to the contact between the penetrating steel and the silica grain may through its splitting action facilitate the steel's further progress. It appears necessary, however, that the gas-forming materials do not have time to burn away from the sand before the steel makes contact with it. Such effects from gas evolution should be facilitated by loose ramming or sand grains being detached by erosion, which in turn may be facilitated by intensive

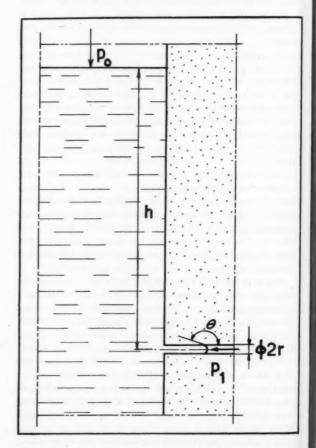
The extremely deep penetration that sometimes occurred on admixture of reducing substances in the mold walls when casting, might be interpreted as due to violent gas generation. Probably, however, the gas evolution is only a contributory cause, as also admixture of reducing agents producing little or no gas could have a bad effect, and as the penetration in immersion tests with equivalent sands was instead mostly counteracted.

C. Factors which counteract penetration — The capillary forces, mechanical obstacles in the sand and the solidification of the steel would appear to be the factors which act against the penetration of the steel. Gas pressure in the sand next to the surface of the molten metal should also be able to act as a hin-

drance, though this was not observed to be the case in the immersed sand cores.*

Interstices in the sand are small and the steel's surface tension is large, so that the capillary forces must assume decisive importance for the process of penetration. Metals do not wet oxide material, so that the capillary forces will be negative and counteract penetration. Provided the metal in contact with the sand is molten the relation between the pressure in the melt at the contact surface and the capillary force decide whether penetration shall occur or not. This is the simplest explanation of the fact that a certain minimum depth was found necessary for penetration and that this minimum depth is the greater the smaller the grain size of the sand is.

The magnitude of the capillary forces is determined by the size and shape of the sand pores and of the surface tension and wetting tendency of the steel. The sand pores are determined in the first place by the size, shape, distribution and packing of the grains, but on contact with the melt thermic and mechanical forces may influence the internal relationship of the grains, so that the size and shape of the pores are



th

fo

m

of

01

ti

gı

di

m

fe

T

m

m

en

an

sa

Fig. 24-Penetration of a molten metal in a capillary.

The preventive effect of reducing substances was independent of their gas contents. Different amounts of water in the sand had little influence on the results.

altered. To begin with, however, it will be assumed that the positions of the grains in the sand are fixed. If we consider an imagined cylindrical pore, Fig. 24, with radius r in the sand at a depth h beneath the surface of a homogeneous melt with a density 3 and surface of a homogeneous melt with a density f and assume the gas pressure over free metal surface and in the pore = p. and p1 respectively, the steel's surface tension $= \gamma$ and the angle of incidence between the molten metal and the sand $= \theta$ and designate acceleration due to gravity by g, then the total impelling force on the melt in the pore will be:

 $K = (h \cdot f \cdot g + p_0 - p_1) \cdot \pi r^2 + 2 \gamma \cdot \cos \theta \cdot \pi \cdot r;$ or per unit of area in the cross section:

$$k = h \cdot f \cdot g - \triangle p + \frac{2 \gamma \cdot \cos \theta}{r}$$

where \triangle p therefore represents the excess of gas pressure in the sand.

As the melt does not wet the contact angle is between 90 and 180 deg, i.e., $\cos \theta$ is negative.

The smallest capillary into which the melt can penetrate is therefore determined by:

Thus a low surface tension, a small angle of incidence between melt and sand (cos θ nearly = 0) and a high pressure mean that the melt can penetrate into fine pores.

In the immersion tests the sand was often submerged 30 cm (approx. 1 ft) below the surface of the melt. If this value is inserted in the above equations and if it be assumed that $\gamma = 1200$ dynes cm, $\cos \theta = -0.9$ and $\triangle p = 0$ then we get a value for r = 0.010 cm, i.e., a pore must be at least 0.20 mm in diameter for penetration to take place. It is of interest that this capillary diameter is of the same order of magnitude as the smallest penetrating steel tips. How the capillary pressure varies with the capillary radius in the above constant conditions is shown graphically by Fig. 25. As an example a line for constant external pressure has been inserted (the difference between the pressure of the melt and the gas pressure prevailing in the sand). The dotted area, therefore, marks the resulting pressure k, which has an impelling effect on the melt in the capillary.

The graph provides a simple explanation why a minimum depth shown is required for penetration. It also shows how the impelling force increases with increasing grain size. The grain size in a sand, however, varies and this applies also to the interstices present in the sand. With low metal pressure or fine grain the capillary resistance is exceeded in only a few pores and sparse penetration should be the result. The dispersion of pore size in a given sand has been marked with dotted lines in Fig. 25 as an example.

Even though the capillary forces decide if the molten metal can penetrate into a given interstice, a couple of important reservations must nevertheless be made against the method of consideration hitherto employed. It is not certain that the capillary forces are constant. While the melt is in contact with the sand there can occur chemical changes both in the

melt and in the sand which influence the surface tension of the metal and the wetting between the melt and the sand. It is thus possible that the metal may gradually penetrate into an interstice which was originally too small. Moreover, regular pores do not occur as was previously assumed but the interstices form a network of pores which thicken and thin alternately. Obviously the penetrating metal finds its way forward where the openings are largest, but it must often happen that the way is blocked in all directions for the advancing metal vein. As stated earlier, the steel has a certain ability to break through such obstacles and enlarge the pores.

An important question is what limits the rate of penetration. If it were ordinary resistance to flow the rate would quickly decrease with depth of penetration. In the constant conditions prevailing with immersion tests, however, the rate was found to be fairly independent of depth of penetration. The simplest explanation of this is that practically the whole resistance to penetration throughout the time is concentrated to the tips of the penetrating veins. This means among other things that the viscosity of the molten metal behind the tips does not influence the rate of penetration, a condition that is quite natural as the viscosity of molten metals is extremely low. (An example of this is provided by the fact that a wash if it is not entirely tight does not decrease penetration, even if the break through takes place

only at a few spots.)

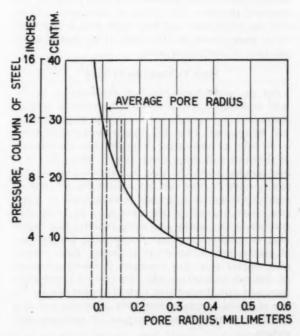


Fig. 25-Variation of capillary force with capillary radius assuming that the surface tension of the molten steel is 1200 dynes per cm and that cosine for the contact angle is -0.9. For comparison a line for a constant counter pressure of 30 cm steel column has been inserted. The pore sizes range and average size of one sand are also indicated.

for the ined

SAND

eel's

orces

s of

1, 50

nter-

with

sure

llary

not.

at a

the teel. e by ins, iical the

are

my.

ndthe

8

ol

bl

af

ex

th

th

re

tra

th

th

ra

tir

als

op

cre

ab

me

bee

cus

for

car

Wit

It seems most obvious therefore to assume that the resistance is due to beginning of solidification and that the rate of penetration would be limited by the sand in the penetration front just having such a temperature that the leading tips of the steel do not solidify (or rather alternately begin to solidify and re-melt as the pores contract and expand). The rate of penetration would then be determined by the speed at which the necessary temperature moves along in the sand. The similarity between the curves for the relation between immersion time and depth of penetration for the steel (Fig. 16) and the "penetration depth" for the sintering front for relatively easily penetrated sands (Fig. 17) also support this assumption, as also the strong dependence on temperature of the rate of penetration.

Nevertheless there are several reasons which are against the suggestion that the rate of penetration is limited in that way. For example, the rate of penetration is dependent to a very great extent on the pressure of the melt. This objection might be met by saying that the heat transmission through the steel that penetrates is entirely dominant over the heat transmission through the sand. That the heat transmission through the steel is considerable is supported, among other things, by the curves in Fig. 22 a and b. It is not, however, certain that increase of pressure will increase the heat conduction. Assuming that the pressure suddenly increases then the steel should. while giving off heat, penetrate so far in that the resistance due to solidification will be equal to the increased pressure. The difference afterwards will be that the penetration can take place with a somewhat lower temperature in the sand. This should cause the rate to increase a little.

Heat Transmission to Sand

On the other hand the heat transmission to the sand in the penetration front will hardly be greater. Again, at low pressures, the steel's rate of penetration will be so small (Fig. 18) that a temperature front in the sand corresponding to the liquidus temperature of the steel will move more quickly (Fig. 22). It is not possible therefore that a beginning of solidification could in this case set a limit for the rate of penetration and as a pressure increase leads to a proportionate increase of the rate it is probable that another factor, which acts more rapidly than beginning of solidification, was determining for the rate of penetration in the immersion tests. Other reasons also support this. For example, it would otherwise be difficult to explain why the rate of penetration increased with diminishing carbon content despite the rise of liquidus temperature then occurring and that the rate decreased on admixture of reducing substances.

If instead it be assumed that it is the capillary forces that limit the rate then the last-named facts will be understandable. One will then however have to be content with stating that the capillary resistance will be smaller with low carbon content than with high, as no measurements of this have been made. Such measurements are moreover difficult to make.

Determinations of the surface tension of cast iron and mild steel or pure iron in vacuum or neutral atmophere have as a rule given lower values for cast iron than for steel and pure iron.

In the immersion tests, however, the atmosphere was normally slightly oxidizing. This should reduce the surface tension and perhaps above all increase the wetting tendency by the formation of a slag skin. The decrease of capillary resistance then occurring will be greater with steel of low carbon content than with cast iron, as the latter because of its high carbon content forms its own protective atmosphere of carbon monoxide.

The counteracting effect on penetration which admixture of reducing substances in the sand had in immersion tests may also be explained by decreasing of oxidation by these substances. That the tendency to penetration was greater in casting than in immersion tests may be *partly* explained by the oxidation being stronger in casting. On the other hand it is difficult to explain why admixture of MnO₂ or Fe₂O₃ did not increase the rate of penetration. One reason may be that the increase of oxygen content with such admixture is too small to have an effect, another that a further increase in oxygen has no effect.

It may also be thought the slag formation and sintering which the admixture caused acted in opposition to increase of oxidation. Another circumstance that is difficult to explain as the cause of capillary forces alone is that the rate of penetration will be much greater if the sand is pre-heated. If the sand has been pre-heated so that the surface temperature is higher than the liquidus temperature of the steel the penetration seems able to take place almost instantaneously to a certain depth.

Even if the capillary forces have a great influence on the rate of penetration, therefore, it is nevertheless theoretically impossible that these alone could limit the rate as long as the surface tension, the wetting properties and pore size are constant. If the impelling force is sufficiently great for penetration to take place at all the capillary forces cannot arrest an increase of speed in a pore with constant section.

Preheating of Sand

The effect of pre-heating of the sand, the fact that penetration increases rapidly with increasing temperature in the melt and the above-stated similarity between the steel's penetration and the progress of sintering show that the temperature is one of the most important factors regulating the rate, even if the limitation is not caused by commencing solidification which should, however, be the case with high metal pressure (great rate of penetration). The factor, or one of the factors that limit the rate of penetration must in any case be governed by temperature.

All changes which contribute to widen the paths for the steel must facilitate penetration. On the other hand, the rate of penetration must be reduced if the resistance to change as regards position and shape of the grains is great. Such changes are most easily observed when the metal pressure has been

so low in relation to the grain size of the sand that penetration of the original pores cannot take place. In such cases, the presence of groups of sand sintered together between the metal veins (Fig. 12) may be taken as indicating that the sintering was primary and that the penetration could not take place until larger pores had been produced through sintering. Probably however, the penetration starts by the melt bulging into the larger pores and so owing to its pressure, pushes the sand grains aside. By cooperation between the pressure of the melt and softening of the grains and reduced coherence between these owing to high temperature it is then possible for the penetration to occur.

Though such changes of position and shape in the grains were not always so easy to observe when the metal pressure was high or the sand was coarse, yet there is an important reason which suggests that even then they play a big part. Even if the average crosssection of the pores is quite sufficient for the capillary forces to be overcome, yet the melt must on account of the irregular shape of the pores be repeatedly blocked during penetration by constrictions which are either too small or which can be broken through after a certain delay. In this way it is possible to explain the effect of the pressure of the melt and the temperature. In a uniformly rammed sand, these checks should occur at intervals that are fairly regular, and it would be natural for the rate of penetration to be constant for constant pressure and temperature.

Resistance to Penetration

From the rectilinear relation between the pressure of the melt and the rate of penetration it appears that the resistance to penetration is proportional to the rate. This circumstance seems to be connected with the number of checks being proportional to the rate.

Increasing wetting tendency in the melt with the time of contact will in principle have the same effect as enlargement of the constrictions, as in this case also the steel can gradually penetrate through an opening which to begin with is too small. Such increase in wetting tendency may arise from several causes. The fact that the steel fills surprisingly well even small irregularities in the pores in which it penetrates suggests that a decrease of capillary resistance really does take place. The fact that changes of position and shape in the grains generally occur, as also the great effect of temperature make it probable, however, that enlargement of the pores is the more important of the two.

It is conceivable that some important factor has been neglected, but the factors that have been discussed are sufficient to enable a theory to be outlined for the mechanism of penetration.

D. Mechanism of steel penetration in the sand—1. General observations: In order that the penetration can take place it is necessary that the steel in contact with the sand is at least partially molten. Owing to the capillary forces there is required further a

certain size in the openings through which the steel is to penetrate. These openings must be comparatively large with low metal pressure, high surface tension and no wetting tendency between the melt and the sand (cosine for angle of incidence almost = 1). If there are not enough openings at the beginning these can, however, be formed under the action of the steels' pressure, of the temperature, of splitting forces such as intensive gas evolution formed locally or through the eroding effects of the melt. Moreover, openings which were too small at the beginning will nevertheless be sufficient after a certain time if the surface tension of the melt or the contact angle between melt and sand decrease with the contact time. Once the opening is so large that sufficient bulging in of the melt can happen the penetration can proceed through joint action of the melt's pressure and one or more of the above factors.

Rate of Penetration

The rate of penetration is determined in the first place by the pressure the melt exercises on the sand and by the mechanical resistance meeting the leading metal tips in the penetration front. With great rate of penetration the rate may be limited by solidification beginning to set in at the tips, which also happens if the delivery of heat decreases or is interrupted. As long as penetration is proceeding it would seem that normally even the leading metal tips are quite molten. The mechanical resistance is then determined by the number of openings per unit of length in the direction of penetration which the melt can only penetrate after a certain pause and of how long these pauses are. The number of openings that check the melt is dependent on the metal pressure, the sand grain size and shape and on the capillary forces, while the time taken on the average to force a way through these barriers is dependent on the sand's temperature in the penetration front, the steel pressure, the presence of splitting forces etc., on the resistance that meets the intervention of these factors and on possible decrease of the capillary forces with the time of con-

2. The course of penetration in castings-(a) Normal course: When the steel is rising in the mold it will probably first solidify over the whole surface of the mold. If the steel temperature is sufficiently high, its mass large enough in relation to adjoining parts of the mold or if the amount of molten metal that runs over a given part is large enough, then parts more or less largely of the steel shell first formed may re-melt. If the pressure of the melt on the sand has in the meantime risen sufficiently then penetration can start. If the time elapsing before penetration can start is long so that the sand is able to get much preheated the first penetration could take place quickly. This requires, among other things, that the casting is very large, so that the time of pouring will be long. Further it is required that the steel shell first formed re-melts very fast, as it is only after this re-melting that the sand can be heated to a temperature in the neighborhood of or higher than the liquidus temperature. The depth of penetration in a given part of

add in sing ency im-

oxi-

SAND

and

mos-

iron

here

duce

the.

The

ll be

with

rbon

bon

or One tent fect,

opumof ion the

em-

of ace nce less mit ing

ing ace ase

em-

itv

of ost the ion tal or

ths the ed

on

ost

the

HSI

sin

ob

cas

cre

Size

ab

wit

the mold will then be determined by the structure of the sand, by the magnitude of the metal pressure and by the time it takes for the penetrating steel tips

to begin solidification.

(b). Other types of penetration. As stated earlier in casting penetration occurs which does not follow the "normal" course as outlined above. Such penetration occurs particularly when organic or other reducing substances are mixed in the sand. A characteristic feature of such penetration is that it may happen with very low metal pressure and with small metal mass (small castings) and still may be very deep, i.e., the penetration rate may be very high.

Mechanism for Penetration

The mechanism for such more incalculable penetration has not yet been possible to explain fully. The same laws must apply in this case as with the "normal" penetration, but some factor which before has been of subordinate significance may have become dominating. Different factors may dominate in different cases. On the basis of observation and assumptions the following types of "incalculable penetration" may be distinguished:

(x) Penetration may be caused by a high gas pressure arising in the metal during solidification.

(B) In conjunction with erosion, penetration may take place in such a way that the sand grains are set in motion, so that sufficient interstices arise at certain spots to enable the melt to penetrate as soon as sufficiently large openings are formed. Movement of the sand grains could be facilitated by gas generation.

(γ) Sometimes splitting of large or small groups of grains may occur, enabling the melt to penetrate. This represents a transitory form between scabbing

and ordinary penetration.

(f) Fin-formation or veining arises when the melt fills up cracks in the sand. After a start of a crack has appeared the melt seems able to exercise a wedge action. Veining occurs mostly when the sand is loosely rammed. It would be favored by the sand being easy to compress and possessing low tensile strength.

The fact that the admixture of organic or other reducing substances can increase the tendency to pene-

tration may have several causes:

Admixture of such substances can, at least with high contents, deteriorate the power of the sand to resist erosion and penetration, both by decreased coherence and owing to splitting effects from gas evolution.

The hydrogen-containing gases generated by the organic substances (and by moisture) in the sand may be expected to involve a gas pressure in the steel on solidification.

Owing to partial combustion the temperature may rise in the sand while at the same time the content of admixed substances decreases. This combustion would not be expected to be great on the admixture of organic substances, as the abundant generation of reducing gases should reduce the oxygen supply to a great extent. On the other hand, oxidation of, e.g., aluminum may be supposed to be considerable. On gasification of the organic substances packing of the sand may be looser. Also the addition of cereals will diminish the flowability of the sand, which will contribute to make the packing loose.

Literature Cited

1. A. H. Dierker, "Reclaiming Steel-foundry Sands," Trans.,

A.I.M.M.E., vol. 90, pp. 83-99 (1930).
2. P. L. Goodale, "Notes on Behavior of Sand Mold in Steel Foundries," Transactions, A.F.A., vol. 38, pp. 471-480 (1930). 3. H. W. Dietert, Contribution to discussion on "Mold At-

- mosphere Control," Transactions, A.F.A., vol. 52, p. 1074 (1944). W. Vaska, "Feuerbeständigkeit und Bewertung von Formstoffen," Die Giesserei, vol. 29, no. 18, pp. 311-313 (1942).
- 5. J. B. Caine, "Mold and Core Washes," Steel Foundry Facts. Dec. 1941, pp. 2-8.

6. J. B. Caine, "A Study of Burnt-on or Adhering Sand." Transactions, A.F.A., vol. 51, 1-p. 647-705 (1943)

7. Foundry Sand Research Committee, "Report of the Subcommittee on Sintering Test," TRANSACTIONS, A.F.A., vol. 52, pp. 1311-1316 (1944).

8. E. E. Woodliff, "Metal Penetration in the Mold," AMERI-CAN FOUNDRYMAN, Nov. 1942, pp. 6-8.

9. S. G. Jones, "Does Metal Vapor Cause Sand Penetration?"

AMERICAN FOUNDRYMAN, Apr. 1948, p. 139.

10. P. R. Beeley and H. T. Protheroe, "The Surface Finish of Steel Castings." Unpublished report from BISRA Steel Cast-

ings Division 1950, also preliminary report of

T. Bishop and K. G. Lewis, "Steel Foundry Research."

Foundry Trade Journal, vol. 86, no. 1693, pp. 111-113 (1949).

11. H. W. Dietert, R. L. Doelman and R. W. Bennett, "Mold Atmosphere Control," Transactions, A.F.A., vol. 52, pp. 1053-

1071 (1944).

12. Mold Surface Committee of A.F.A., "Changes in Chemistry of Liquid Steel in Contact with Sand," Transactions, A. F.A., vol. 56, pp. 260-262 (1948). 13. R. E. Savage and H. F. Taylor, "Fayalite Reactions in

Sand Molds Used for Making Steel Castings," Transactions,

A.F.S., vol. 58, pp. 564-577 (1950).

14. A. M. Portevin and P. G. Bastien, "Capillarity as a Factor in Foundry Practice," The Inst. of British Foundrymen, 33rd Annual Conference, Paper No. 582, Glasgow 1936, pp. 88-116.

15. T. P. Hoar and D. V. Atterton, "Penetration of Molten Metal into Compacted Sand," J. Iron Steel Inst. 166 (1950) September, pp. 1-17.

16. S. L. Gertsman and A. E. Murton, "An Investigation of

Metal Penetration in Steel Sand Cores," TRANSACTIONS, A.F.S., vol. 58, pp. 595-603 (1950). 17. S. Hj-son Ljungren, "The Sand Problem in Steel Cast-

ing." Swedish Ironmasters Assn. Annals, vol. 129, no. 8, pp. 457-530 (1945).

18. N. A. Berglund, "Casting of Steel in Green Sand." Swedish Ironmasters' Ass'n Annals, vol. 134, no. 2, pp. 41-83 (1950)

19. J. B. Caine, "Behavior of Molding Sand in Contact with Liquid Steel," The Foundry, vol 75, no. 7, pp. 72-77, 143, 146. 149, 153, 155, 156, 158 (1947).

20. J. B. Caine, "Sand in the Mill vs Sand in the Mold,"
The Foundry, vol. 77, no. 10, pp. 68, 70, 232-240 (1949).
21. F. Körber, W. Oelsen, "Die Grundlagen der Desoxydation

mit Mangan und Silizium," Mitt. Kais. Wilh. Inst. Eisenforschg. vol. 15, pp. 271-309 (1933).

DISCUSSION

Chairman: J. A. RASSENFOSS,* American Steel Foundries, East Chicago, Ind.

Co-Chairman: H. W. DIETERT, Harry W. Dietert Co., Detroit. S. L. GERTSMAN and A. E. MURTON: 1 The author made an excellent contribution to our fundamental knowledge of metal penetration in steel sand cores. The method described appears to be a useful extension of the Caine hot dip test.

We feel that there should be some clarification of the term "capillary action" or "capillarity." The author has pointed out that with steel in contact with sand, capillary action acts in a negative manner. This introduces a terminology which is really based on

² Foundry Consultant, Wyoming, Ohio.

This paper was presented by Mr. Rassenfoss in the absence of the author. Dept. of Mines and Technical Surveys, Ottawa, Ont., Canada.

AND

will

con-

ans.

Steel

30).

At-

944).

orm-

acts.

nd."

Sub-

52.

ERIm?" nen astch." 9). old)53em-A. in DNS, tor 3rd 16. ten 50) of .S., st-57. ed-0) ith 46. 1, " on ig.

es, oit. an tal

ed ets is 1. The non-wetting characteristic of steel on sand;

2. The surface tension of steel;

3. The effect of pore size of the sand.

Since the surface tension of a particular steel, such as mild steel, does not appear to vary to any great extent, capillarity is then predominantly a function of effective pore size.

Hoar and Atterton have come to the same conclusion by using the same basic equation as the present author. The former similarly conclude that the penetrating pressure required to obtain metal penetration is related to the effective pore radius or the size of the largest pores.

Our empirical work, using the tentative standard A.F.S. test casting, has led us to the same conclusion. We have had to increase the height of the casting whenever we decreased the grain size of the sand in order to obtain metal penetration.

J. B. CAINE: There seems to be a misunderstanding, probably due to translation of the paper. Mr. Pettersson disagrees with my findings regarding the influence of silicates. His basis is the chemical analysis of sand in contact with molten steel after the steel has solidified. The question is, did the oxidation and formation of silicates occur before, or after initial skin

formation. Scattered, but not conclusive evidence indicates that this oxidation of the sand occurs after initial solidification, and if true, would then have no bearing on penetration.

MR. Pettersson (Author's Closure): Though the different physical properties behind the term "capillary action" have been treated in some detail in my paper, an introduction like Messrs. Gertsman and Murton present might have been valuable. It should be pointed out, however, that the contact angle between molten steel and sand, and also the surface tension of the steel are affected by the atmosphere and eventual chemical reactions tt the steel-sand interface.

I am not sure what misunderstandings Mr. Caine has in view. My attempts to study the effect of the chemical reactions have led me to the conclusion that the oxidation of the steel and the following formation of silicates generally take place almost exclusively after the steel has solidified and so cannot have a great effect on the pore size of the sand in contact with molten steel. As stated above, however, though the quantity of the products of chemical reaction generally will be small as long as the steel is molten at the surface, they may have an appreciable effect on the surface forces acting between the sand and the liquid steel.

MECHANICAL EQUIPMENT FOR MEDIUM - SIZE GRAY IRON FOUNDRY

By

H. W. Zimnawoda*

ABSTRACT

Mechanical equipment for production and material handling representing practice of progressive medium size jobbing gray iron foundries is discussed in three sections dealing with core room, molding and melting. Most of the described units are illustrated. The importance of proper layout of the selected units is emphasized and the typical sand, mold and core handling arrangements are shown on sketches.

THE PRESENT SHORTAGE OF CASTINGS makes imperative the utilization of mechanical equipment to increase the output of existing foundries. The diminishing supply of skilled labor requires that machines be used to the greatest possible degree.

Every foundry may be characterized by the figure which indicates its production of salable casting per man-hour.

The productivity will increase when the performance of skilled operators will not be delayed by unproductive functions. Improved methods should be offered to ease manual effort and provide for better working conditions. The heavy operations should be handled by horsepower instead of manpower.

The secret of success in each foundry section lies in the reduction of variables to a minimum. This can be accomplished by controlled sand mixture preparation for cores and molds, reliable core baking ovens, controlled melting methods, and adequate hot metal handling system.

The gray iron jobbing foundries melting 20 to 30 tons of iron per day represent a large segment of the foundry industry in need of modernization of their production methods. Except in cases involving specialized work, most of these foundries employing hand methods of operation have found it difficult, if not impossible, to compete against mechanized shops.

If the foundry cannot afford to install a mechanized system at once, it should proceed gradually. The savings obtained in the first period of modernization should be used to finance the next step of the predetermined program. A long-range view of the final

stage of the mechanized system should be kept in mind so that the step-by-step modernization assures that the preceding step fits the following one, no matter how much time elapses between the installation of each.

A careful study of existing methods of a particular foundry and a complete analysis of the requirements connected with the anticipated production program should result in a proper layout of the selected equipment.

The purpose of this paper is to indicate the essential types of equipment used in the core room, the molding and melting departments, and to show some typical arrangements and layouts. This survey may bring nothing new to the foundrymen already enjoying the benefits of mechanization, but it may serve as a guide for those who are now ready for modernization of their shops. Because of restriction in the scope of this paper, the vast field of molding machines, core making and casting cleaning equipment was omitted. After establishment of a general layout and production capacity, these machines can be selected from catalogs according to the type of castings to be produced. The outstanding features of each make has to be compared in the light of performance, initial cost and maintenance requirement.

Core Room

Economy of coreroom operation depends upon location in relation to other departments, selection of proper equipment and arrangements of units according to the flow of materials or sequence of operations. Mechanical equipment, supplemented when necessary by unskilled labor, will relieve the skilled coremakers of time-consuming and unproductive activities. Mechanical handling of finished cores will save time and labor, and ease of movement will eliminate distortion and breakage of cores.

It is desirable that the grain size of core sand mixture be comparable to that of the system sand mixture so that cores broken up in shake-out operation will not appreciably affect the permeability of prepared molding sand mixture. This sand mixture should also be free from moisture, and if delivered wet should be dried in an oven.

There are three principal designs of dryers: one

H.

of is

dis

ing

Ir

bu

he

thi

of

the

int

CEL

sto

Th

sha

Th

cha

Mu

Th

^{*} National Engineering Company, Chicago.

Presented at the 24th Foundry Congress, in Paris, June 5, 1951, as the Official Exchange Paper of the American Foundrymen's Society to the French Technical Foundry Association.

ind

the

low

lar

nts

am

uip-

en-

the

me

nay

oy-

as

on

of

ore

ed.

on

ogs

he

ed

in-

of of of on s. ry ers fend on

ot dbe

ne

48

with a parallel gas and sand flow, another with a counter flow of gas and sand, and a through flow type of dryer. The initial cost of the first and second type is lower but they are somewhat less efficient because of considerable heat loss (for this reason they will not be discussed). The principle of the third type is shown in Fig. 1. This type of dryer possesses the advantages

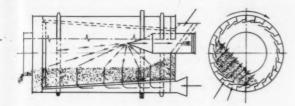


Fig. 1-Sketch illustrating sand dryer.

of both counter-current and parallel-current flow drying methods, but without the disadvantages of either. It provides not only high efficiency and rapid drying, but a uniform treatment without danger of overheating. Its design allows the hot gases to pass through the channels covered by overlapping plates, then through the sand into the inside of the drum and out of the exhaust duct. The dryer utilizes practically the total heat from the hot gases which comes into intimate contact with all particles of sand. In this type of dryer the moisture can be reduced to 11/2 per cent. It is recommended to cool the sand before storage.

Mixers for core sand fall into two groups: (1) blade (paddle or sigma-type); and (2) muller-type mixers. The first type consists either of a trough with revolving shaft and mounted paddles or two sigma-shaped shafts. These mixers are used for small batches and are charged by hand; they are effective for oil binders. Muller-type mixers are built for larger capacities. They provide better distribution of ingredients, espe-

cially colloidal type binders, and develop better green strength of prepared sand. These mixers consist of a circular crib with two revolving mullers, and usually are equipped with bucket loaders for charging.

The variety of sand mixtures requires individual study for proper handling. The distribution of prepared sand should be frequent and in small quantities. Because many sand mixtures are unstable, storage of them should be limited to amounts to be used in a short period of time.

Manual handling of prepared sand can be replaced by the installation of the mixer on a platform. In this arrangement the dry sand is shoveled from the storage into a bucket of the loader, where the bonding additions are made. The prepared sand is distributed on the platform by dumping the contents from a rubberwheeled buggy into the coremaker's hoppers suspended from the platform. It can be also discharged into a holding hopper for use by floor coremakers or for distribution to small core blowers.

Another method of core sand distribution is by monorail carrier, which stops in front of the mixer, lowering a bottom-drop bucket under the discharge chute. When filled the bucket is hoisted to the transporting position and stopped at the coremaker's stations to refill the hoppers.

Rapid core production with core blowing machines is well known. When considering this type of equipment, the additional expense for metal core boxes and usually special core dryers must be kept in mind. The sand mixtures used should have a low moisture content to prevent sticking. The air supply must be at a constant, usually 80 psi for simple cores and 100 to 120 psi for intricate cores. Modern core blowers are very fast. One movement of the air valve handle controls the air, automatically performing the operation of clamping, filling the core box and releasing it after the core is made. The selection of proper type and size will depend upon the shape of cores and their

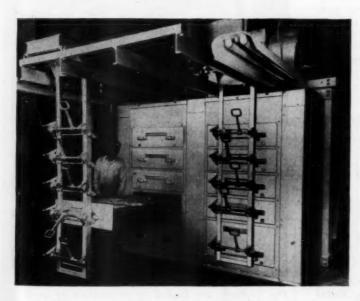


Fig. 2-Drawer-type oven for baking small cores.

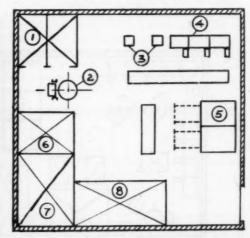


Fig. 3—Coreroom layout: (1) sand storage; (2) mixer and bucket loader; (3) coreblowers; (4) core benches; (5) drawer-type ovens; (6) core inspection; (7) core box storage; (8) finished core storage.

amount. The core machines of rollover type are especially suitable where a deep draw of the core box requires considerable skill of the operator.

After the selection of proper sand mixing and core making equipment, suitable core baking ovens have to be considered. The most important features of the modern core ovens is the uniformity of the temperature inside of the oven, its automatic control and low fuel consumption.

The time and temperature of baking process depend upon the grain size of core sand, moisture content, type of binder used, humidity of the atmosphere and size of cores. The cores made from coarse sand will reach maximum strength in a shorter baking time than those made from finer sand. The moisture, usually up to 6 per cent has to be evaporated and exhausted from the oven. The baking time is in straight proportion to the amount of moisture. Some of the binders will produce best results when heated longer, perhaps 3 to 4 hr at a moderate temperature of 300 to 350 F; others will require short baking time of 1 to 2 hr at a temperature 450 to 500 F. The core oils, which undergo chemical reactions of oxidation and polymerization, in order to form a hard film and hold the grains together, require a supply of fresh air to be introduced into the oven. The baking time is in direct relation to the size of cores. At the average baking temperature of 400 F, the heat penetration may be figured at approximately 1 in. per hr.

All these factors must be considered in selecting the proper type of oven and determining baking temperature and time. Exact analysis of the production will enable any foundry to obtain cores with required strength, made from a sand with a minimum of binder, and baked in the oven at a proper temperature in the shortest time.

For the description of modern core ovens we have selected representatives of three main groups: drawer-type, rack-type, and vertical conveyor-type ovens. These ovens are built with recirculating systems for uniform temperature distribution and baking with low temperature differential. The fuels used are gen-

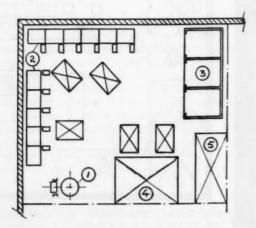


Fig. 4—Coreroom layout (1) mixer and bucket loader; (2) core benches; (3) rack-type ovens; (4) core finishing; (5) core storage.

erally oil, gas or electricity and occasionally coal, any of which can be accurately controlled by means of modern automatic temperature control. Coke is not an "automatic" fuel and is therefore not used in modern recirculating ovens.

The heat is developed in a combustion chamber located outside of the oven and the gases produced in this chapter are mixed with recirculated air from the oven with the addition of such fresh air as is necessary. The resulting mixture is then delivered by means of the recirculating air fan into the oven where it is distributed through ducts to provide uniform temperature throughout the oven.

Drawer-type ovens are used for baking small cores. Within the oven the drawers are supported by rollers which assure a smooth movement without danger of breaking small, intricate cores. The separate baking areas permit selective baking time and handling of rush jobs. With this type of oven the coremaker's benches or core blowers should be grouped close to the oven. The core plates with green cores are deposited by the coremakers on set-off tables. The oven tender transfers the cores to the ovens and, after baking, deposits the cores on the table located in front of the ovens. There they are inspected and delivered to the molding stations. See Figs. 2 and 3.

Rack-type ovens were developed to reduce the cost of transportation of larger cores by the use of portable racks of the average size 4x6 ft (usually with 5 shelves) and a capacity of 1 to 2 tons. The racks are loaded by coremakers at their working stations, and moved by hand- or power-operated trucks into the oven. See Fig. 4 and 5.

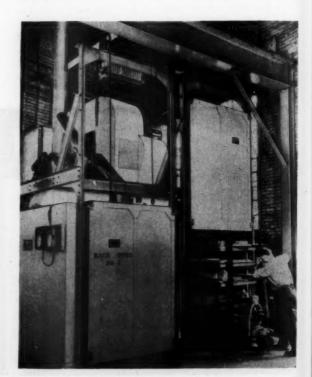


Fig. 5-Racks are loaded by coremakers at working stations and moved by truck into rack-type oven.

DRY

anv

of

not

od-

lo-

in

the

ces-

ans

t is

era-

res.

lers

of

ing

of

er's

the

ted

der

de-

the

the

Ost

ble

es)

by

by

See

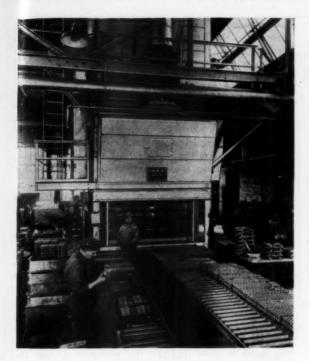


Fig. 6-Vertical conveyor-type oven commonly used for baking cores of approximately the same size.

Vertical conveyor type ovens are commonly used in foundries where the manufacturing program is established and the cores are of approximately the same size. See Fig. 6. It is not unusual, however, for vertical conveyor ovens to be used in jobbing shops where a variety of core shapes and sizes must be baked. In such cases the occasional large or chunky cores may be permitted to remain on the conveyor for two or more trips through the vertical oven, while the smaller cores are removed after one trip. Often the coreroom operating vertical ovens has other types of ovens for handling odd sizes of cores.

Among the major advantages of the vertical core oven are its continuous operation and the minimum floor space required. The bench coremakers and the core blower operators usually are grouped around the upgoing section of the conveyor. A passageway through the middle of the oven provides for convenient access to both sides of the ascending shelves, eliminating congestion at the loading station.

The complete cycle is divided into the baking period and the cooling period. Usually 25 per cent of the conveyor carriers are in the cooling zone. The combustion chamber, heated by oil or gas burners, is located in the central part of the oven. The hot gases, tempered by recirculated air are forced into the heat distributing chamber and across the ascending carriers. In the cooling period a forced-air system provides for the cooling of the cores and the complete removal of smoke before the cores reach the unloading station.

The conveyor movement can be continuous or intermittent in operation. Continuous conveyor travel is almost universally used, but for heavy classes of cores, which require definite periods of time for loading and unloading, conveyor travel may be intermittent. Upon reaching the unloading station, the cores are removed from the oven and placed on gravity roller conveyors as shown in Fig. 6.

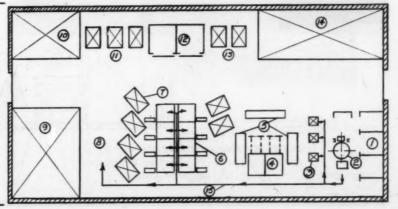
In determining the type and capacity of core oven to be used, the following factors must be considered: number and size of core plates per day or per hour; maximum height, weight and composition of the cores; and the approximate consumption of core sand per day.

Usually at least two types of core ovens are required to cover the entire production of a medium-size gray iron foundry. The accompanying layout (Fig. 7) shows the arrangement with drawer-type and rack-type ovens. The sand prepared in a mixer equipped with bucket loader is delivered to the coremakers' stations by a monorail system or by carts pushed on the floor. Coremakers producing small cores are grouped around the drawer-type ovens. Medium-size cores are made on benches and loaded on racks to be delivered to rack-type ovens. Space is reserved for core finishing, inspection and storage, as well as for storage of sand and core boxes.

Molding Section

Handling of used and prepared sand and molds before and after pouring are the most important functions contributing to efficient operation of the foundry. The right selection of proper mechanical units and layout will reduce the total volume of sand needed

Fig. 7—Coreroom layout: (1) sand storage; (2) sand preparing unit; (3) core blowers; (4) drawer-type ovens; (5) tables; (6) core benches; (7) core racks; (8) space for special cores; (9) core storage; (10) finishing and inspection; (11) cooling; (12) rack-type ovens; (13) empty racks; (14) core box storage; (15) prepared sand delivery to blowers and benches.



and the quantity of molding equipment. Proper layout will also allow for increase of production with no increase in building space. It will reduce the number of scrap castings and provide for better working conditions.

The first step in a modernization program is to employ a molding sand preparation method which will insure uniformity of each batch of prepared sand mixture. The next stage should deal with the mechanical conveying system for shake-out sand, its cleaning and storage, and the last modernization change should consider the equipment used to distribute prepared sand mixture and handling molds.

The chief advantage in mulling of sand lies in the close control of the required properties of the prepared sand mixture. Its high qualities are developed by the intimate contact of the sand grains and the plasticized bonding materials. Maximum benefits from the bonding agents will be obtained only by properly mulling and blending with the sand under pressure and controlled additions of water, checked by a metering device. A centralized sand preparing arrangement will also make a direct material saving since the amount of bond used to maintain the strength of the sand can be measured and controlled.

Many factors must be considered in the selection of the proper sand mixer. The most important characteristic of a good mixer is its ability to provide uniform distribution of bonding materials and moisture throughout the batch. The production per hour and cost per ton of prepared sand, figured on the basis of operating costs including repairs, maintenance, and clectric current consumption, should also be considered.

Efficient operation of the mixer depends upon its auxiliary equipment for automatic charging of screened sand cleaned of magnetic refuse. Figure 8 illustrates a complete sand preparation arrangement. It consists of a vibrating screen with magnetic belt conveyor running underneath, a bucket loader for charging the mixer, the muller-type mixer with dust hood connected with the dust collecting system, and an aerator for fluffing of discharged prepared sand. The mixer's drive is located underneath, and the concrete pedestals are constructed to suit the required discharge height.

A more developed sand preparation system consists of a holding hopper which can be filled by a special truck, or by a box suspended from an overhead crane (Fig. 9). The hopper feeds the sand through the regulating gate into a vibrating feeder. This feeder is a two-level unit; it has a perforated top for removal of oversize scrap and core butts and allows usable molding sand to pass through the perforation into the lower deck of the vibrating trough. The screened sand is passed over a magnetic pulley for removal of me-

0

q

te

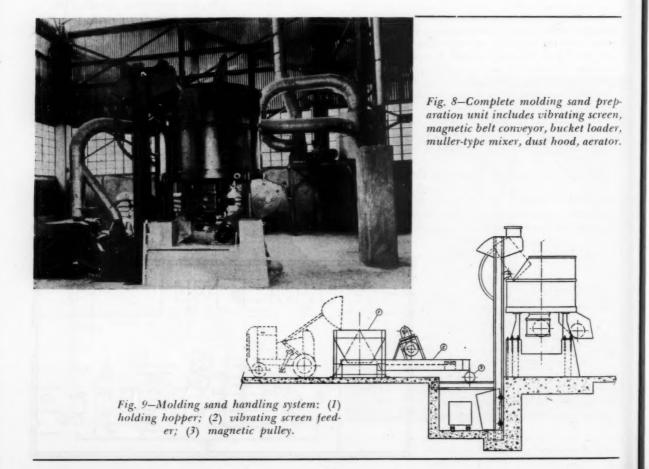
T

fo

in

St

lay me loc



RY

of of one of some of s

or

ist

nd

id.

n-

sts ial ne ua of ldhe ad

r.

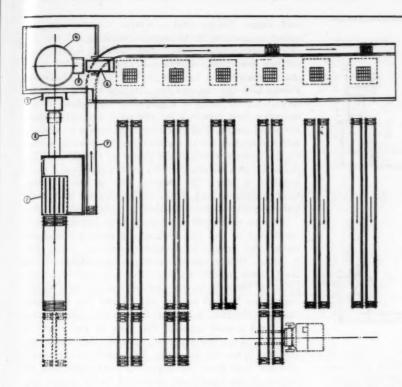


Fig. 10-Sand handling layout; (1) vibrating shake-out; (2) magnetic belt conveyor; (3) bucket loader; (4) mixer with dust hood; (5) aerator; (6) distributing buggy; (7) flask return roller conveyor.

tallic impurities, and then discharged into the bucket of the loader.

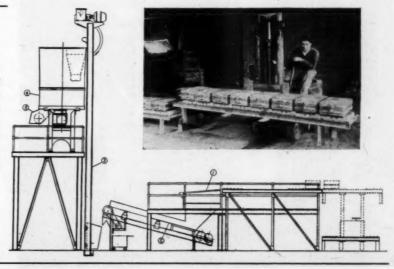
Both the above described arrangements, while compact in size and simple in design, fulfill the basic requirements for cleansing and conditioning of molding sand.

One of the layouts shown comprises a complete system for handling sand and molds (Fig. 10 and 11). The arrangement does not require pits or special foundations, and can be installed in almost any building. Conventional roller conveyors are used for the storage and pouring of molds. The cooled molds are pushed onto the transfer section of the roller conveyor, which is then picked up by a monorail hoist or lift

truck and carried down to the end of the foundry (Fig. 12). An operator on the elevated platform moves the molds to the vibrating shake-out while the truck operator goes back for another load. The sand passes through the shake-out bars onto the magnetic belt conveyor to be delivered to the bucket loader. The castings are delivered via a chute to a container on the floor, and the empty flasks are returned on the roller conveyor line to the molding stations.

Many times foundrymen have hesitated to install an overhead distributing system in their foundry because they felt that their building was too low; they were afraid that such an arrangement would be too expensive. Figure 13 illustrates suggested solution in

Fig. 11-Elevation of sand handling layout: (1) vibrating shake-out; (2) magnetic belt conveyor; (3) bucket loader; (4) mixer with dust hood; (5) aerator. The photograph (Fig. 12) shows a lift truck about to move poured molds to position for elevating to shake-out level.



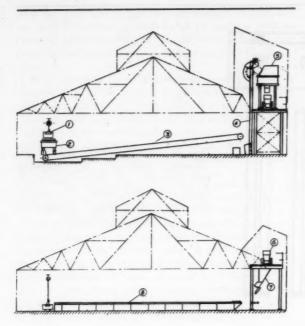


Fig. 13—Sand handling layout; (1) box for poured molds; (2) vibrating shake-out; (3) magnetic belt conveyor; (4) bucket-loader; (5) mixer with dust hood; (6) distributing buggy; (7) molder's hopper; (8) roller conveyor.

such cases. Notice that a simple construction was added on one side of the foundry building to cover the sand distribution system. The poured molds after cooling, are dumped into a bucket suspended from the monorail and brought to a vibrating shakeout. From there the castings are directed to the cleaning room and the sand is transported by means of the magnetic belt conveyor to the sand preparing unit. Prepared sand is discharged from the mixer to a buggy which distributes it to the molders' hoppers by dumping its content through the grates in the floor of the platform; this installation can serve a group of 4 to 12 molders or more. It presents one of the simplest, most adaptable and least expensive sand and mold handling systems for a medium-size gray iron foundry.

A more developed sand preparing arrangement is illustrated in Fig. 14. This arrangement includes the facility for storage of shake-out sand in a bin. The molds are shaken out on the vibrating shake-out. The sand, after passing through the magnetic belt conveyor, is lifted by an elevator, discharged to a revolving screen located on the top of the bin and then stored in the bin. A holding hopper with air-operated gate, and a sand capacity the same as the mixer, is attached to the lower part of the bin. It is filled with the sand from the bin gate. By a push-button operation the sand from the hopper is discharged into the bucket of the loader and delivered to the mixer. Figure 14 shows the upper part of this hopper, the bucket loader located on the side of the bin, and the mixer

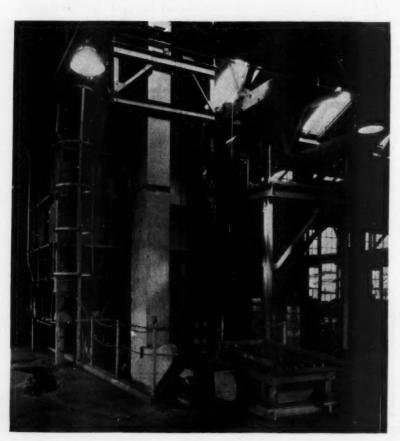


Fig. 14—Sand handling arrangement includes vibrating shake-out, elevator, storage bin, measuring hopper, bucket loader, mixer, with dust hood, distributing platform and molder's hopper.

on per sid arr is 1 Th two due

H.

jolt roll of the ont pot ferriflas and con

cop

bet having stor

ran

ret

s the graR)

he ter om ut.

he

it.

g

p

he

12

St.

ld

ry.

is

he

he he nlv-

en ed is th

a

he

et

er

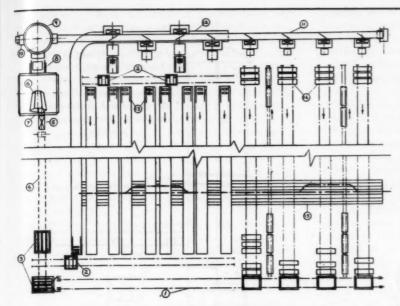


Fig. 15—Sand and mold handling layout: (1) apron conveyor; (2) transfer car; (3) vibrating shake-out; (4) magnetic belt conveyor; (5) elevator; (6) hexagonal revolving screen; (7) sand storage bin; (7A) measuring hopper; (8) bucket loader; (9) mixer with dust hood; (10) aerator; (11) distributing belt conveyor; (12) flask return roller conveyor; (13) roller conveyors; (14) pallet lines, (15) pouring platform.

on the distributing platform. A molders hopper suspended over a molding machine is visible on the right side of the picture.

The layout as illustrated in Fig. 15 is applicable to a foundry having small and medium-sized work. The arrangement comprises a minimum of equipment and is recommended for a foundry starting to mechanize. The mold handling and shake-out system is shown in two variations. It can be applied to suit different production methods.

The molds are made in tight flasks on two sets of cope and drag machines, and in snap flasks on four jolt squeezers. In the first section the drags made on rollover and jolt machines are pushed on short sections of roller conveyors for finishing and setting of cores. The copes made on jolt machines and brought over the drags for closing. Each completed mold is passed onto a transfer car and directed to the storage and pouring lines. The poured and cooled molds are transferred by another car to the vibrating shake-out. Snapflask molds, produced in the second section, are stored and poured on pallets which take the place of the conventional roller conveyors. These pallets can be returned on tracks under the pouring lines or can be stacked and returned on special return lines located between the pouring lines, as shown in Fig. 15.

The advantage of the pallet arrangement consists in having the molds located close together and in allowing the use of all of the entire floor space for mold storage. This layout offers almost unlimited flexibility in mold production for intermittent or continuous pouring. The capacity of the sand preparation arrangement, the type of machines, and the space for the mold storage can be adjusted to the required production output.

Since we have limited this paper to a discussion of the sand and mold handling system for a medium-size gray iron foundry, we will not enter the field of the power mold conveyors and elaborate sand transportation systems. The proper application for mold conveyors take place generally in a foundry having established type of production and continuous pouring; very seldom in the case of a jobbing medium-size gray iron foundry.

Melting Section

The melting section of the gray iron foundry offers numerous opportunities for mechanization. Handling of raw materials in the yard, mechanical charging of the furnaces, application of auxiliary equipment for controlled melting, and use of modern pouring units will contribute to the economy and efficiency of foundry operation.

Handling of raw materials starts in the yard when cars of scrap, pig iron, coke and limestone are delivered. The crane equipped with magnet and clam-shell bucket will perform the unloading at much less cost and time than the manual operation. The tonnage of materials handled will determine if the investment can be justified.

Cupolas today are essentially the same as those built in the early days of the gray iron foundry. Elimination of hand charging is made possible by mechanical charging equipment. A properly designed charging system will pay for itself in a short time through reduction in labor, saving in materials and elimination of the investment in a charging platform.

The mechanical charger simplifies the material handling problem since coke, iron and limestone are put into the same bucket. It also improves melting practice by allowing the raising of the cupola charging door, making possible an increase in height of the charge above the melting zone. Such an arrangement permits preheating of larger amounts of iron, gives greater melting speed and hotter metal since the charge receives full benefit of the fuel.

Cupola chargers are either skip hoist or crane type. The size of the unit and the selection of the right type

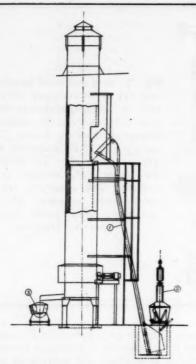


Fig. 16—Charging arrangement for cupola: (1) skip charger; (2) weigh lorry; (3) holding U-ladle.

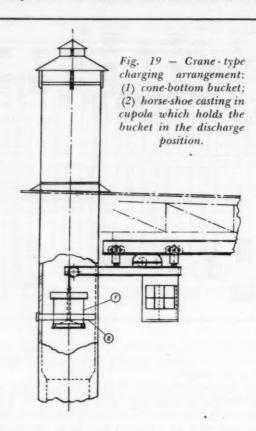
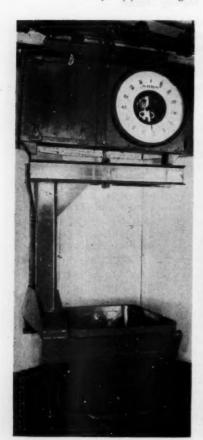




Fig. 17—Weigh lorry with bottom- Fig. 18—Handling and charging equipdump bucket is pushed along mono- ment for the foundry material yard rail loop for loading with weighed includes charging crane with magnet amounts of scrap, pig iron, coke, and and clamshell buckets, and hopper limestone.





(D.1.)

bu is di

ch

a

de

de

fo

ra

tr

le

lo co 17 ab

ha us cla la fo up scribin tiv

of cas slo un tro

gre the

car

is

•

depends upon many factors such as yard space, size and number of cupolas, kind of scrap available, etc.

The skip hoist charger (Fig. 16) is of inexpensive design, simple in operation and particularly suited for the small or medium-size foundry. The bucket is raised and lowered by a cable attached to an electric hoist. It is mounted on wheels for travel on vertical track, curved at the top. The bucket is filled at floor level by means of a wheelbarrow or weigh lorry.

The monorail type weigh lorry with bottom-dump bucket is pushed by the operator along the monorail loop, loading the required amount of scrap, pig iron, coke and limestone into a bucket on the lorry (Fig. 17). After bringing the loaded lorry into a position above the skip hoist bucket, the operator releases the charge by hand lever.

Some of the medium-size foundries may be able to have a yard crane for unloading of incoming material, using a magnet for handling of pig iron and scrap, and clam-shell bucket for coke and limestone. The typical layout of handling and charging equipment for such a foundry is shown on Fig. 18. The iron charge is made up on a platform having space for various types of scrap and pig iron. The required amount of coke and limestone is discharged into the bucket from respective bins.

The most popular form of bucket for a crane-type charger has a cone-shaped bottom which is attached to a center stem suspended from the hoisting mechanism (Fig. 19). The horse-shoe casting, which holds the bucket in the discharging position, is located in the cupola stack. It serves to support the shell of the bucket while the bottom is lowered and the material is discharged.

The main advantage of this charging system is the distribution materials inside and around the periphery of the cupola with minimum impact. The horse-shoe casting permits the bottom of the bucket to be lowered slowly and in this way makes it possible a controlled unloading of the bucket's content.

ird

ne.

In general, because of more accurate melting control, a properly mechanized handling system will improve the quality of castings. It will provide also for greater safety and simplify the supervisor's job.

The accurate cupola operation should be based on the supply of a constant weight of air. Variation in atmospheric temperature and barometric pressure may cause the weight of oxygen in the given volume of air to vary as much as 20 per cent. When the air supply is controlled a uniform melting rate can be obtained.

Two types of measuring systems has been developed

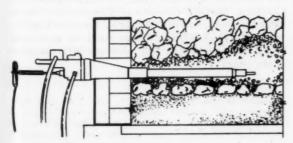


Fig. 20-Electric coke ignitor.

and are giving satisfactory service. One system of air control consists of an electrical system which is connected to the current supply of a constant speed motor driving a centrifugal blower. The volume of air is varied automatically by means of a small motor-driven gate in the blast line. The system is based on the principle that a constant weight of air is delivered when the electrical input is constant. If the volume tends to increase or decrease, due to changes in the resistance through the cupola or due to the atmospheric changes, the power meter causes the gate to open or close, enough to keep the power constant and, thereby, the weight of air going into the cupola. This arrangement can be used in connection with fan type only.

The other system of air control measures the flow of air by means of an orifice plate or Venturi tube installed in the blast line. This type of control can be applied to either a positive displacement blower or fan-type blower. In the case of the positive blower, the excess air is bled to atmosphere and the controlled valve is placed above the outlet flange of the blower. In the case of fan-type blower the valve is located in the blast line. The compensating mechanism includes a bellows whose expansion or contraction under varying conditions provides changes in the leverage of the controlling instrument.

There is another improvement in cupola operation which is worth mentioning, namely, an electric coke



Fig. 21—U-type tilting ladle, receiving metal from the cupola, has sufficient capacity to hold hot metal during intermittent pouring.

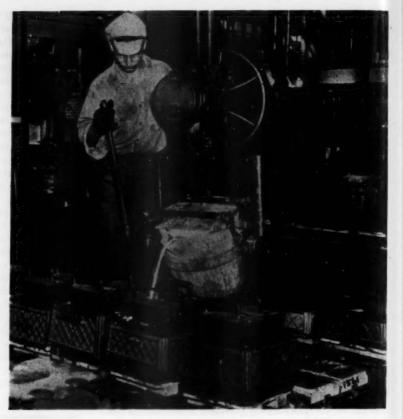


Fig. 22-Molds are poured from covered ladle equipped with raising and lowering mechanism and suspended from a monorail system.

ignitor as shown on Fig. 20. The firing of the coke bed by kindling of wood planks was always an unpleasant operation, causing excessive smoke and usually many unburned wood pieces and ashes mixed with coke. The electric ignitor replaces the tedious operation and lights the coke thoroughly in a short time without use of wood. The barrel of the ignitor is inserted through the breast of the cupola and placed approximately 5 in. above the sand bottom in a layer of coke breeze. The electric spark ignites a small amount of fine coke in front of the barrel. This fire is spread through the coke bed by means of compressed air; the electric torch being replaced by air tube.

Metal melted in the cupola usually is received in a U-type tilting ladle (Fig. 21), or directly into larger ladles to be transferred in the pouring ladles at the hot metal distribution point. The U-ladle is heavily lined with refractory, and has sufficient capacity to serve as reservoir of hot metal during intermittent

Molds are poured from covered ladles, equipped with raising and lowering mechanism, and suspended from a monorail (Fig. 22). This type of ladle makes it possible for one man to do the pouring. It provides also for easy handling of hot metal with greatest safety

and minimum fatigue.

The foregoing descriptions and illustrations give a general survey of basic units and layouts recommended for a progressive medium-sized gray iron foundry. It should be emphasized that each particular foundry needs a "tailor-made" arrangement of suitable units combined in an efficient and smoothly operating system. This final system has to be established for the entire foundry, and the modernization should start in all sections simultaneously. The final layout drawing should show in dotted lines the space reserved for equipment being considered for future stages of mechanization. The layout should be flexible enough to provide the effective use of the units under changing conditions and varying rates of production.

Din eve

ble

eff

me

mil

ba

T

fre

en

or

su

U

co fo

ca

Acknowledgment

The author expresses his thanks to Mr. B. L. Simpson, President of National Engineering Company, Mr. Charles A. Barnett, President of The Foundry Equipment Company, and Mr. Walter R. Jaeschke, Consulting Engineer of Whiting Corporation for permission to use the photographs and assistance in preparation of the paper.

MELT QUALITY AND FRACTURE CHARACTERISTICS OF 85-5-5-5 RED BRASS AND 88-8-4 BRONZE

By

R. D. Shelleng,* C. Upthegrove,** and F. B. Rote ***

ABSTRACT

This is the third progress report on a research project sponsored by the Research Committee of the A.F.S. Brass and Bronze Division on the development of a fracture test procedure for evaluating the melt quality of brass and bronze alloys.

In the first progress report 1 it was shown that fracture characteristics of sand cast flats, 1/2 to 2 in. thick, could be correlated with melt quality. However, the rather long time required for this test decreased its usefulness for production control.

In the second progress report 2 a more rapid (3-min) test for melt quality was proposed. This test utilizes a 3/4 in. x 3 in. x 6 in. block, cast against a gray iron chiller. Instructions for the use of this test and evaluation of the fractures were prepared in a separate pamphlet.

The present report presents data obtained in evaluating the effects of variations in the size and shape of the chill block casting and variations in the temperature of the chiller from room temperature to 700 F on the fracture appearance of 85-5-5-5 alloy chill blocks. A determination of the cooling rates in a 3/4 in. x 3 in. x 6 in. chill block has also been made. The relations between melt quality appearance of 88-8-4 alloy have been investigated. A study was also made of the influence of degassing with dry nitrogen on the fracture appearance and melt quality of 85-5-5-5 and 88-8-4 alloys.

Introduction

The DETRIMENTAL EFFECTS of dissolved reducing gases, particularly hydrogen, on the quality of copperbase alloys are well known. These alloys have the ability to dissolve gases in varying quantities and, therefore, to suffer varying degrees of impairment of quality. The quality of a particular alloy might vary widely from melt to melt, due to small uncontrolled differences in melting practice, with the result that the brass or bronze foundryman is generally uncertain as to the success he might expect when he casts a particular melt. Until recently there has been no satisfactory test which could be used to evaluate the quality of the metal before casting, so that most foundrymen simply made the castings and tested the metal at a later time.

The present project, sponsored by the Research Committee of the Brass and Bronze Division of the American Foundrymen's Society was initiated at the University of Michigan, to investigate the possibility of the development of a test procedure by which melt quality in red brass could be evaluated before the metal was cast.

In the early stages of the project a study was made of the interrelationship of melt quality of 85-5-5-5 red brass and fracture characteristics of sand cast flats ½ in. to 2 in. thick. A definite correlation was observed, and it was found that the tensile properties of standard test bars could be predicted from fracture appearance in sand cast flats. However, the time required for the solidification, notching, and fracturing of the test flats was too long for the test to be valuable as a production control.

In an effort to speed up the test, the use of a chilled fracture block was investigated. It was found that melt quality could be correlated with the fracture appearance of a chilled ¾ in. x 3 in. x 6 in. casting, and that tensile properties could be predicted. The test using the chilled casting requires only about 3 min, which makes it useful as a production control, since it is not necessary to hold molten metal for unduly long time periods while the test casting is examined.

The present report includes data secured in attempts to refine the chilled fracture test bar procedure and to investigate its usefulness for evaluating the melt quality of other copper-base alloys.

In the earlier stages of the project, melts were intentionally made either of high or low quality. No attempts were made to secure data from melts of poor quality which were later improved by degassing with dry nitrogen. Since this practice is relatively common in commercial foundries, it was desired to secure data from heats in which this practice was used. Such data are presented in this report.

H. H. Fairfield ³ reported in a discussion of the second progress report, the successful use of a chilled fracture test bar as a production control in a commercial foundry casting several copper alloys, including 85-5-5-5 and 88-8-4. Part of the work in the present project has been conducted to evaluate the usefulness of the chilled casting for 88-8-4 alloy and to determine a procedure for interpreting fractures of this alloy. The findings of these investigations are presented in this report.

NDRY

ed

It

rv

its

yshe

in

ng

h.

to

ng

P

^{*} Research Fellow and ** Professor, Department of Metallurgical Engineering, and *** Associate Professor, Departments of Metallurgical Engineering and Metal Processing, University of Michigan, Ann Arbor, Mich.

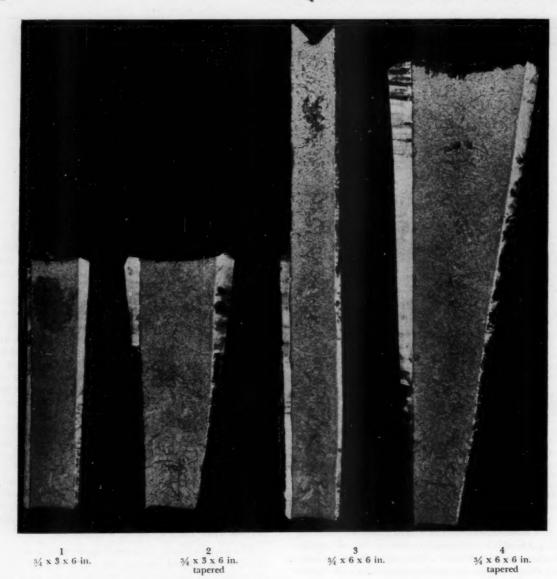


Fig. 1a-Effect of size of chill block on fracture characteristics of 85-5-5 red brass. (Induction furnace melt No. 85, high quality.)

Procedure

A total of 50 melts were made in an induction furnace with a magnesia lining or in a gas furnace with a clay-graphite crucible. The chill blocks were cast in core sand molds which were made of lake sand (A.F.S. 50 grain size) bonded with 1 per cent oil, 1 per cent cereal and 1 per cent bentonite. The tensile test bars (ASTM B 208–49T Fig. 4) were cast in dried molds made of bank sand (A.F.S. 90 grain size) bonded with 3 per cent bentonite. The charges were standardized at 50 per cent new ingot and 50 per cent scrap from previous melts. The melts were heated to 2250 F and tapped into a preheated ladle. Additions of 2 oz of phosphor-copper per 100 lb of metal for deoxidation, and 1 lb of zinc per 100 lb of metal to compensate for melting loss, were made. Chill blocks and tensile test

bars were poured at temperatures ranging from 2175 F to 2050 F. Many variations in melting practice other than those steps detailed above were used to produce heats of varying melt quality; however, the above steps were followed in all melts.

Chill Block

The effects of decreased thickness or height of the chilled block on the fracture characteristics as indicators of melt quality were investigated previously and reported in the second progress report. It was concluded from the data presented that the 3/4 in. x 3 in. x 6 in. block gave a more reliable fracture for the prediction of melt quality. In order to evaluate the effect of tapering the chilled block and increasing its dimensions, chill blocks of the following dimensions were made:

lt

F ner ice eps

he dind n-. X licof enere

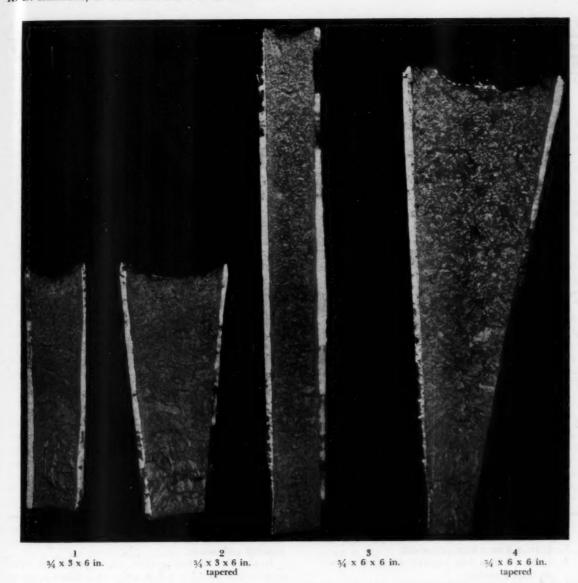


Fig. 1b-Effect of size of chill block on fracture characteristics of 85-5-5-5 red brass. (Gas-fired furnace melt No. 73, low quality.)

TABLE 1

85 High	3/4 x 3 x 6	Fine.		
		Fine	13/4	Slight
	3/4 x 3 x 6 tapered	Fine	2	Slight
	3/4 x 6 x 6	Fine	21/4	Intermediate
	3/4 x 6 x 6 tapered	Coarse	3	Intermediate
73 Low	3/4 x 3 x 6	Fine	11/8	Intermediate
	3/4 x 3 x 6	Coarse	11/8	Intermediate
	3/4 x 6 x 6	Coarse	11/16	Intermediate
	3/4 x 6 x 6 tapered	Coarse	11/4	High

d

is

tu

ar

TABLE 2-EFFECT OF SIZE OF CHILL BLOCK ON FRACTURE CHARACTERISTICS OF 85-5-5-5

Heat Num- ber	Melting Furnace		Per cent Elongation in 2 in.		Pouring Temp.,		Block Dimen- sions	Grain Size	Blue Gray	Columar Grain Depth, in.	Color Above Blue Gray	Degree of Mottle	Oxidized Shrink
60	Induction	31900	19	8.86	?	Low	1	Fine	15/8	1/2	B and O	2	No
					Tapped		4	Fine	13/4	1/2 1/2 8/8 1/2	B and O	3	No
			1		at		5	Fine	21/2	3/8	B and O	2	Rust
					2280		6	Coarse	21/2 21/2	1/2	B and O	3	No
62	Induction	36000	22	8.85		Low	1	Fine	17/8	3/16 1/2 3/8 11/16	Red Brown	2	No
					2130		4	Fine	23/8	1/2	B and O	2	No
							5	Fine	25/8	3/8	B and O	2	No
							6	Coarse	11/2	11/16	B and O	3	No
73	Gas	33800	21	8.79	2150	Low	1	Fine	1½ 1½	5/8 5/8 3/4 11/4	B and O	2	No
							4	Coarse	11/8	5/8	B and O	2	No
							5	Coarse	11/16	3/4	B and O	2	No
							6	Coarse	11/4	11/4	B and O	3	No
74	Gas	34800	23.5	8.67		Low	1	Coarse	13/4	11/8	B and O	- 2	No
					2150		4	Fine	11/4	11/8	B and O	2	No
							5	Coarse	19/16	11/8	B and O	2	No
							6	Coarse	23/8	11/8 11/2	B and O	3	No
75	Induction	37000	27	8.95	2050	Inter-	1	Coarse	13/4	1/4	В	2	No
						mediate	4	Coarse	13/4	3/8	В	2	Rust
							5	Fine	15/8	9/16	B and O	2	Rust
							6	Coarse	2	1/4 3/8 9/16 1/2	Gray	3	Rust
76	Induction	39200	35	8.9	2110	High .	1	Fine	2	1	В	2	Rust-Green
								Fine	17/8	3/4	В	2	Rust
							5	Fine	17/8 28/4	3/4	В	2	Rust
77	Induction	36000	24.5	8.76	2150	Low	1	Coarse	15/16	7/8 1/2	B and O	2	No
							4	Coarse	11/2	1/2	B and O	2	No
							5	Coarse	15/8	0	B and O	2	No
							6	Coarse	11/2	1/2	B and O	3	No
80	Induction	35300	26	8.72		Inter-	1	Coarse	15/16	3/8	O and B	2	No
					2100	mediate	4	Coarse	15/8	0	O and B	2	No
							5	Coarse	7/8	0	O and B	2	No
							6	Coarse	1	0	O and B	3	No
85	Induction	38700	33	8.9	2140	High		Fine	13/4	716 1/2 1/2 13/6	B and O	1	Rust
								Fine	2	1/2	B and O	1	Rust
								Fine	314	1/2	B and O	2	No
							6	Coarse	3	13/16	B and O	2	No
	1-34 x 3 x		-1			‡ B—Blue							
	4-34 x 3 x					O-Orange							
	5-34 x 6 x					† Rated 1-1							
	6-34 x 6 x	6 Tapered					-Mediur		2				
		700				3-	-Heavy	mottle					

1. 3/4 in. x 3 in. x 6 in. (standard)

2. 3/4 in. x 3 in. x 6 in. tapered to 11/4 in. at the air-cooled surface

3. 3/4 in. x 6 in x 6 in.

4. 3/4 in. x 6 in. x 6 in. tapered to 11/2 in. at the air-cooled surface.

All were cast under the procedure detailed above. Data from these melts are presented in Table 2.

Representative fractures are shown in Fig. 1a (high melt quality) and 1b (low melt quality). The depth of the blue-gray zone in the 3/4 in. x 3 in. x 6 in. block of the high quality melt is 13/4 in. The area above the blue-gray shows only a small amount of mottle. The overall texture of the fracture is fine. The blue-gray depth in the fracture of the tapered $\frac{3}{4}$ in. x 3 in. x 6 in. block is 2 in., about the same as in the straight 3/4 in. x 3 in. x 6 in. bar. There is more mottle in the region above the blue-gray, and the fracture is coarser in this area. The fracture of the 3/4 in. x 6 in. x 6 in. block is blue-gray to a depth of 21/4 in. The upper end is somewhat more mottled and coarser above the blue-gray than in the 3/4 in. x 3 in. x 6 in. block. The blue-gray depth in the fracture of the tapered 3/4 in. x 6 in. x 6 in. block is 3 in., and the area above the blue-gray is very

mottled and coarse.

In the Fig. 1b (low quality), the depth of the bluegray zone in the 3/4 in. x 3 in. x 6 in. chill block is 11/8 in. The area above the blue-gray zone is quite mottled, and the overall texture of the fracture is coarser than the corresponding high quality fracture in Fig. 1a. The blue-gray depth in the fracture of the tapered 3/4 in. x 3 in. x 6 in. chill block is 11/8 in. The area above the blue-gray is more mottled and coarser than either the high quality tapered 3/4 in. x 3 in. x 6 in. block or the low quality straight 3/4 in. x 3 in. x 6 in. block. The fracture of the 3/4 in. x 6 in. x 6 in. block is blue-gray to a depth of 1-1/16 in. The area above the blue-gray zone is more mottled and coarser than in the corresponding high quality fracture. The blue-gray depth in the fracture of the tapered 3/4 in. x 6 in. x 6 in. block is 11/4 in. The area above the blue-gray zone is much coarser and more mottled than in the corresponding high quality fracture. A summary of these data are presented in Table 1.

The above data demonstrate that increasing the height of the chill block from 3 in. to 6 in. tended to increase the depth of the blue-gray zone in high-quality metal but had little effect on the blue-gray zone in

ZE

in.

nd he he . x he he he he ay ay rein is ch ng re he

to

al-

in

low-quality melts. Tapering the chill block gave a coarser and more mottled fracture above the blue-gray region.

A comparison of the same size chill blocks from high and low quality melts and high and intermediate quality melts shows that there generally is a proportionately greater difference in the depth of blue-gray in the 6-in. straight blocks than in the other size blocks. However, there generally is a proportionally greater difference in the depth of blue-gray in the 3-in. chill blocks when the comparison is made between low and intermediate quality melts. The 6-in. blocks are, therefore, more sensitive in differentiating high quality from low and intermediate quality melts, but less sensitive when differentiating between intermediate and low quality melts.

In general, the fractures of the higher 6-in. blocks were not found to be as reliable as the fractures of the 3-in. blocks, because there was greater variation in the depth of blue-gray for a particular range of elongation in tensile test bars poured with the fracture test bars. The fractures of the 3-in. tapered blocks were substantially the same as the 3-in. untapered, except for the increase in mottle and coarseness above the blue-gray zone. It is believed that, of the sizes investigated, the 3/4 in. x 3 in. x 6 in. chill block gives the most reliable fracture, is most economical, and provides the most rapid test.

Variation in Chiller Temperature

In view of the fact that, in a commercial foundry, many chill blocks might be cast in succession on the same chiller, it was desired to evaluate the effect of the temperature of the chiller on the fracture characteristics. Several heats were made, casting 3/4 in. x 3 in. x 6 in. chill blocks on chillers at room temperature, 400, 600, and 700 F. Data from these heats are presented in Table 3.

Fractures from a high quality and a low quality heat are shown in Fig. 2a and 2b. The high quality fractures, Fig. 2a, have blue-gray zones of 2 in., 2 in., 17/8 in., and 17/8 in. for the blocks poured on chillers at room

temperature, 400, 600, and 700 F., respectively. The coarseness and mottle are about the same for all four blocks. None of the low quality fractures, Fig. 2b, have blue-gray zones. The grain size and mottle in these fractures are also about the same in all four blocks.

Variations in depth of the blue-gray zone for the blocks poured on chillers at varying temperatures are within the variation encountered among several heats of the same tensile strength and elongation. Therefore, it is believed that the temperature of the chiller, up to 700 F, has no significant effect on fracture appearance of chilled castings.

Extension of Test to 88-8-4 Alloy

Several heats of varying quality were made to secure information from which a correlation between melt quality and chill block fracture characteristics of 88-8-4 alloy might be obtained. Data from these heats are presented in Table 4. Considerable difficulty was encountered in producing heats of low quality. The addition of wet clay to an induction melt just before tapping seemed to have little or no effect on the properties obtained.

The tensile properties have been classified in three ranges—high, intermediate and low. A high quality melt has at least 47,000 psi tensile strength and 50 per cent elongation in 2-in. Intermediate quality includes melts from 40,000 psi to 47,000 psi tensile strength and 35 to 50 per cent elongation. Low quality melts fall below 40,000 psi tensile strength and 35 per cent elongation.

In order to produce a poor quality heat it was necessary to melt in a gas furnace with a strongly reducing atmosphere. When the melt temperature reached 2250 F, wet clay was pushed beneath the surface of the melt to obtain an intimate contact of the metal with the steam formed. The melt was then cooled to about 2150 F and poured.

Fractures of 88-8-4 Alloy

Typical fractures for 88-8-4 alloy chill blocks are shown in Fig. 3. The high quality fracture, No. 1, has a

Table 3-Effect of Chiller Temperature on Chill Block Fracture Characteristics of 85-5-5-5 Red Brass

Heat Number	Melting Furnace	Tensile Strength, psi	Per cent Elongation in 2 in.	Density	Pouring Temp., °F	Quality	Grain Size	Chiller Temp., °F	Blue Gray	Columnar , Grain Depth, in.	Color Above Blue Gray	Degree of Mottle*	Oxidized Shrink
59	Induction	39250	36	8.84	2150	High	Fine Fine Fine Fine	Room 400 600 700	134 134 2 2	3/8 1/2 1/2 1/2	Blue-Orange Blue-Orange Blue-Orange Blue	2 2 2 3	No Rust No No
64	Gas	37500	32	8.87	2150	High	Fine Fine Fine	Room 400 600 700	2 2 17/8 17/8	1/2 5/8 3/8 3/8	Blue-Orange Blue-Orange Blue-Orange Blue-Orange	2 2 2 3	Rust-Gr. No No No
97	Induction	37500	26	89	2100	Inter- mediate	Coarse Coarse Coarse	Room 400 600 700	2 2 2 ¹ / ₈ 1 ⁷ / ₈	3/8 1/4 1/4 1/8	Gray Gray Gray Gray	3 3 3	Orange Orange Orange Orange
98	Gas	22200	11	8.25	2200	Low	Coarse Coarse Coarse	Room 400 600 700	0 0 0	0 1/2 1/2 1/2	Orange-Blue Orange-Blue Orange-Blue Orange-Blue	2 2 3 3	No No No No

^{*} Rated 1-No mottle

^{2—}Some mottle

^{3—}Heavy mottle

Chiller Temp., F

Chiller

Temp., F

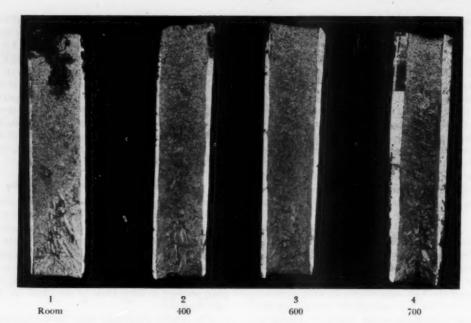


Fig. 2a-Effect of chiller temperature on fracture characteristics of 85-5-5 red brass. (Induction furnace melt No. 64, high quality.)

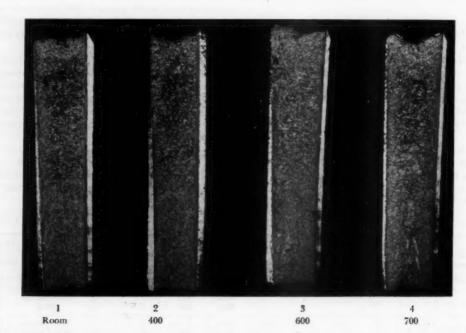


Fig. 2b-Effect of chiller temperature on fracture characteristics of 85-5-5-5 red brass. (Gas-fired furnace melt No. 98, low quality.)

fine, dense structure and is light yellow in color. The texture is finest and the color lightest at the chilled surface. The area near the air cooled surface is somewhat mottled and appears yellow-tan. The fracture is also coarser near the air cooled surface. The texture of the intermediate quality fracture, No. 2, is coarser than the high quality and the color is yellow-tan. The mottle in the upper part of the fracture is gray. The low

quality fracture, No. 3, is quite coarse and has a reddish-gray color.

A type of fracture which occurred frequently is shown in Fig. 4 (2). The area next to the chiller is fine textured and light yellow. A short distance from the chiller the color changes abruptly to gray, mottled with tan. At the top of the block near the air cooled surface a bright orange oxide spot appears. The abrupt

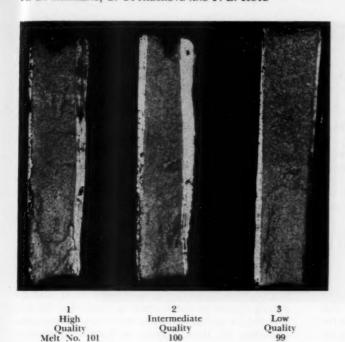
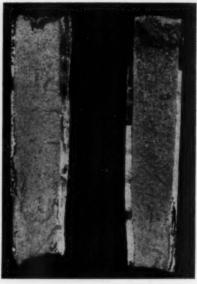


Fig. 3—Fracture characteristics of 88-8-4 bronze chilled blocks 3/4 x 3 x 6 in.



1 High Quality High Quality

Fig. 4—Typical fracture and zoned fracture of 88-8-4 bronze chilled blocks 3/4 x 3 x 6 in.

TABLE 4-TENSILE PROPERTIES AND CHILL BLOCK FRACTURE CHARACTERISTICS OF 88-8-4 BRONZE

Heat Number	Melting Furnace	Tensile Strength, psi	Per cent Elongation in 2 in.	Density	Pouring Temp., °F	Quality	Fracture Color	Fracture Texture*	Color of Mottle	Oxidized Shrink	Zones
86	Induction	50,500	63	8.76	2170	High	Yellow-Gray	MF	Gray	Yes	Yes
87	Induction	46,600	59	8.76	2175	Intermed.	Yellow-Tan	MF	Tan	No	No
88	Induction	47,000	65	8.80	2100	High	Yellow-Gray	MC	Gray	Yes	Yes
89	Induction	47,000	65	8.80	2100	High	Yellow-Gray	MC	Gray	Yes	Yes
90	Induction	47,000	50	8.85	1975	High	Yellow-Gray	C	Gray	Yes	Yes
91	Gas	43,500	46	8.82	2150	Intermed.	Reddish-Tan	MF	Tan	No	No
92	Gas	46,200	61	8.8	2150	Intermed.	Yellow	F	Tan	No	No
93	Gas	47,700	57	8.75	2150	High	Yellow	MF	Tan	No	No
94	Induction	47,500	58	8.74	2145	High	Yellow .	MF	Tan	No	No
95	Induction	41,100	36	8.73	2130	Intermed.	Yellow-Tan	F	Tan	No	Yes
96	Induction	48,000	52	8.73	2100	High	Yellow-Gray	MF	Gray	Yes	Yes
99	Gas	29,500	18	8.19	2175	Low	Reddish-Tan	C	Tan	No	No
100	Induction	44,600	40	8.80	2160	Intermed.	Reddish-Tan	MF	Tan	No	No
101	Induction	55,600	60	8.74	2155	High	Yellow-Tan	F	Tan	No	No
105	Induction	46,250	52.5	8.78	2145	Intermed.	Yellow-Tan	MF	Tan	No	No
105N	Induction	46,200	58	8.83	2120	Intermed.	Yellow-Tan	F	Tan	No	No
106	Induction	46,100	46.5	8.76	2150	Intermed.	Yellow-Tan	F	Tan	No	No
106N	Induction	45,700	61	8.79	2145	Intermed.	Yellow-Tan	F	Tan	No	No
107	Gas	48,000	59	8.76	2145	High	Yellow-Tan	MF	Tan	No	No
107N	Gas	43,400	33	8.85	2130	Low.	Yellow-Tan	F	Tan	No	No
109	Gas	37,700	24	8.70	2050	Low	Yellow-Gray	C	Gray	No	Yes
109N	Gas	45,900	49	8.70	2050	Intermed.	Yellow-Tan	MC	Gray	Yes	Yes

MF—Medium Fine
MC—Medium Coarse
C—Coarse

transition from yellow to gray occurred only when the oxide spot was also present. This type of fracture generally occurred when the metal was poured cold. For example: heats 88-90, 96, 109N were all poured at 2100 F or lower, and all had this type fracture. There were some exceptions to the generalized statement that this type of fracture is the result of pouring at 2100 F or less, which indicates that other factors, as yet undisclosed, might contribute to its appearance. It is believed that the oxide spot is caused by shrinkage dur-

ing solidification, which draws air into the chill block from the air cooled surface and produces the oxidized surface.

A high quality fracture is characterized by a fine texture and light yellow color. As the melt quality decreases, the fracture becomes coarser, and the color tends to be a gray-tan or reddish-tan. The fractures of 88-8-4 chill blocks, with the exception of the pouring high and low quality as does the 85-5-5-5 alloy. No measureable characteristic such as the blue-gray zone

in the 85-5-5-5 was found in the 88-8-4 alloy chill blocks, and no columnarity was observed in any of the fractures.

The same precautions which applied to the casting of 85-5-5-5 chill blocks must be observed when pouring 88-8-4 chill blocks, with the exception of the pouring temperature.

These precautions are:

1. The pouring temperature should be in the range 2120 to 2200 F.

2. All chill blocks should be examined for evidence of surface blows, due to gas pickup from the chiller or from the metal strip used for notching.

3. The pouring basin must be removed from the mold as soon as the metal is poured, in order to prevent the basin from chilling the top of the block.

4. The chilled blocks must not be quenched at too high a temperature or an unreliable fracture will result.

Degassing with Dry Nitrogen

The possibility of improvement of melt quality by sweeping dissolved gases from the molten metal with dry nitrogen is well known. In order to investigate the influence of degassing with nitrogen on fracture appearance in the chill block casting, several heats were made using the procedure previously outlined for producing a poor quality melt. After the wet clay was added, about half the metal was poured into a preheated ladle. The zinc and the phosphor-copper additions were made, and the two chill blocks and one test bar mold were poured. The remaining metal was returned to the furnace and degassed with a measured amount of nitrogen. Phosphor-copper and zinc additions were made, and two more chill blocks and a test

bar mold were poured.

Data obtained from these heats are presented in Table 5. Representative 85-5-5-5 alloy chill blocks are shown in Fig. 5a. Chill blocks 1 and 2 were poured after the addition of wet clay, but before degassing with nitrogen. Chill blocks 3 and 4 were poured after degassing. The height of the blue-gray zone in block 1 and 2 is ½ in. The area above the blue-gray zone is rather coarse and mottled. The tensile properties of the bars cast with chill blocks 1 and 2 averaged 36,000 psi tensile strength and 21 per cent elongation. After degassing with 2.4 cu ft of nitrogen per 50 lb of metal the blue-gray zone was raised to 15% in., and the area above the blue-gray zone became somewhat finer than that in blocks 1 and 2, and had less mottle.

The tensile properties of the bars accompanying chill blocks 3 and 4 averaged 39,400 psi tensile strength and 29.5 per cent elongation. The nitrogen degassing brings about an increase in the tensile properties of the metal and an accompanying increase in depth of the blue-gray zone in the chill block. The color of the blocks poured after degassing is much bluer, and the overall texture of the fracture is finer, also the area near the air cooled surface is much less mottled after degassing. All these changes in fracture appearance indicate improved melt quality.

Figure 5b shows typical fractures accompanying the improvement of the melt quality of 88-8-4 alloy by degassing with nitrogen. Fractures 1 and 2, poured before degassing, have a coarse granular surface, which is dull gray-tan. Fractures 3 and 4 were poured after degassing with 2.5 cu ft of nitrogen per 60 lb of melt. These fractures are more yellow and less gray than fractures 1 and 2, and the texture is finer. The fractures obtained after degassing a low quality melt appeared the same

Table 5-Effect of Degassing with Dry Nitrogen on Tensile Properties and Chill Block Fracture Characteristics

Heat Num ber	Melting Furnace	Tensile Strength, psi	Per cent Elongation in 2 in.	Density	Temp.,	Degass- ing Time Min.		Quality	Grain Size	Depth of Blue Gray in.	Columnar Grain Depth	Color Degree Blue Gray	Degree of Mottle*	Oxidized
	Induction Induction	38400 37600	32.5 32.0	8.95 8.92	1975 1975	10	1.3	High High	Coarse Coarse	3 3	1/4 3/16	Blue Blue	3	Orange Rust
	Induction Induction	38300 38700	36 36	8.89 8.87	2170 2170			High High	Fine Fine	13/4 15/8	3/4 5/8	Blue Blue	2 2	No No
	Gas Gas	34200 38000	17.5 25.5	8.65 8.61	2130 2130	10	2.4	Low	Coarse Coarse	1/2	12	Blue-Orange Blue-Orange		No No
108N	Gas Gas	39500 39300	29 30	8.83 8.85	2130 2130			Inter. High	Fine Fine	15/8 15/8	3/4 3/4	Blue-Orange Blue-Orange	2 2	No No

88-8-4 BRONZE

Heat Num- ber	Melting Furnace	Tensile Strength, psi	Per cent Elongation in 2 in.	Density	Temp.,	Degass- ing Time, Min.		Quality	Fracture Color	Fracture Texture	Color of Mottle	Zones	Inclusion
2 106N1	Induction Induction Induction	45800 46300 45700 45700	48 45 55.5 61	8.77 8.75 8.77 8.58	2150 2150 2145 2145	15	3.5	Intermediate Intermediate Intermediate Intermediate	Yellow-Tan Yellow-Tan Yellow Yellow	Med. Fine Med. Fine Fine Fine	Tan Tan Tan Tan	No No No No	No No No No
109 1 2 109N1	Gas Gas	38100 37300 45800 45900	24 24 46 49.5	8.7 8.7 8.82 8.7	2050 2050 2050 2050 2050	10	2.5	Low Low Intermediate Intermediate	Gray Gray Tan Tan	Coarse Coarse Med. Fine Med. Fine	Tan Tan Tan Tan	No No No Yes	No No No Orange

N indicates melts degassed with nitrogen.

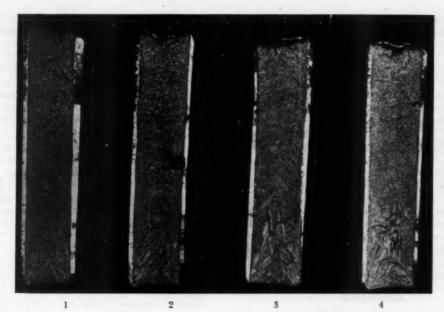
* Rated 1—No mottle 2—Some mottle 3—Heavy mottle in are ter ith

18 he 120 lehe ve at ng th ng of of ne 1e ea er

ce

ne

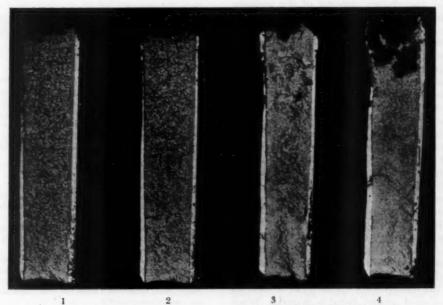
e-re ll ng se es ed ne



Low Quality
I and 2 were poured from poor quality melt; 3 and 4 were poured from same melt after degassing

Fig. 5a-Effect of degassing with nitrogen on fracture characteristics on 85-5-5-5 red brass. (Gas-fired furnace

melt No. 108.)



Low Quality

1 and 2 were poured from poor quality melt; 3 and 4 were poured from same melt after degassing

Fig. 5b-Effect of degassing with nitrogen on fracture characteristics of 88-8-4 alloy. (Gas-fired furnace melt No. 109.)

as fractures from melts of similar tensile properties made under improved melting practice, such that degassing was not required. The tensile properties of the test bars accompanying blocks 1 and 2 averaged 24 per cent elongation and 38,000 psi tensile strength. After degassing, these properties rose to an average of 48 per cent elongation and 45,900 psi.

Degassing a poor quality melt will change the fracture obtained in the chilled block, as the tensile properties are improved. In the case of 85-5-5-5 alloy the depth of the blue-gray zone is increased, the overall texture is refined and the mottle in the upper area is decreased. When a poor heat of 88-8-4 is degassed with dry nitrogen, the color of the fracture changes from

gray mottled with tan to a yellow-tan, and the texture of the fracture is refined.

Conclusion

The results of the investigations to date may be summarized as follows:

1. The $\frac{3}{4}$ in. x 3 in. x 6 in. chilled block appears to be the most satisfactory of the sizes investigated for foundry quality control.

2. There is no effect on the fracture appearance of varying the chiller temperature from room tempera-

ture to 700 F.

3. A definite correlation has been found between melt quality of 88-8-4 alloy and fracture appearance of chilled blocks. A high quality fracture is characterized by a fine, dense texture and a light yellow color. A low quality fracture is reddish-tan or gray-tan in color and has a coarse texture.

4. Heats degassed with dried nitrogen show an improvement in fracture characteristics, accompanying the improvement in melt quality.

ADDENDUM

Practical Applications of Test

It is realized that considerable care must be exercised when performing the fracture test. Several precautions concerning pouring temperature, gas pickup from the mold and chiller, removal of the pouring basin, and quenching of the chill blocks were mentioned earlier in this report. In addition to these precautions the men using the test should be trained in the evaluation of the fractures. The color changes in the fracture are rather difficult for the untrained eye to see. Some experience is necessary before the operator is able to make an accurate prediction of melt quality. To facilitate the examination of the fractures good lighting is essential. A blue fluorescent light has been found fairly satisfactory for this purpose.

The manner in which the fracture test is applied in a commercial foundry will depend upon the foundry practice utilized. There are, of course, many possible ways in which the test could be used. Three methods

will be discussed here.

If the melting is done in a large production unit, a chill block can be cast from the first ladle tapped. This ladle can then be used for a casting in which high properties are not of particular importance. Meanwhile the chill block will have solidified and can be fractured and examined. If the fracture indicates that melt is of high quality, the remainder of the melt can be used for castings where good quality metal is essential. When the test is used in this manner there is no delay involved, while the chill block cools, and is fractured and examined.

If all melts must be of high quality, and degassing procedures are used for the improvement of melt quality, the fracture test can be used as a control on the amount of nitrogen or other inert gas which is bubbled through the melt. When the melt reaches the tapping temperature a chill block can be poured, fractured, and examined. If the fracture indicates that the melt is of high quality, it can be poured immediately. If it is of low quality, the melt can be degassed. The time of degassing necessary can be estimated from the chill block

fracture. At the end of the estimated degassing time period another chill block can be poured and the fracture examined to determine if sufficient improvement in melt quality has occurred.

The fracture test can be used in conjunction with other types of smaller melting units to determine the optimum melting practice or to determine the average quality of the metal cast. A chill block could be cast from each melt before pouring. At a later time the melt quality, as indicated by the fracture, could be correlated with the melting practice used or, if a pit type furnace is used, a correlation might be found between the position of the pot in the furnace and the melt quality. Such a procedure over a long period of time would give an indication of the average melt quality of the metal produced in the foundry and provide data upon which improved practices could be based.

Acknowledgment

The writers wish to thank the American Foundrymen's Society for financial support of the research project upon which this report is based, and the Research Committee of the A.F.S. Brass and Bronze Division for suggestions and guidance in the prosecution of the research work.

We also acknowledge the contributions of the Department of Metal Processing, University of Michigan, which supplied foundry facilities and testing equip-

ment for the project.

The assistance of W. B. Pierce, L. Ruffin, P. L. Jackson, R. Knecht, and Y. P. Telang was of great value in accumulating the data for this report.

Bibliography

1. J. F. Ewing, C. Upthegrove, F. B. Rote, "Research Progress Report on Melt Quality and Fracture Characteristics of 85-5-5-5 Brass," Transactions, A.F.S., vol. 57, p. 433 (1949).

2. F. M. Baker, Jr., F. B. Rote, C. Upthegrove, "Research Progress Report on Melt Quality and Fracture Characteristics of 85-5-5-5 Brass," Transactions, A.F.S., vol. 58, p. 122 (1950).

3. H. H. Fairfield, discussion of Ref. 2 in Transactions, A.F S., vol. 58, p. 131 (1950).

DISCUSSION

fr

bi

bl

th

Chairman: B. N. AMES, New York Naval Shipyard, Brooklyn. Co-Chairman: J. W. BOLTON, The Lunkenheimer Co., Cincinnati.

J. J. MAYER: ¹ Does analysis, be it on the low or the high side of the zinc, or the low or high side of the tin, of those chill block fractures have much influence?

Mr. Shelleng: That is one of the things we have not investigated.

H. C. AHL: ² What is the effect of water quench on these tests? Can these test blocks be water quenched without effect on the fracture itself?

MR. SHELLENG: In the case of the 85-5-5-5 alloy, if quenched too scon, the fracture is apt to be erroneous, due to cracking at or near the chilled surface. However, they can be quenched in about 2 min. after pouring.

MR. AHL: Is there any indication that this fracture test will hold for other alloys? Can there be any correlation between these groups such that we can end up with one set of specifications instead of a different set for every type of alloy?

MR. SHELLENG: Since we have investigated only two alloys that is rather difficult to answer. We feel confident that the change of texture of the fracture from fine to coarse, as the melt quality varies from high to low, very likely will be an indicator of quality on all alloys.

¹ Vice President, Lumen Bearing Co., Buffalo. ² Lt., USNR, Washington, D. C.

ne

ac

th

he

20

Si

elt

re.

pe

en

elt

ne

tv

ta

ch

e-

vi-

of

e-

n.

p-

k.

in

ch

n. n-

se

ot

ed

at

in

ill

en

at

de

G. K. EGGLESTON: 8 As a matter of interest, what procedure was used to produce low quality metal for these tests?

MR. SHELLING: We found that melting in a gas-fired furnace with a reducing flame would usually produce low quality metal. We also produced low quality metal by melting it in an induction furnace and pushing balls of wet clay beneath the surface of the metal, and some steam got into the brass. It was very difficult to produce low quality metal from the 88-8-4 alloy. That seemed to turn out good, in spite of us. However, melting in the gas-fired furnace with a reducing flame usually did the trick.

W. L. Rudin: ⁴ I tried this test in a few foundries on the 81-3-7-9 alloy assuming that what had been done on the 85-5-5-5 alloy would hold for the 81-3-7-9. The results we got were very inconsistent.

Mr. SHELLENG: We have had no experience with the 81-3-7-9

MR. RUDIN: In a discussion with Mr. Halliwell, I find that he experienced the same problems that I did in the 81-3-7-9 alloy. Even though there are only small variations between 85-5-5-5 and 81-3-7-9 alloys, we could not use the data from one and apply them to the other alloy.

G. P. HALLIWELL: ⁸ We tried the fracture test on the 81-3-7-9 alloy. We obtained samples from heats melted under good (oxidizing) conditions and also from heats melted under reducing conditions with glass slag over the metal and under smoky flame. I know of no better way to get gassed metal than that, but on our fractures we could not tell the difference. Yet the properties of the test bars indicated that we had a very poor metal and also a good metal.

We did, however, get some of the distinguishing features of the 85-5-5-5, but they were not as marked as I would have liked to see them.

MR. ROTE: The tests we have run so far at the University of Michigan have been confined to leaded red brass and 88-8-4. We selected those two alloys largely with the idea in mind they would have two distinctly different metals; one with lead in the fracture, and the other in which we have true bronze structure.

I believe from what we have seen in brasses and also some very inconclusive preliminary tests on aluminum alloys, that the main indication in all fracture tests for nonferrous materials using the chilled block is the texture of the fracture. Invariably, the high quality fracture is fine; the low quality fracture poured exactly the same way is coarse.

We might add a note of caution here. Anyone using this test should follow definite and closely controlled procedures. If you are very careful to have a dry chiller and thoroughly baked cores, adequately slotted so that, during the pouring, the basin does not chill the upper surface, most of your problems will be solved. With these you can get adequately good fractures, which can be used to determine melt quality. If you slip in any one of those steps, you may get into trouble, and the slip will mean that you will have poor quality indication. In other words, the test as it now stands is a test for high quality. A fracture indication of high quality means high quality metal. A low quality indication is not necessarily a true indication. It may be due to failure to control the test conditions.

C. L. FREAR: ⁶ I am convinced that the fracture test is one of the best, if not the one best test for melt quality in tin bronzes. I usually pour a small block, in sand, which I break and examine the fracture for over-all color, which should be a blue-gray-yellow color if no oxides are present. By examining the fracture under about 10 diameters magnification interdendrite shrinkage is usually indicated by definite dendrites. If no shrinkage is present, no dendrites will be seen. I was wondering whether the writers have noted the presence or the absence of these dendrites.

MR. ROTE: Chill blocks from high quality melts are dendritic throughout with dendrites going in from the chilled and core surfaces. In low quality metal the structure becomes more diffuse. The upper sections of low quality blocks show very

definite voids. The cooling conditions concentrate the contaminating gas into the upper regions of the chill block. That is the reason that in 85-5-5-5, the depth of the blue-gray structure is the measure of the melt quality.

C. A. ROBECK: 7 We used this fracture test in our foundry on castings that were particularly troublesome to our customers. We found the test more than worthwhile.

CHAIRMAN AMES: I have one observation to make. The authors reported that test bar fractures poured at 2100 F and below yielded poor fracture characteristics, even on good quality metal. That goes back to our old thesis. Solidification molds play a big part. You can take good quality metal and get bad character if the pouring temperature is not right. The authors specify a particular pouring range.

When you deviate from the class of alloys studied by the authors, you should give more study to the optimum pouring rates and crystallization before you begin to evaluate solidification molds rather than the quality of the metal itself.

Mr. ROTE: I would like to add that in 85-5-5-5 we found that the minimum pouring temperature was 2000 F. In 88-8-4, 2100 F was the minimum safe pouring temperature.

I think Mr. Ames' point is very important. When going to another class of alloys, you might as a first step check your effective pouring temperature. I know, again from very inconclusive tests on light alloys that pouring temperature has very important bearing on fracture appearance, and there is a minimum pouring temperature for aluminum as well as copperhase alloys which we have discussed in this report.

base alloys which we have discussed in this report.

D. G. SCHMIDT: 8 How soon do you recommend removing the pouring basin?

MR. ROTE: It should be removed immediately. We, in fact, recommend that you slot the top of the core used for this chill block, so that the metal will not touch the pouring basin. But, for safety, we recommend that you get the basin off at once.

for safety, we recommend that you get the basin off at once.

MR. SCHMIDT: We poured 85-5-5-5 at 2200 F to 2150 F. What is the maximum pouring temperature of 88-8-4?

MR. ROTE: We poured blocks from the 85-5-5-5 at temperatures as high as 2275 F and as low as 1950 F and the only blocks in this series which showed uncertain fracture indications were those poured below 2000 F. 88-8-4 was poured at 2250 F to 2100 F and all blocks showed certain fracture indications.

Mr. Schmidt: Have you noticed any consistent relationship between shrink area and metal quality?

MR. ROTE: There is a consistent relationship, but the shrink area is more consistently related to the pouring temperature. We have not observed shrink in very low quality metal. We have observed it in high quality metal which was poured at low temperatures. We consider the presence of the shrink area an indication that we are coming close to the lower limit of pouring temperature for production of a satisfactory chill block.

We showed shrink areas in some of our chill blocks. That fact that there is a shrink area present does not indicate that the block is not satisfactory or reliable, but it is an indication that you may be approaching trouble.

CHAIRMAN AMES: Have you ever taken a low quality melt and poured it at 2250 F or higher?

Mr. ROTE: Yes, and we got about the same kind of fracture that we got with a lower temperature, with one reservation, that it tends to somewhat better fracture than would be indicated if the block were poured at 2100 F.

CHARMAN AMES: That has been our experience with other tests. You can take a bad quality melt and get better fracture indication than you would by pouring it from excessively high temperatures.

Mr. ROTE: That is one other precaution that would be injected here. A poor quality melt poured at high temperature never has the fracture indications that a high quality melt has. The degree of mottle may vary. The brown or gold mottle may be less on pouring at high temperature, but it will not be eliminated. The texture and color of the overall fracture will not change.

WM. ROMANOFF: DO I understand correctly that if you take a poor quality melt, dropping temperature will not help, and that poor quality melt poured at higher temperature will

³ Engineer, Barnes Mfg. Co., Mansfield, Ohio.

⁴ Chief Metallurgist, Elesco Smelting Corp., Chicago.

⁵ Director of Research, H. Kramer & Co., Chicago.

Bureau of Ships, Department of the Navy, Washington, D. C.

⁷ Vice President, Gibson & Kirk Co., Baltimore.

⁸ H. Kramer & Co., Chicago:

give better results? That is entirely different than we have ever found.

MR. Rote: The fracture appearance of chilled blocks poured at high temperatures are better than those poured at lower temperatures. However, metal allowed to stand in the pot out of the furnace and poured into tensile bars at varying temperatures from 2275 F to 2100 F show better quality when poured at lower temperatures.

MR. HALLIWELL: Was the heating of poor quality melts to high temperatures done under reducing or oxidizing conditions?

Mr. Rote: These melts were heated to high temperature under reducing conditions.

Mr. HALLIWELL: What are the implications of this, since you get a good fracture?

MR. ROTE: We get a better fracture, not a good fracture.

H. J. Roast: ¹⁰ It does seem rather a shock to me since invariably we have cast 85-5-5-5 at much lower temperatures, to wit, 1900 F for pressure test work, and had very good results. These are castings of perhaps not less than ½ in. in cross-section. We would pour at 1900 F.

MR. Rote: The appearance of the fracture will be better, if it is poured at a high temperature. Based on test bar results, the higher pouring temperature gives lower tensile properties than lower pouring temperature. I could not make a statement on pressure tightness. We are talking about two different things.

⁹ Vice President and Technical Superintendent, H. Kramer & Co., Chicago

¹⁰ Consultant, London, Ontario, Canada.

CHILL TESTS AND THE METALLURGY OF GRAY IRON

By D. E. Krause*

ABSTRACT

The design and application of chill tests for the control of the quality of gray iron castings are discussed. Much of the data contained in this paper constitutes a report of the activities of the Chill Test Committee of the American Foundrymen's Society.

The preliminary work done in regard to developing a standardized design of wedge test by the Committee has shown that the sensitivity of the wedge test can be increased by decreasing the acute angle of the wedge without appreciably changing the chill value. Chill tests appear better suited to very soft irons than wedge tests whereas the latter are better adapted to the harder irons.

The influence of the composition of the iron, melting practice and ladle additions on the chilling tendencies of gray iron is discussed. The chilling tendency of gray iron increases progressively with decreases in carbon and silicon content. As melting conditions become more strongly oxidizing, the chilling tendency of the iron increases. Additions of silicon alloys greatly reduce the chilling tendency of gray iron.

The simplicity of the test, the rapidity with which it can be

The simplicity of the test, the rapidity with which it can be made, and the relative ease with which the results can be interpreted should make the chill and wedge tests especially attractive to the smaller shop.

THE VALUE OF CHILL AND WEDGE TESTS as an aid to controlling the quality of gray iron has been recognized by a number of foundrymen for quite some

time. There are still a great number of foundrymen who are either not using these simple, rapid, and inexpensive tests or not using them to their fullest advantage. Since only the simplest type of equipment is needed and the necessary skill to use these tests is acquired readily, these tests should appeal especially to operators of the smaller shops. The rapidity of the test and the low cost of making the test are fully appreciated by the large production shops.

Although chill and wedge tests are a great aid to the gray iron metallurgist, they are sometimes misinterpreted by operators not thoroughly familiar with what the chill test does and does not indicate. Chill and wedge tests are not a substitute for chemical analyses. Inability to always find a close correlation between the amount of chill on the test piece and the carbon and silicon content of the iron has often led some operators to feel that the test was of very little value. The chill test primarily shows carbide stability of iron or the tendency for the iron to solidify as a hard, white iron instead of a soft gray iron.

Chemical composition is only one of the factors which affects carbide forming tendency of iron and hence one should not expect the relation of chill to composition to be necessarily close. Since one of the other important factors affecting the carbide forming tendency of iron is melting conditions, correlation of amount of chill with chemical composition becomes better as control over melting practice is improved. Chill and wedge tests therefore show the combined effect of melting practice and chemical composition. If one of these variables is carefully controlled, the test becomes a more reliable measure of the other variable.

The term "chill tests" is often loosely applied to include two types of tests. In one test a small rectangular casting is poured with one edge against a chill plate while in the other a wedge-shaped piece is cast in a sand core. In the latter type of test piece, the rate of cooling depends on the thickness of the wedge at various distances from the apex of the wedge. Since use of the term in this manner may lead to some confusion, the term "chill test" will be applied only to that test piece which is cast against a chill plate. The term "wedge test" will be used exclusively for the

Columbus, Ohio

ZES

ince ires, ults.

ults, rties nent rent

cago

The work reported in this paper was done in cooperation with the A.F.S. Chill Test Committee whose personnel consists of the following members:

J. L. Brooks, Chairman, Muskegon Piston Ring Co., Sparta, Mich.

H. H. Wilder, Vice-Chairman, Vanadium Corp. of America, Detroit

H. C. Aufderhaar, Electro Metallurgical Div., Union Carbide

and Carbon Co., Chicago
H. C. Bement, Campbell, Wyant & Cannon Foundry Co.

H. C. Bement, Campbell, Wyant & Cannon Foundry Co., Muskegon, Mich.

H. W. Gorman, Allis-Chalmers Mfg. Co., Milwaukee

H. H. Hursen, Association of Manufacturers of Chilled Car Wheels, Chicago

D. E. Krause, Gray Iron Research Institute, Columbus, Ohio

W. W. Levi, Lynchburg Foundry Co., Radford, Va.

H. R. Saurer, Dayton Malleable Iron Co., Dayton, Ohio Wm. J. Sommer, Plainville Casting Co., Plainville, Conn.

H. W. Stuart, U. S. Pipe & Foundry Co., Burlington, N. J. R. E. Wilke, American Radiator & Standard Sanitary Corp.,

Executive Director of Gray Iron Research Institute, Inc.,

test piece cast in the form of a long wedge without the use of a chill plate.

Although the amount of chill in the test piece is one of the most easily measured characteristics of the test piece and is sometimes the only one measured, the gray portion on the fractured test piece also indicates some of the properties of the iron. The appearance of the fracture between the fully white and fully gray zones as well as the depth of chill gives some indication of melting conditions.

Application of Chill and Wedge Tests

The use of chill and wedge tests may be divided into three classifications. The greatest use and the one with which this discussion is primarily concerned is in the application of the chill and wedge tests to the control of the quality of gray iron. A second classification would include the production of various types of castings in which a portion of the surface is to be chilled to a specified depth and the balance of the casting is gray iron. Examples of castings in this application are automotive tappets, rollers, chilled car wheels, chilled iron rolls, and castings to resist abrasion. Quite often, chill tests for this application are highly specialized and in the case of an automotive tappet or other small casting, the casting is often the test piece. Whatever test piece is used, it is highly important that the depth of chill on the test piece be the same as that on the casting or bear a very close relation to it. A third classification which is somewhat limited is found in the production of white iron castings. In this application, a test piece which may be a wedge is cast to determine whether the metal is suitable for making a casting free of primary graphite.

The applications of the chill and wedge test to the control of the quality of gray iron may be roughly subdivided into three groups. The three groups are closely related and differ from each other principally in the extent to which the tests are used. In the first group and the simplest application, the tests are merely used to indicate the suitability of the iron for the type of castings being made. The primary aim is to ensure production of castings free of chilled or hard edges in fairly light sections. If the chill is high, the iron is used for heavier castings. No particular attempt is made to change the chilling tendency of the iron through ladle additions.

Applications of the chill and wedge tests in the second group are possibly more numerous than in the other two groups. In this group, the chill or wedge test is used to determine the quality of the iron, as in the first group, but in addition either graphitizers or carbide stabilizers are added to the ladle to produce an iron of more uniform chilling tendencies. Such corrective measures do not guarantee an iron of constant properties but do however bring the properties within closer limits than would be the case if the corrective measures had not been taken. It is erroneous to assume that irons showing the same chilling tendencies as a result of ladle additions will have identical mechanical properties.

The application of the chill and wedge tests in the third group is a very important and specialized one. In the production of high-strength processed irons through ladle inoculation of a properly melted base iron, chill and wedge tests, most frequently the latter, are indispensable. Since some of the higher strength irons produced by this method have a rather high chilling tendency as-melted, the wedge test is better adapted to this application than the chill test. Considerable experience is needed to fully utilize the information obtained from tests for this application.

Work of A.F.S. Chill Test Committee

Survey of Chill Test Practices—During the past years, this Committee has made a survey of various types of chill and wedge tests in use with the thought in mind of establishing some standard sizes and shapes. Virtually each foundryman using a chill or wedge test developed one which suited his needs without giving any thought of correlating his information with that obtained by another foundryman. The Committee felt that it would be helpful to have some degree of standardization of these test pieces for the benefit of foundries who are not using such tests as well as to provide for more ready transfer of information among gray iron foundrymen. In practically all cases, standardized design of chill or wedge tests could be adopted without any difficulty.

is

iı

SE

of

de

te

fo

of

Po

Ra

len

fre

to

It was found that the various types of chill and wedge tests in use could be divided roughly into three groups. Tests in the first two groups were used primarily as a means of evaluating the chilling tendency of the iron. These tests were either of the type cast against a chill plate or the wedge type in which the chill is induced by progressively decreasing the thickness of the casting. A third type of test piece had an enlarged section which in some cases was tested for Brinell hardness and in other cases the fracture was examined for the grain structure.

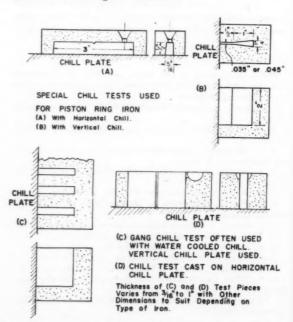


Fig. 1-Representative types of chill tests used by gray iron foundries.

ıt

d

1

n

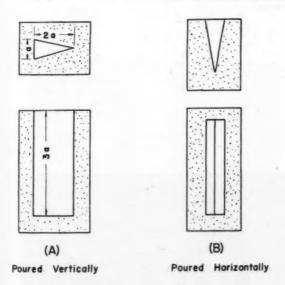
d

The Chill Tests—Test pieces cast against a chill were found to be of many shapes and sizes. Some of the tests were cast with the chill plate horizontal and the core resting on the plate while others were cast with the chill plate in a vertical plane. The chill plates of gray iron, steel, copper, or graphite, gave satisfactory results.

Examples of representative chill test types are shown in Fig. 1. Types (A) and (B) are used for a soft, hypereutectic iron having a very low tendency to chill such as piston ring iron. Type (B) is used with a water-cooled chill plate. Type (C) chill tests are widely used in production foundries where tests are poured at very close intervals. The chill plates are often water cooled. The same type of test piece is sometimes cast in separate cores. Type (D) chill test is most frequently used in jobbing shops and is made in a range of sizes to suit various types of iron. A small pouring basin or button is provided at one end of the test piece to avoid pouring iron directly onto the chill.

Wedge Tests—Wedge tests are most frequently cast in individual cores although gang cores are also used. Most wedges are cast vertically although a few foundries cast wedges in the horizontal position. Representative examples of these wedges are shown in Fig. 2. Although the ratio of the width of the base of the wedge to the height is varied, a ratio giving an angle of approximately 30 degrees is most frequently used. The wedge poured horizontally is used to a very limited extent.

Special Chill Tests—Quite often a chill test piece is designed to include a heavier section which may be tested for Brinell hardness or the fracture examined for grain structure. Such test pieces give an indication of the grain structure of the iron in heavier sections.



Ratio of base (a) to height (2a) of wedge and to length (3a) of wedge (A) as shown is used most frequently. Dimension (a) will vary from $\frac{3}{8}$ inch to over $\frac{1}{2}$ inches.

Fig. 2-Representative types of wedge test pieces.

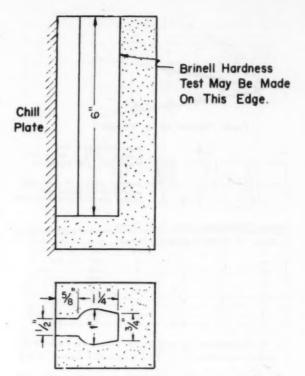


Fig. 3-Representative chill test of the keyhole type.

An example of such a test, often called a keyhole test piece, as used in a large automotive foundry for cylinder block work is shown in Fig. 3. This test piece is often used in conjunction with a wedge test. The wedge test is used as a rapid means of checking the iron and the keyhole test piece may be used at less frequent intervals as a check on the general structure of the iron. A record is usually made of the depth of chill on the keyhole test piece while the results of the wedge test are used as a guide and may not be recorded.

Chill and Wedge Tests

As a result of reviewing control tests for gray iron, Sub-Committee T.S. 6 of the Technical Council of the Institute of British Foundrymen developed recommendations for chill and wedge tests. The series of wedge tests shown in Fig. 4 was developed. The reason given for making the angle at the apex of the wedge sharper as the wedge becomes larger was to obtain the same degree of chill on wedges of various sizes for the same iron. The amount of chill is determined by measuring the width of the wedge in millimeters at the junction of the white iron zone and the gray zone of the wedge.

The reason for the relatively large number of wedges is that as small a wedge as possible for a given iron may be selected, thus taking advantage of the rapidity of the test. The wedge must not be so small, however, that the chill exceeds one-half the width of the base.

It was found that for soft irons having a very low tendency to chill a test piece cast against a chill was

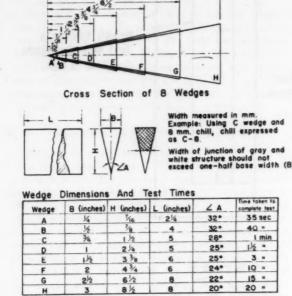


Fig. 4—Dimensions of wedge test pieces recommended by Sub-Committee T. S. 6, Technical Council, I.B.F.

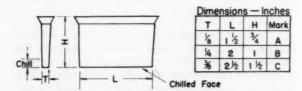


Fig. 5-Chill test pieces recommended by Sub-Committee T. S. 6, Technical Council, I.B.F.

more satisfactory than the wedge. Recommended dimensions for a chill test piece are shown in Fig. 5. A series of tests was made to determine the relation between the chill in the wedge and chill tests. The results indicated that the depth of chill on the chill test was approximately double the wedge value on the wedge test. It will be shown later that this relation may vary with the chilling tendency of the iron.

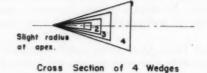
Development of Standard Wedge Tests

Discussion by members of the A.F.S. Chill Test Committee of the relative merits of chill tests using a chill plate and wedge tests made without such a plate led to the conclusion that the wedge test would be preferable. Inspection of a large number of chill test practices revealed considerable carelessness on the part of the operator in keeping the chill test plate in satisfactory condition. Unless the test plate is kept free of scale, dirt, and burned core oil, a low chill value will be obtained. The use of a wedge test would eliminate this variable in measuring the chill-forming tendencies of gray iron.

The A.F.S. Chill Test Committee in reviewing the various designs for test wedges felt that the wedge test

might be made suitable for the softer irons by making the angle at the apex of the smaller wedges more acute than for the larger wedges. It was known that the more common wedge designs with a wedge angle of about 30 degrees made it difficult to measure the very low chills obtained for the softer irons. It was also felt desirable for the initial work to use as few wedge sizes as possible. Although five wedges were originally considered, these were subsequently reduced to four. The dimensions of the four wedges under consideration are shown in Fig. 6. The fifth wedge had dimensions intermediate between the largest (W4) and second largest (W3) of the present series of four wedges. It should be noted that the angle of the wedge becomes more blunt as the size of the wedge increases. This is directly opposed to the principle followed by the Institute of British Foundrymen.

Fractures of the four A.F.S. Chill Test Committee wedges together with a fracture of a chill test from the same ladle of iron are shown in Fig. 7. The influence of wedge test size and design on the amount of chill is quite evident. Although not too apparent from the illustration, the depth of chill from the edge of the wedge for the two smaller wedges is about the same as the depth of chill on the chill test. The three smaller wedges have a very small radius at the chilled



Vedge No.	В	н	LA	Length	
WI	0.20	1.00	11.5*	4	
W 2	0.40	1.25	18 *	4	W - F
W3	0.75	1.50	28 °	4	LANT
W'4	1.25	2.00	34.5*	4	

ir

th th

de

Fig. 6—Test wedges under investigation by A.F.S. Chill Test Committee.



Fig. 7—Fractures of four A.F.S. Chill Test Committees wedges and a chill test showing method of measuring chill.

re

at

le

a

M

re

ed

eı

(c)

11

16

n-

ee

e

ce

e

le

1e

ee

ill

edge. Through an error in the drawings for the largest wedge (W4) shown in Fig. 7, it has a much larger radius at the apex than intended. The rather large radius may account for the absence of even a small amount of chill.

Wedge tests with fairly sharp edges and with rather small included angles often have an irregular edge as shown in Fig. 8. This tendency was pronounced for the W1 and W2 wedges but not evident in the W3 wedge. The irregular edge is more likely a result of the small included angle and sharpness of the edge rather than the small size of the wedge. Because of the possibility of sand grains sticking in the sharp edge of the test core and the test piece having an irregular edge by virtue of not filling out, the Committee felt it desirable to express the chill in $\frac{1}{32}$ in measured across the wedge at the junction of the chilled area and the gray area. The method of measurement is shown in Fig. 7. This measurement will be termed "wedge value."

The four wedge designs under investigation by the A.F.S. Chill Test Committee as shown in Figs. 6 and 7 may require some modification in order to avoid the irregular edges. This is particularly true of the two smaller wedges W1 and W2.

Results Obtained with A.F.S. Chill Test Committee Wedge Tests

Wedge test pieces of the design described were cast together with chill test pieces from various grades of iron to determine the relation between the amount of chill and the design of the test piece. The relation of the depth of chill on a chill test piece of \(\frac{3}{6}\)-in. thickness and the wedge value obtained with the first three wedges for 22 tests are shown in Fig. 9. Although the relation is an approximate one only, the depth of chill on a chill test is about 3.3 times the wedge value. The relation of the chill of the W1 and W2 wedges to that on the W3 wedge is shown in Fig.

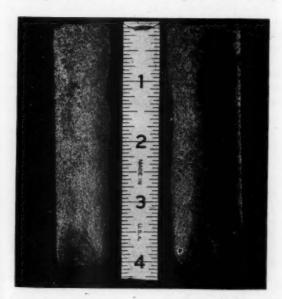


Fig. 8-Irregular edges often found in the W1 and W2 A.F.S. Chill Test Committee wedges.

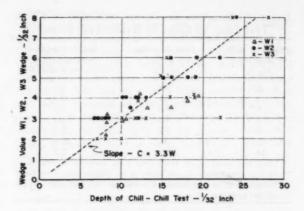


Fig. 9-Relation of depth of chill on chill test to chill on wedges W1, W2, and W3.

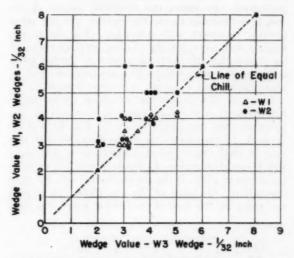


Fig. 10-Relation of chill on W3 wedge test to chill on W1 and W2 wedges.

10. For some reason which is obscure at the present time, the W2 wedge generally shows a higher wedge value than the W1 and W3 wedges for the same iron as indicated in Fig. 10. The irons from which the data in Figs. 9 and 10 were obtained came from a variety of sources and varied in carbon content from 3.10 to 4.07 per cent and in silicon content from 0.93 to 2.50 per cent. All of the irons were melted in a cupola and most of them were unalloyed. Some irons were inoculated and others were not.

Results based on a large number of tests of iron used for small castings indicate that the ratio of depth of chill in a chill test piece to the chill in a wedge varies with the chilling tendency of the iron. The relation between the two types of tests is shown in Table 1.

Procedure for Making Chill and Wedge Tests

Good practice is required if dependable information from chill and wedge tests is to be obtained consistently. A satisfactory practice is easy to acquire and the results are worth the little extra attention given to various details in making these tests.

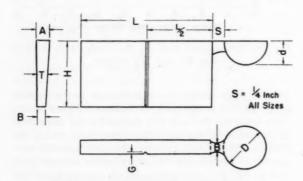
TABLE 1—RELATION OF DEPTH OF CHILL IN CHILL TEST PIECE TO CHILL IN WEDGE TEST PIECE.³

Depth of chill in chill test,* 1/32 in.	Chill in wedge test piece,** 1/32 in.	Ratio of depth of chill in chill test piece to wedge value
2	2.0	1.0
4	2.7	1.5
6	3.2	1.9
8	3.7	2.2
10	4.2	2.4
12	4.6	2.6
14	5.1	2.7

Chill test piece ½-in. thick and 3 in. long on chilled face.
 Wedge dimensions, 7/16-in. base, 7/8-in. height, and 3 in. long. Wedge chill is measured as width at the junction of the chilled and gray zones.

Selection of Test Piece Size—One of the inherent advantages of the chill and wedge tests for control purposes is the quickness with which the test may be performed. To retain this desirable feature of the test, it is desirable to use the smallest size test piece possible consistent with reliability. The use of too small a test piece for the chilling characteristics of the iron will result in erratic values. If the test piece is too small, only a white and a mottled zone will be obtained whereas a definite gray zone is desired.

Based on the experience in about 20 foundries melting iron over a wide range in composition, the series of chill test pieces shown in Fig. 11 was developed. It was found that it was desirable with this type of chill test to select a thickness of test piece which would be within the limits of one-half to twice the depth of chill. If the depth of chill was less than one-half the thickness of the test piece, too large a test piece for the type of iron being tested was used



	Re	ecom	nende	ed E	imen	sions	- 1	nches	3
Chill Test No.	Т	A	В	н	L	D	d	G	Recommended Chill Depth Range 32 ^{nd*}
IC	3/6	1/4	1/8	14	2/2	3/4	1/2	1/32	3 - 12
2 C	1/4	\$	3.	1/2	3	1/8	1/2	1/32	4 - 16
3 C	3/8	7/6	16	13/4	3/2	7/8	1/2	1/16	6 - 24
4C	1/2	%	7/6	2	4	1	3/8	16	8 - 32
5 C	3/4	13/6	1/16	2½	5	1	%	3/32	12 - 48
									# T.

Casting to be made in dry sand core.

Allow 4 inch of sand on all sides.

Fig. 11-Suggested dimensions for Type C chill test.

with a subsequent sacrifice in speed of testing. If the depth of chill exceeded twice the thickness of the test piece, the depth of chill became erratic and the line of demarcation between the gray and white areas of the fracture became less distinct.

In selecting a wedge test piece, the smallest test piece adaptable to the type of iron being melted should be used. The chill on a wedge type of test piece should not exceed one-half the width of the base. This is illustrated in Figs. 4 and 6. For metallurgical control of gray irons having a chill on the wedge consistently less than $\frac{3}{32}$ in., the chill type of test piece would be more suitable. For irons having wedge chill values consistently in excess of $\frac{4}{32}$ in., the wedge test would be preferable to the chill test as a smaller test piece could be used with corresponding increase in speed of testing.

Test Piece Cores—Although the chill or wedge tests could be poured in green sand molds, the use of cores is preferred. Dietert and Woodliff⁴ found the properties of green sand to have an appreciable effect on the depth of chill of gray iron and for this reason a dry sand core would be preferred to a green sand mold. As chill and wedge tests become smaller, a finer core sand should be used. Cores should be well baked and an excess of binder avoided. The core for a single test piece need not extend more than ¾ in. on all sides of the test piece for the smaller pieces and 1 in. for the larger pieces.

iı

Chill Plates—Clean chill test plates are very important if dependable results are to be obtained from a chill test. Selection of the chill plate material depends somewhat on the type of application. Gray iron, steel, copper, or graphite will give satisfactory results if properly used. Graphite has the advantage of resisting "burning-in" when a test is poured rather hot. It also has a lesser tendency to pick up burned core oil deposits than the metal plates. If chill tests are not poured any more frequently than one in 10 minutes, water cooling is not necessary. Several plates can be provided and used in rotation which will prevent them from being overheated.

The temperature of the chill plate has comparatively little effect on the depth of chill for plate temperatures up to approximately 500 F. The tendency for the test pieces to burn onto the chill plate

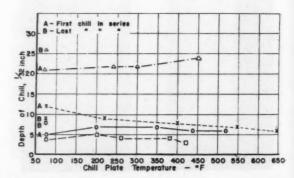


Fig. 12-Influence of chill plate temperature on depth of chill of 3/2-in. thick test piece.

RON

the

test

inc

01

est

ted

est

the

tal-

the

of

ng

he

sa

ng

sts

res

p-

a

nd

a

ell

OI.

n.

nd

n-

m

e.

ay

ge

ed

ts

10

es

e-

te

d.

te

increases however, with an increase in plate temperature. The influence of chill plate temperature on depth of chill is shown in Fig. 12. A test was poured on a plate at room temperature at the start and end of each series. This procedure accounts for the two tests at plate temperatures of 80 F.

Water cooling may be used with either horizontal or vertical chill test plates. When water cooling is used, it is more desirable to use a water reservoir than circulating water. The use of circulating water for cooling, particularly in summer, may reduce the temperature of the plate below the dew point of the air and difficulty with moisture condensing on the plate will be experienced. Sand and other loose material will stick to the wet plate causing a low chill.

Metal chill test plates must be kept clean. Condensed core oil can be removed quite easily with coarse steel wool if it is done immediately after the test piece is poured. If the plate becomes fire checked, it should be machined off or replaced.

Sampling Ladles and Sampling Practice

A clean sampling ladle of proper construction is just as important as other details of chill testing technique. For sampling of an iron with a low chilling tendency using a small test piece, a malleable iron melting ladle coated with a thin layer of fireclay or other refractory will prove satisfactory. With a little skill and dexterity, no difficulty will be experienced in cutting out a sample from the stream of iron from the cupola spout. This method of obtaining a sample as practiced in a piston ring foundry is shown in Fig. 13. The use of a thinly-lined melting ladle as illustrated is not suitable for obtaining larger samples of



(Courtesy of Muskegon Piston Ring Co., Sparta, Mich.)

Fig. 13—Method of obtaining sample of iron for chill test from stream of iron.

a harder iron.

For pouring larger wedge and chill tests, a small ladle made of a piece of 4 to 5-in. diameter steel tubing about 5 in. long with a bottom welded-in and a section of 3/4-in. pipe welded-on for a handle and lined with 1/4 to 1/2 in. of a good refractory works very well. The use of a 50-lb hand ladle to catch a very small quantity of iron for a chill or wedge test is generally not satisfactory as the iron cools off too quickly. A sampling ladle should not be used when it has a skull or heel of solid iron in it. The ladle should also be free of scale and small oxidized pieces of iron.

The technique of cutting a sample out of a stream of iron can be applied to any size cupola. If a lined steel ladle as described above is used, it is important that the lining at the top edge be brought to a fairly sharp rim to avoid splashing when the ladle is held in the edge of the stream of iron. No difficulty will be experienced in getting a sample if the ladle is held at either edge of the stream rather than on the top or bottom of the stream.

A sample of iron for the chill test is often obtained by dipping it out of a large ladle. If this is done, the sampling ladle should be periodically dipped in a coating of blacking to protect the steel from the molten iron. Since iron in the ladle is frequently sampled after a ladle addition, it is important that the sample obtained be representative of the iron in the ladle. Unless inoculants go into solution thoroughly, there is apt to be a concentration of silicon in the iron at the top of the ladle. The iron should either be stirred in the ladle or the sampling ladle dipped far enough below the surface to get a representative sample.

Sampling of the stream of iron at the spout must be done with care in order to obtain a representative sample. The easiest sample to get is one from a continuously-tapped cupola with rear slagging and a full well of iron. In such an operation, the iron is fairly well mixed as it leaves the spout. For an intermittent-tap cupola, the sample should probably be taken when one-third of the tap has been taken out. If the sample is taken too soon, it may show a lower chill than representative for the entire tap. It is, of course, desirable to obtain a sample as soon as possible so that if some ladle treatment is necessary, it can be made in time. Obtaining a representative sample of iron from a front-slagging cupola is rather difficult. This is particularly true if the charge contains an appreciable amount of steel and highly concentrated silicon alloys in which case a true sample is almost impossible to obtain. A forehearth will, of course, smooth out some of the fluctuations in iron composition likely to be encountered with a front slagging cupola and will aid materially in getting a representative sample.

Pouring and Cooling of Test Piece

Although normal variations in pouring temperature do not markedly affect the depth of chill, extremes in pouring temperature should be avoided. If the iron being sampled is about 2800 F, the iron should be allowed to cool to 2650 to 2500 F or thereabouts before pouring the test piece. Low carbon equivalent irons are more sensitive to low-pouring

temperature than softer irons.

If the principal interest is in the depth of chill, it is possible to quench the test piece in water as soon as it is solid. If the test piece is quenched too soon, the interior of the test piece may be still liquid and the quenched piece may show spots of white iron in the center of the specimen. If the test piece is quenched in water as soon as it is fully solid, the fracture of the test piece will have a lighter color than if it was allowed to cool in the core to below approximately 1300 F. The depth of chill is not affected by quenching from a high temperature. The appearance of test pieces quenched in water as compared with similar test pieces allowed to cool in the core is shown in Fig. 14. The left piece of each pair was water quenched from about 1700 to 1800 F while the right one was left to cool in the core. The few light-colored spots in the chill test at the very left of Fig. 14 are white iron areas resulting from quenching too soon.



Fig. 14—Effect of water quenching of chill and wedge test pieces on appearance of the fracture. Left piece of each pair was quenched from about 1700 F. Right piece of each pair was allowed to cool in the core.

T.C.-3.39%; Si-1.74%.

If time is an important factor, water quenching should be used, and again, the smallest test piece suitable for the type of iron being tested should be selected.

Measurement of Chill

The amount of chill whether measured as depth on a chill test piece or as the width of the wedge at the junction of the chilled zone with the gray zone should be expressed in ½32-in. Some irons do not have a very sharp demarcation between the gray area and the white area. The intermediate mottled area varies considerably in appearance and amount. Where a fairly broad mottled zone is obtained, the depth of chill is sometimes measured to the point where only a few gray spots are evident. In other cases, two measurements are made. The first measurement being made to the first gray speck in the chill and is termed clear chill while the second measurement is made to the last white speck in the gray part of the fracture.

Relation of Chilling Tendency to Metallurgy of the Iron

The amount of chill in the test piece is directly related to the tendency to freeze white. Since composition of the iron is only one of the factors which influences the carbide forming tendency of the iron, the relation between the amount of chill and the composition is not necessarily close. Other factors which affect the carbide forming tendency of the iron are: nature of the materials which make up the charge, melting conditions, ladle treatment, and cooling rate. In foundry practice, cooling rate is related to section size. The chill test and the wedge test introduce a variable cooling rate through the design of the test pieces.

In the chill test, the cooling rate is greatest at the chill plate face of the test piece and gradually becomes less as the distance from the chill plate increases. In the wedge test, the variable cooling rate is obtained through a continuously increasing section size from the apex of the wedge to its base. The cooling rate is therefore highest at the apex of the wedge and gradually decreases as the distance from the apex of the wedge is increased. The cooling rate at the junction of the white iron zone and the gray iron zone is that rate which is just sufficient to counteract the graphitizing forces in the iron. This cooling rate may be termed a critical cooling rate as faster rates will result in the formation of white iron while slower cooling rates will result in gray iron.

Effect of Composition of Iron on Amount of Chill

Although the composition of the iron is only one of the factors which affects the carbide forming tendency of the iron, it is an important one. If the other factors are fairly well controlled, the relation between composition and amount of chill is fairly good. Carbon and silicon are two of the more important elements in gray iron and are present in the largest amounts. The amount of these elements present has considerable influence on the mechanical properties of the iron as well as on the amount of chill observed in a chill or wedge test piece.

A number of investigators have reported the effect of silicon and carbon on the chilling tendency of gray iron. The effect of silicon content of gray iron melted in an induction furnace under closely controlled conditions on the depth of chill as reported by Burgess. and Bishop⁵ is shown in Fig. 15. The marked effect of silicon on depth of chill, particularly for the lower carbon irons is quite evident. The reduction in depth of chill effected through a late addition of 0.25 per cent silicon-manganese-zirconium alloy to a 3.05 per cent carbon iron containing 1.5 per cent silicon is as great as an increase of 0.6 per cent silicon in the charge. The inoculated iron also shows a lesser change in depth of chill with changes in silicon content than does the base iron. The influence of carbon on depth of chill may also be determined from Fig. 15. It should be noted that the effect of carbon on the depth of chill varies with the silicon content, at low-silicon contents the influence of carbon on depth of chill is much greater than at higher-silicon contents.

Since the effects of silicon and carbon are similar in regard to their influence on the chilling tendency of gray iron, the expression, carbon equivalent (carbon plus ½ silicon) may be used to combine the effect of the two elements if the ratio of carbon to

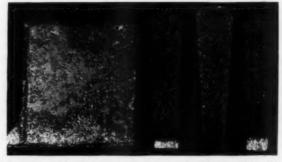
1.0

1.2

Fig. 15-Relation of depth of chill to carbon and silicon content of untreated irons and inoculated irons.⁵



Fig. 16—Iron showing tendency towards inverse chill. Melted in basic-lined cupola. T.C.-4.07; Si-0.93.



2.6

2.2

2.0

per cent

Fracture at left and outer surface of test piece show the appearance of iron which has rejected carbon (kish). T.C.—4.10; Si—2.01.

Fracture in center is of a very soft iron to which 0.35 Si was added in the ladle showing complete absence of chill. T.C.—3.85; Si—2.04.

Fracture at right is of an iron which not quite reached the point of rejecting kish. T.C.—3.87; Si—1.92.

Fig. 17-Chill tests of high-carbon irons.

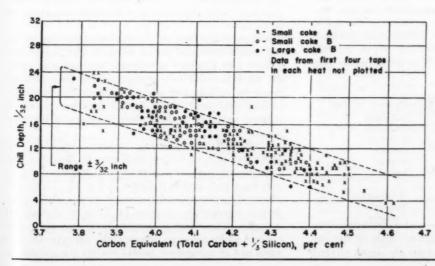


Fig. 18 — Relation of chill depth to carbon equivalent of irons melted with three test cokes in 10-in. cupola with coke ratio of 7.5/1 and blast rate of 12.5 lb per min.

hich are: arge, rate, tion varieces.

RON

the

com-

mes
In
ned
com
e is
iduthe
cion
hat

the

be repol-

one tenher een Carelegest has

ect ray ted oness ect

in

as he ige an oth

oth per

is ar cy

th

on

efto silicon remains within certain limits. It has been found that irons with a high carbon to silicon ratio usually show a greater chilling tendency than irons with a lower ratio of carbon to silicon for the same carbon equivalent. Irons with rather high carbon to silicon ratios have a tendency to show an inverse chill. An example of such a chill is shown in Fig. 16.

In addition to carbon affecting the depth of chill, it also affects the appearance of the chilled zone as well as the gray zone. High carbon irons often develop a coarse columnar structure in the chilled area whereas lower carbon irons have a finer chill structure. The gray area also becomes progressively darker as the carbon is increased. If the carbon content exceeds the limit of solubility, kish may be formed. The presence of kish is usually evident on the outside of the test piece as well as in the fracture and an example of such an iron is shown in Fig. 17. Although very soft irons usually show some chilling tendency, inoculation of such irons can produce an iron without any chilling tendency as shown in the center fracture of Fig. 17.

The relation between the carbon equivalent of iron and the chilling tendency becomes better as other factors are more closely controlled. The data plotted in Fig. 18 were obtained from irons melted in an experimental cupola with identical melting stock for all the irons. The data represent 10 different melting conditions. The variation in carbon equivalent came through changes in blast temperature, blast moisture, and coke properties. If the chill test values for any one particular set of the operating conditions are averaged, the relation between carbon equivalent and depth of chill is considerably better as shown in Fig. 19. The greater range for the values in Fig. 18 could be attributed to minor changes in melting conditions.

Effect of Other Elements on the Chilling Tendency of Gray Iron

Although the cupola operator is most concerned with the control of the carbon and silicon content of the iron through the use of chill and wedge tests, other elements do affect the chilling tendency of gray iron. A discussion of elements which reduce the chilling tendency and elements which increase the depth of chill in gray iron may be found in the A.F.S. Cast Metals Handbook.⁶ For convenience, elements which either reduce or increase the depth of chill when present in amounts likely to be found in gray iron are as follows:

Reduce depth of chill	Increase depth of chill
Aluminum	Tellurium
Titanium	Chromium
Silicon	Antimony
Nickel	Bismuth
Copper	Sulphur
Cobalt	Selenium
Phosphorus	Tin
Zirconium	Molybdenum
	Manganese
	Tungsten

Under certain melting conditions, some of the elements which may normally increase the depth of chill

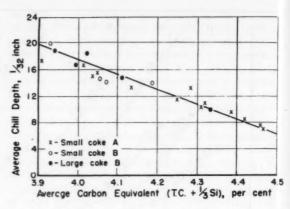


Fig. 19—Effect of average composition of iron upon average chill depth of irons melted with three test cokes in 10-in. cupola with coke ratio of 7.5/1 and blast rate of 12.5 lb per min.

may actually reduce the depth of chill when added in small amounts. This is true of sulphur and boron. Depending on the sulphur content, an increase in manganese may also reduce the depth of chill.

Influence of Melting Conditions on Depth of Chill

Two melting conditions which tend to increase the depth of chill are superheating and melting under oxidizing conditions. The effect of superheating on the depth of chill is especially marked when melting in an induction furnace where contact with hot coke is absent. Massari and Lindsay⁷ reported an increase in chilling tendency in a car wheel iron as the melting temperature was increased. Gray irons of higher silicon content are similarly affected.

ti

ti

th

tio

pr

50

Any change in melting practice which tends to make conditions more oxidizing will result in an increase in chill. Irons melted under strongly oxidizing atmospheres or slags will tend to show in addition to a greater depth of chill, a coarser mottled zone and may also show a rather sooty fracture in the gray zone of the test piece. Such irons will have lower strength and toughness than expected from the chemical composition. Such irons are also markedly benefited by inoculation.

Although an increase in the moisture content of the cupola blast was shown by Eash and Smith⁸ and Krause and Lownie⁹ to increase the depth of chill, the increase could be explained on the basis of a decrease in carbon and silicon content. For the same carbon equivalent, there was no consistent trend in chilling tendency as the moisture content of the blast was varied.

Under certain oxidizing melting conditions, an abnormal chill or inverse chill may be obtained. An example of how very high-carbon, low-silicon irons may have such a chill is shown in Fig. 15. Zapffe and Phebus¹⁰ attributed some cases of inverse chill to hydrogen segregation. Auguste Le Thomas¹¹-found that a high-sulphur content coupled with low-manganese content might cause an inverse chill. He reported that additions of manganese would correct the condition.

RON

4.5

on

test

ind

in

on.

in

1

the

der

on

ing

oke

ase

elt-

er

to

in-

ng

to

nd

ne

th

m-

by

of

nd

ill,

de-

ne

in

ast

an

An

ns

nd

ıy-

at

ese

at

m.

Control of Depth of Chill Through Ladle Additions

Ladle additions to control the chilling tendency of the iron are used extensively. The additions normally used for reducing the depth of chill are: graphite, ferrosilicon, and a number of proprietary graphitizing alloys which may contain in addition to more than 50 per cent silicon one or more of the following elements: aluminum, zirconium, calcium, manganese, and titanium. The most common means of increasing the depth of chill is by ferrochromium additions. Quite often, complex iron-chromium alloys are used.

In regard to the use of graphitizers as ladle additions to iron which may show rather wide variations in depth of chill as it leaves the spout, it is often wrongly assumed that obtaining a uniform depth of chill after the addition of varying amounts of graphitizers will assure an iron of identical properties. Since irons having some of their silicon content introduced as a late addition will show a lower depth of chill than irons melted down to the same silicon content, the depth of chill is no longer indicative of the carbon equivalent value of the iron. The use of chromium or silicon additions to compensate for minor variations in melting practice and analysis is a proper and good one. Experience in any one shop may show that iron having less than a certain depth of chill as it leaves the spout may be too soft for certain types of castings. The addition of the proper amount of chromium to such an iron will improve the quality of the casting. On the other hand, an iron which shows a higher than normal amount of chill may result in hard edges or corners on some castings and the addition of a graphitizing alloy will assure freedom from machining difficulties. The use of over 5 lb of aluminum containing inoculant per ton of iron should be done cautiously as pinholes or shrinkage defects may be encountered. This tendency is more pronounced in soft irons. There are instances where even lesser amounts of such inoculants may cause difficulties.

Relation of Depth of Chill to Hardness and Tensile Strength

Since a fair relation exists between chilling tendency and carbon equivalent providing melting practice and ladle practice are fairly uniform and since there is a fairly good relation between carbon equivalent and tensile strength, it has been found that there is a relation between tensile strength and chilling tendency. A similar relation has been found between Brinell hardness of gray iron and depth of chill. In some cases, the relation is very good while in others the relation is quite broad.

Processed Irons—A special and important application of chill and wedge tests is in the manufacture of processed irons. It would be very difficult to produce a processed iron of uniform and predictable properties without extensive use of either chill or wedge tests. Since greatest interest in these irons is in the range of 50,000 psi tensile strength in 1.2-in. arbitration bars, the depth of the chill of the iron as melted is rather high so that the wedge test is better suited and more widely used than the chill test. One manner in which wedge tests may be used for controlling the properties of a processed iron has been described by Smalley and Reece. 12 The use of wedge tests for controlling the manufacture of processed iron is such a highly specialized one that it must be considered outside of the scope of this paper. The application should be of current interest as it provides a means of producing high strength irons without the use of scarce alloys.

Conclusions

Chill and wedge tests are among the simplest and most economical control tests available to the gray iron foundryman. Because of the ease with which the test may be made and the rapidity with which the results are obtained, the tests should be of special interest to the operators of smaller foundries.

Chill and wedge tests are not a substitute for chemical analysis although they will greatly extend the usefulness of chemical analyses. Because the depth of chill is influenced by melting conditions as well as chemical composition of the iron, the relation between the depth of chill and properties of the iron must be worked out for each particular melting practice. Such a relation can be established in a relatively short time and the results obtained are very much worth the effort required.

As in the conduct of any other type of test, good testing technique must be established and adhered to or otherwise the results will be misleading and of relatively little value.

Chill tests using a chill plate are better adapted than wedge tests to the control of the softer irons in the range of 4.20 per cent carbon equivalent or over. Wedge or chill tests may be used in the range of 4.0 to 4.2 per cent carbon equivalent while the use of wedge tests is much preferred for irons having carbon equivalents of less than 4.0 per cent.

Chill and wedge tests are a necessity when using high percentages of scrap in the cupola charge. Such tests will quickly indicate the presence of undesirable elements in the iron and often corrective measures can be taken at once to make the iron acceptable for use in the castings.

Acknowledgments

The author wishes to acknowledge the assistance of various members of the A.F.S. Chill Test Committee in supplying test cores and helpful information. He also wishes to thank H. W. Lownie, Jr. of Battelle Memorial Institute who supplied much useful information and E. H. Nielsen of the Gray Iron Research Institute who assisted in preparation of the material required for this paper.

Bibliography

- Handbook of Cupola Operation, American Foundrymen's Society 1946, pp. 65-68.
- 2. Sub-Committee TS 6 of Technical Council, Institute of British Foundrymen, "Control Tests for Grey Cast Iron," *Proceedings*, Institute of British Foundrymen, vol XXXIX, pp. A28-A51 (1945-1946).
- 3. H. W. Lownie, Jr., Private communication.
- H. W. Dietert and E. E. Woodliff, Transactions, American Foundrymen's Society, vol. 48, pp. 393-424 (1940).

5. C. O. Burgess and R. W. Bishop, "Alterations in Cast Iron Properties Accompanying Use of a Strong Inoculant of the Silicon-Managanese-Zirconium Type," TRANSACTIONS, American Foundrymen's Society, vol. 52, pp. 671-710 (1944).

6. CAST METALS HANDBOOK, American Foundrymen's Society,

1944, pp. 367, 546.

7. S. C. Massari and R. W. Lindsay, "Some Factors Influencing the Graphitizing Behavior of Cast Iron," Transactions, American Foundrymen's Society, vol. 49, pp. 953-991 (1941).

8. J. T. Eash and R. E. Smith, "Effects of Moisture and Preheat of the Cupola Blast on Gray Cast Iron," Transactions, American Foundrymen's Society, vol. 49, pp. 1022-1054 (1941).

9. D. E. Krause and H. W. Lownie, Jr., "Blast Humidity as a Factor in Cupola Operation," Transactions, American Found-

rymen's Society, vol. 57, pp. 11-132 (1949).

10. C. A. Zapffe and R. L. Phebus, "Cause and Cure of Inverse Chill and Hard Spots in Cast Iron," *Transactions*, American Society for Metals, vol. 41, pp. 259-284 (1949).

can Society for Metals, vol. 41, pp. 259-284 (1949).

11. Auguste Le Thomas, "Abnormalities in Depth of Chill of Cast Iron and their Relation to Manufacture Uniformity," La Founderie Belge, vol. 3, May-June 1934, pp. 58-73.

12. Oliver Smalley and Herbert A. Reece, "Process for Producing Cast Iron Castings," U. S. Patent 2,371,654, March 20, 1945.

DISCUSSION

Chairman: J. L. Brooks, Muskegon Piston Ring Cc., Sparta, Mich.

Co-Chairman: H. H. WILDER, Vanadium Corp. of America, Detroit.

T. E. EAGAN: ¹ The author should be congratulated for presenting such a good paper. This chill test is a real tool if it is used properly, and the author has explained how to use it. However, I want to caution you about quenching the wedge test in water. What the author said is true, you can quench chills of these little wedge bars in water, but be sure to quench them slowly because they may otherwise crack and be valueless. Also, when you quench them too fast you obtain martensite in the upper parts making it hard to read the chill. In quenching, take hold of the wedge big side down and immerse only the big side in the water, hold it there until the wedge becomes black, then quench all over. It takes only a few seconds but it makes considerable difference in what you get in your readings.

makes considerable difference in what you get in your readings. C. F. WALTON: The main advantage of the chill test is the speed with which it supplies information. However, when observing the fractured test piece particularly in a dark corner by the cupola it is sometimes difficult to distinguish the extent of the white or mottled iron. The ordinary 100-watt yellow insect-repellent light bulbs produce a light which makes the "reading" of a chill test piece much easier. Similarly, the photography of iron fractures is improved in contrast by the use of a yellow

(k) filter.

W. F. Bohm: * I would like to have the author comment on the effect of pouring rate on depth of chill. Anyone who has made automotive tappets which are cast against a chill plate, chilled at one end, can appreciate the fact that it is possible to alter the depth of chill 200 or 300 per cent, depending upon the actual metal entry into the mold cavity and how the iron actually strikes that chill plate. I am referring to the speed and the uniformity of pouring, the entry of iron into the mold cavity.

MR. KRAUSE: In the oral presentation of the paper I inadvertently skipped over many of the details of chill testing. In regards to the chilled plate type, I think anyone who is familiar with it has on occasions seen a double chill, i.e., a shallow chill with a thin layer of gray iron followed by another chill layer. That is often a result of a splash of iron hitting the chill and then followed by some more iron. It is quite important in pouring this type of chill test that it is poured rapidly and uniformly. The chill test cores often have a little pouring basin at one end to avoid pouring directly onto the chill plate and from this standpoint the vertical chill plate may be better than

the horizontal plate. But here again it is desirable not to pour against the chill. You should pour away from it.

In regard to the pouring temperature, I have sometimes found it is quite important and at other times I was unable to find a consistent relationship between the depth of chill and pouring temperature. It would nevertheless be desirable to pour chill tests at about the same temperature.

MR. BOHM: I think a casting of the type I was discussing is perhaps a little more sensitive than the average chill bar. The chill is probably in the order of, say $\frac{7}{32}$ in. and you get the anomolous gray streaks and low chills if you do not pour properly and have the metal actually entering as fast and symmetrically as you can. In other words, chill formation seems to be very sensitive in some conditions to the manner in which that metal comes in contact with that chill.

MR. KRAUSE: I think it is impossible to over-emphasize the importance of some of these details. I have seen foundries use chill test ladles with an inch of cold iron in the bottom. That gives you very erratic results. A clean chill test ladle is necessary to obtain reliable results from the chill test. And if you want to obtain consistent results, then you must also keep the cores and chill plates clean.

J. T. MACKENZIE: ³ The condition of the ladle must also be taken into consideration.

W. R. MOGGRIDGE: 4 The author showed an illustration of parallel-sided chill tests which had an inverted chill at the middle. Would you consider such a chill which had an inverted section, as a representative chill? Would you measure such a chill test from the surface to end of the clear chilled section and call it representative? We frequently find such conditions in 3.70-3.50 carbon and 0.70-0.90 silicon cast iron chill tests.

MR. KRAUSE: There is no consistent practice that has been followed in measuring depth of chill. Some have reported chill as so much clear chill or the distance from the chill face to the first gray speck, and then to the last white speck. If there is a very deeply mottled zone, it is about all you can do.

In the case of an inverse chill, I do not know how you could satisfactorily measure it. This type of chill is obtained quite often with low-silicon, high-carbon irons. I do not know how to express the measurement.

MR. MOGGRIDGE: If you design the chill so as to avoid parallel sides, does it decrease the tendency to inverted chills in such irons?

Mr. Krause: Sometimes you see a white area in the center of a wedge.

A. A. EVANS: 5 We did some work on the correlation between chill depth and average hardness. Did the author do any work along that line?

MR. KRAUSE: For a given foundry and cupola practice there is a fair relationship between the hardness of the casting and depth of chill but you can greatly upset that relation by varying the ladle treatment. You can divide chill test applications into three groups, one of which is merely a selection process. You might cast a chill test, break it quickly and if the chill is greater than a certain amount, pour heavier castings, and with less than a certain amount you can pour the lightest jobs.

The next step would be to make some alterations of chilling tendency by ladle additions, either adding ferrochromium or something of that nature to harden the iron, or ferrosilicon or some other innoculant to reduce the chill.

The third application is in the manufacture of processed iron which is rather a complex specialized one. We have conducted surveys in a large number of foundries, trying to correlate depth of chill with Brinell hardness and strength of the iron. Since there is a general relationship between the depth of chill and carbon equivalent, you will find some relation between tensile strength and depth of chill. But it is rather broad. You cannot look at any chill test and without knowing something of the melting practice, predict the tensile strength from the depth of chill.

Some foundrymen are misled somewhat if they feel they are getting uniform tensile strength by obtaining a uniform depth of chill through ladle additions of either ferrochromium or ferrosilicon. Strength and chilling tendencies are related but

¹ Research Metallurgist, Cooper-Bessemer Corp., Grove City, Pa.

² Professor, Case Institute of Technology, Cleveland.

^{*} Asst. Fdry. Met., Buick Motor Co., Flint, Mich.

^{*} Technical Director, American Cast Iron Pipe Co., Birmingham, Ala.

⁴ Chief Inspector, Plant 5, Ford Motor Co. of Canada, Windsor, Ont.
⁵ Foundry Quality Control Engr., International Harvester Co., Indianapolis,

ON

our

ne

ind

our

is

he

the

op

tri

be

hat

the

ase

ise

es

OH

the

be

of

he

ed

nd

in

en ill he a ild ite ow lel ch of

th he to ou is th

on ed th ce ad le ot he th

or

the relation can be broad. But the more control you exercise in melting practice the better the relation becomes between chilling tendency and properties of the iron. You can work out a good relationship in one foundry and then find in the next toundry that the relation is too broad to be useful because of inadequate control.

DR. MACKENZIE: In that connection, I think Dr. Austin's bismuth treatment is one of the best examples I have seen; that also goes for the tramp elements. Lead will reduce depth of chill on soft iron and increase it on hard iron. It is surprising how much lead goes into the cupolas these days.

CO-CHAIRMAN WILDER: Dr. MacKenzie would you care to comment on the effect of antimony on depth of chill?

DR. MACKENZIE: Spencer and Walding made quite an investigation of this and reported their finding in A.F.S. TRANSACTIONS (1932). They added antimony, lead, and combinations of them and I am quite sure you will find it changed the chill. It just depends upon what kind of iron you happen to be running as to which way it will change it. For instance, tin and antimony do not affect the strength very much, whereas lead will increase the chill and reduce the strength, just like bismuth which, as Dr. Austin showed, increases the chill enormously but the strength is very much lower and back field of the chill more open than the one without the bismuth.

C. D. SEARS: 6 As a matter of interest, we have had better luck reading and interpolating chill test blocks using a gage made by ourselves. It is simply a lever arrangement giving an extended reading. It is graduated in ½2 in. Ten divisions of

* Foundry Metallurgist, International Harvester Co., Indianapolis, Ind.

 $1_{32}'$ in. is expanded over about 3 in. Through extensive correlation we are able to predict casting hardnesses from chill depths. So critical is the chill, we have to divide the $1_{32}'$ in. division into four subdivisions.

We would be glad to send anyone interested a drawing of our

gage so they might make their own.

MR. KRAUSE (Author's Closure): The author wishes to thank Mr. Eagan for his remarks concerning proper procedure in quenching of hot test wedges. The suggestions made by Prof. Walton should prove very helpful and since they are very simple to apply they should be tried. The suggestion to round off the back of the wedge test piece should increase the utility of the wedge test.

The description of a simple device for measuring the chill on a wedge piece by Mr. Sears is much appreciated as anything that can be done to simplify measurement of the chill and reduce chances of error is a worthwhile contribution to the art.

Chill and wedge tests are so relatively simple and inexpensively made that many foundrymen have developed designs to meet their own needs. Standardized designs would meet their needs as well and in addition would allow interchange of information with other foundrymen. Many foundrymen have undoubtedly developed unique practices or devices for measuring depth of chill which would be of considerable interest to the A.F.S. Chill Test Committee and other gray iron foundrymen. These foundrymen may think their ideas too simple to be of any interest to others but it is only through sending in such ideas to the A.F.S. Chill Test Committee that rapid progress can be made. Your comments and suggestions are therefore very much appreciated.

FATIGUE DATA SUMMARY

Report No. 2

By

M. E. Annich*

ABSTRACT

This report is an interpretation of data on fatigue factor values used by the Foundry Industry in its production operations. The A.F.S. Timestudy and Methods Committee circulated a questionnaire in the Foundry Industry and compiled the required data herein presented.

This report is a presentation of several special methods of determining fatigue allowances to be used on various operations in the foundry. These special methods were reported to the American Foundrymen's Society Time Study and Methods Committee in answer to a questionaire circulated among Society members during the latter part of 1949.

Report No. 1, presented in 1950 at the Cleveland Convention was an attempt to summarize and present fatigue values used in plants of reporting members in which fatigue is applied on an overall rather than an elemental basis.

1. Overall Basis

The following special method of determining fatigue allowances on an overall basis by a point rating system similar to that used in job evaluation, was previously presented in the November 1948 issue of the American Foundryman. Its author is John R. Walley, Supervisor of Standards of the Ohio Steel Foundry Co., Springfield, Ohio.

This method is again presented here because, although it may be subject to some question as to the coverage of the factors used or the values assigned to these factors, it is a step in the right direction in approaching the problem of fatigue allowances more analytically and objectively, and, as the author states, "taking the guesswork out of Fatigue Allowances."

The complete article, just as it appeared in American Foundryman follows:

"FATIGUE ALLOWANCE granted to employees in compensation for lost time, due to physical tiredness, is a controversial subject among industrial engineers, and a source of ridicule and grievance among workers. This situation is further aggravated by the fact that many fatigue allowances have been established arbitrarily, with no basis for substantiation.

It is no longer necessary to guess at what a fatigue allowance should be for a certain occupation. The procedures used in point plan job evaluation can be carried over to the appraisal of the various fatigue factors within an occupation. All the benefits usually ascribed to improving rate alignment apply to fatigue allowances. The allowances can be considered fair and equitable for they have a reasonable and logical basis, and they can be explained in a manner that most workers appreciate.

When establishing a basis for fatigue allowances, it is a sound plan to appoint a committee of six at the start. One member should be responsible for gathering data, and the other members should assist in evaluating fatigue content. The opinions of a group are better than those of one man.

The first job of the committee is to select a series of not less than eight (8) key jobs within the foundry. The key jobs should include what is considered the most fatiguing and the least fatiguing occupations to which fatigue allowances will be applied; then between the extremes will be a number of occupations with varying fatigue content. The committee should appraise the list of key jobs and place them in rank order.

A representative key job list, ranked with the occupation having highest fatigue content first, might be as follows:

- 1. Squeezer molder operator, large work.
- 2. Chipping, large castings.
- 3. Flogging heads.
- 4. Swing grinding.5. Stand grinding.
- 6. Squeezer molder operator, medium work.
- 7. Burn-off man, small castings.
- 8. Bench coremaker.

The committee should then determine what factors cause fatigue within the key occupations. Foundry

^{*} Supt. of Standards, American Brake Shoe Co., Mahwah, N. J. This is a report of the Timestudy and Methods Committee of the American Foundrymen's Society. Personnel of this Committee are as follows: E. G. Tetzlaff, Chairman, M. E. Annich, Roy Diffenbaugh, J. J. Farkas, Wally E. George, H. Reitinger, M. T. Sell, Dean Van Order, Jeff Alan Westover and H. R. Williams.

CHART 1-FATIGUE APPRAISAL CHART

I. STRENGTH

A measure of the heaviest weight which must be lifted by an employee during his regular work. Where a hoist is used, the next heaviest object must be considered.

- 1. Over 150 lb.
- 2. 100 to 149 lb.
- 3. 75 to 99 lb.
- 4. 30 to 74 lb.
- 5. To 29 lb.

in

S

S.

at

i.

ie

7-

r

rs

d

SE

it

e

n

p

ef

0

2

S

d

II. MONOTONY

Refers to the repetitiveness of work without changes in job.

- 1. Only one job a day, no changes.
- 2. More than one job, but not over four per day.
- 3. Usually four to eight different jobs.
- 4. Generally eight to 15 jobs a day.
- 5. More than 15 jobs per day.

III. MENTAL STRAIN

A measure of the employee's alertness due to danger, and the attendant nervous and mental strain. The required use of safety equipment is a guide to danger within an occupation.

- Must use goggles and leggings during entire work cycle.
- 2. Must use goggles during entire work cycle.
- 3. Must use goggles, gloves, leggings, or other equipment during part of work cycle.
- 4. Must use protective equipment occasionally.
- 5. Never needs protective equipment on regular job.

IV. UNITS HANDLED

A measure of the units handled over a full working day, based on the average number of units per man.

- 1. 409 units and over.
- 2. 190 to 408 units.
- 3. 88 to 189 units.
- 4. 41 to 87 units.5. To 40 units.

V. Noise

A reference to the volume and frequency of sounds due to equipment which is used by the employee on the job.

- 1. Constant loud noise, with very little relief.
- 2. Loud, shattering noise most of the time, but with some relief.
- 3. Some type of noise on the job, but not loud or shattering.
 - 4. Loud, shattering noise intermittently.
 - 5. Very little noise.

VI. DIRT

The amount of dirt to which employee is subjected over a full day's work.

- 1. Covers clothes and body. Dirt is of a clinging nature, and cannot be brushed off.
 - 2. Covers face or hands mostly. Clings.
- 3. Nuisance dirt on clothes, face, or hands, but which can be brushed off easily.
 - 4. Generally on hands only.
 - 5. Very little dirt encountered.

work indicates a number of factors such as strength, monotony, mental strain, units handled, noise, and dirt. Other factors and other names for the same factors may be used; however, the important thing is that a clear-cut distinction exists between each factor. A degree of each factor should exist in each occupation.

Fatigue factors must be clearly defined. Each member of the rating committee should understand the definitions. Further, five degrees of each factor should be established so that it will be possible to tell which degree exists in an occupation. A sample set of defined factors and degrees may be seen in Chart 1.

The rating committee member who gathers information concerning occupations should have a simple form for making notes. In collecting his information, he can then look for the specific information needed. After appraisal of fatigue content, another one of the same forms can be used for the permanent file copy. All reasons for designating a certain degree of a factor, to a specific occupation, should be recorded. This information is often necessary for substantiating a rating, and for rating new occupations for which a fatigue allowance is needed. A form that serves this purpose is shown in Chart 2.

It is now possible for the committee to appraise fatigue content on all jobs which are to have a fatigue allowance. The tools are a set of defined factors, with varying degrees of the factor, and a write-up of each occupation showing what causes fatigue. It is a simple matter for the committee to complete this job phase.

Next, after appraising the difference in fatigue content of the occupations, a cross comparison of degrees should be made. The cross comparison is done by list-

CHART 2-FATIGUE FACTOR WEIGHT APPRAISAL

Job Title	DepartmentDate
STRENGTH-	Degree Points
MONOTONY-	DegreePoints
MENTAL STRAIN-	DegreePoints
UNITS HANDLED-	DegreePoints
Noise-	Degree Points
DIRT-	Degree Points
Job Grade	Fatigue Range% to%
	Total Points

CHART 3-FATIGUE APPRAISAL WEIGHT

	Degree									
Factor 1	2	. 3	4	5						
Strength300	240	180	120	60						
Monotony250	200	150	100	50						
Mental Strain150	120	90	60	30						
Units Handled150	120	90	60	30						
Noise100	80	60	40	20						
Dirt 50	40	30	20	10						

CHART 4-POINT VALUE-FATIGUE ALLOWANCE CORRELATION

Point Range	Per Cent
900-1000	33
810-899	30
725-809	27
654-724	24
587-653	21
529-586	18
476-528	15
428-475	12
386-427	9
To 385	6

ing occupational titles vertically, and fatigue factors horizontally. Then the degree of each factor contained in an occupation is shown opposite the title and below the factor. Each appraisal of a degree should be questioned as to how it compares with other degrees on other jobs. The cross comparison is an overall check on judgment, and affords an opportunity for revisions when necessary.

Another job now confronts the committee. Point weights must be given to each factor and its degrees. The committee should determine which factor causes the most fatigue, and which the least. The factor which causes the most fatigue should be allowed the most points, for example, 900. The 900 points should then be broken down into five classes corresponding to the five degrees of the factor. The same procedure can be used in determining point weight for each factor used. It is important that different weight values be given the various factors in order to clarify the actual difference between fatigue factors, and degrees. A sample form is shown in Chart 3.

The point weights assigned to each degree should be marked on the form which shows substantiating information. The points per occupation are to be

Various industrial engineers who are in accord with the practice of making an allowance for fatigue state that a minimum of 6 per cent should be established. The maximum allowance has not been so clearly specified. If 33 per cent were determined as the maximum, a range from 6 to 33 per cent could be used. The range should be established by the committee prior to rating.

Total point values per occupation can be correlated with a fatigue allowance by means of a simple chart. Then it is necessary only to know the total points in order to determine the proper allowance (Chart 4).

While this method of determining fatigue allowances is quite lengthy when put into writing, nevertheless it is simple to explain, and is easily understood by workers. Guesswork is not needed in an application of fatigue allowances developed in this manner."

2. Compromise Method

Many time study people are averse to using fatigue allowances on an overall basis because it is impossible to get the accuracy of values between jobs that can be obtained by using an elemental method. Moreover, the elemental method lends itself more readily and more logically to any changes in the fatigue allowance that may be made necessary through a change of practice in a job.

The following brief example of what one foundry is doing as a compromise between the overall method and the elemental method is presented here. This plan recognizes the need for a greater amount of fatigue on certain elements, and in varying degrees, based on the physical exertion required.

The same sort of a plan might be followed on other operations although this particular one applies only to molding operations.

The writer states his plan briefly, as follows:

"To the normal basic elemental times is added a

minimum of 15 per cent for personal needs and relaxation (5 per cent is personal). There may also be added, up to a total of 25 per cent, an extra fatigue allowance on elements of the operation requiring extra physical exertion. These elements are, riddle sand, air ram, strike off, and flask carryout."

Under Section 5 of this report in which is summarized elemental fatigue values used by several of the reporting members, this same approach to the problem will be noted where, on certain elements, there is a range of values used which vary directly as the flask size, weight, etc.

3. Elemental Basis

The following method of applying fatigue on a purely elemental basis was discussed in "Fatigue Data Summary, Report No. 1," A.F.S. TRANSACTIONS, vol. 58, p 169 (1950) and is included in this report to complete the data.

This method applies fatigue based on the class of motion and the weight handled or force exerted. It is a simplified presentation of two almost identical methods which were submitted in answer to the questionaire.

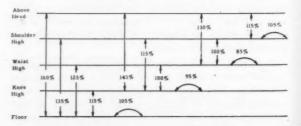
In this method, the per cent fatigue allowed for two basic movements, Knee to Waist, or Waist to Shoulder, are given in Table 1, varying according to the weight lifted or force exerted.

TABLE 1—PER CENT FATIGUE ALLOWED FOR KNEE TO WAIST OR WAIST TO SHOULDER MOVEMENT

Weight Lifted or Force Exerted in Lb.	Per Cent Fatigue Allowed	Weight Lifted or Force Exerted in Lb.	Per Cent Fatigue Allowed
0	4.0	60	34.5
5	5.5	70	40.5
10	7.5	80	46.5
15	9.5	90	52.0
20	12.0	100	57.5
30	17.0	110	61.5
40	22.5	120	65.5
50	28.5	130	69.0
		140	71.0

If, however, the elemental movement is different from the basic movement, the amount, in per cent, by which the basic values must be changed is given in Table 2.

TABLE 2-MODIFICATION FACTOR BASED ON THE CLASS OF MOTION INVOLVED



4. Another Elemental Basis Method

Another method of determining the fatigue allowance to be applied on an elemental basis was pre-

Jo Lo Li

H

M Oi Pa

Per Pic Pos

Pos Pos Pos Pos

Posi W Posi epe ne ng le

bre

ta ol. to of It al

or to

O

nt nt, en

55

W-

TABLE 4-ELEMENTAL FATIGUE ALLOWANCES IN MOLDING reent Elemental Fatigue Allowances Used in 4 Plants. Each value given indicates use by one plant

Marting Mart			Small						Small			
Squeeze Sque			Machine						Machine			
Element			Molding:						· ·			
Element Bench Rollover, Special Rollows, Rollover, Rollover, Rollover, Rollover, Special Rollows, Rollover, Roll			JSS;	Medium			Sq	ueezei				
Fig.												
Name	Element	Bench		Rollover	Machine	Shirger	Element B	Bench		Rollover	Machine	Slinge
Note Description Descrip	diust Pins		5				Prepare Floor for Patt	tern B	Board			10
Normal N		Drag		5		10						
Namehic Flask Flattern 10 10 10 10 10 10 10 1				-								25
Repetition 10 10 10 10 10 10 10 1							0 /	vevor	10			
Bloard Squeeze 0		10	10									
On: Off 5 5 5												
Carryout (not Mechanical) 15 to 35' 15 co 35'		5	5				Raise Table for Draw					
Content 15 to 35 20												
Clamp Pate 15		15 to 35	1						15		15	
Clamp 10	not interminent,						3 .	10				
Clamp Plask to Board	Clamp		10						5			
Clamp Piask to Board			10		15	15	accum opinio					
Clamp Drag		ard					Remove Clamps			10		10
Classystant Cope Bars		M C	15			40		8 to 15	1 126	**		
Clean & Position Plate		rs	40		5							
Close Mold						10						
10 to 25 15 to 25			9			10						
15 20** 20	CHOSC MORG			3			Remove Sprues & Riser	I'S		10	15	15
Cau Sprue 5 9							The special section					2.0
Return to machine 5 5 5 5 5 5 5 5 5	Cut Sprue											
Riddle			9				Return to machine			5		
Machine Manual 15 to 25* 15 to 25* No.		ten 3						10				
Rollover 15 to 35 15 to 35 5		110	15 to 954	15 to 951			Riddic					
Machine or Pattern		ai)	13 10 23	13 10 23			Pollover 15 to			5	15	10
Door Cope from Machine		1993					Ronorci			J	10	10
Cope from Machine to Table		TII			15	15	Sand in Shovel	20	10			
To Table 5 20** 20 to 35*					15	15		15	10	90	95	30
Sand in-From Hopper 5 5 7 7 5 5 7 5 7 5 7 5 7 5 7 5 7 5 7 5 7 5 7 5 7 7			nowh				пош неар			20	43	30
Set Chaplets 5 7	to Table			4			Sand in From Honner	1.0				
Downgates 10 to 25 15 to 25 15 15 15 15 15 5 16 17 17 17 17 17 17 17	7211	2						-				
Draw Cope							Set Chaplets		1			
Draw Match 15						15	6 61-111-					
Draw Match 15 15 15 15 15 15 16 10 10 15 10 10 10 15 10 10	Draw Cope						Set Chills		5			
Draw Pattern 10			5		15		6 . 6		-			
10 5 10 10 10 10 10 10						-	Set Cores (small)		7			
10)raw Pattern			5		15	6 . 6		***			
Draw Spike in—out 5 16 16 15 16 16 16 16		10					set Cores (meaium)		10		15	15
Set Core Crane Set Core Crane Set Core Crane Set Gate Core Set Gate Core Set Gate Core Set Strainer Core Set Set Strainer Core Set Set Strainer Core Set Set Strainer Core Set Set Set Strainer Core Set Set Set Strainer Core Set			10									
Center Crane		5					6.6.4	13				
Set Ramup Core 5 5 5 5 5 5 5 5 5												15
Set Strainer Core 5											15	
Set Gaggers 10 15 Evel Sand 5 5 5 Evel Mold Weights 7 Evel Pattern in Match 5 Evel Pattern in Match 10 Evel Pattern	olt	5		5	10							
Set Nails 5 5 5 5 7												
See Sand							00			15		
Set Mold Weights 10 Set Mold Weights 5 Set Mold Weights 5 Set Mold Weights 5 Set Mold Weights 5 Set Mold on Conveyor Car (Crane) 10 Set Pouring Basin 10 10 15 Set Sprues & Risers 5 10 15 Set Sprues & Risers 5 10 15 Set Sprues & Risers 5 Set Mold Weights 5 Set Mold Se			5				Set Nails					
Set Mold Weights 7 Set Mold Weights 7 Set Pattern in Match 5	evel Sand	5	5					5				
Set Pattern in Match 5 Set Pattern in Autonical Set Pattern in Auto	ift by Hoist		9									
Set Pallet on Conveyor 5 Set Pouring Basin 10 10 15 15 15 16 16 16 16 16	latch on Bench	15							7			
C(Crane) 10			20wh									
Set Sprues & Risers 5 10 15 10 15 10 15 10 15 10 15 10 15 10 15 10 15 10 15 10 15 10 15 15		Car						5				
Machine			10				0					15
Spray Squeeze S S S S S Squeeze S S S Squeeze S S S Squeeze S S Squeeze S S S Squeeze S S S S Squeeze S S S Squeeze S S S S S S S S S			10 to 20 ^s				Set Sprues & Risers			10	15	
Squeeze 5 5 5 5 5 5 5 5 5			5						10			
Squeeze 5 5 5 5 5 5 5 5 5	artine	5	5	5	10	10	Spray					10
Strike off 5 5 5 5 10 15 15 15 1		5	10				Squeeze	5				
Strike off 5 5 5 5 5 5 5 5 5	atch		5			10			5			
Strike off 5 5 10 15 20									5			
15 16 15 15 15 15 15 15	een	10					Strike off	5	5			10
15				1				10	15			20
15 15 15 16 16 16 17 17 18 18 19 19 10 10 10 10 10 10	een Sprues & Rise				15				15			
10 9 15 Vents, Push 5 5 5 16 16 17 16 17 16 17 17 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18 18						15	Vents (Rod or Pipe)		5		10	10
Vents Push 5 5												
manually alone sition Metal Upset (56 lb) manually alone sition Wooden Upset sition Mold (Crane) sition Plate (90 to 130 b) (Manually-alone) sition Plate (over 130 lb) (Manually- sition Plate (over 130 lb		et (56 lb)		15			Vents, Push	5	-			
vents, Scratch 5 7 (56 lb) manually alone osition Wooden Upset 10 15 Wash Mold With helper osition Mold (Crane) 10 Range of flask volume 1440 to 2600 cu in. 15 Range of cope volume 800 to 2600 cu in. 16 Solition Plate (over 130 17 Solition Plate (over 130 18 Solition Plate (over 130 19 Solition Plate (over 1440 to 3000 cu in.		(00 10)										
(36 lb) manually alone sition Wooden Upset 10 15 wh With helper sition Mold (Crane) 10 25 Range of flask volume 1440 to 2600 cu in. lb) (Manually-alone) 25 25 a Range of cope volume 800 to 2600 cu in. sition Plate (over 130 b) (Manually-alone) 25 25 a Range of flask volume 800 to 1800 cu in. b) (Manually-alone) 25 25 a Range of flask volume-1440 to 3000 cu in. c) Depending on weight—no range reported. Range of flask volume—1440 to 3000 cu in.		et		15			Vents, Scratch					
solution Wooden Upset 10 15 wh With helper Range of flask volume 1440 to 2600 cu in. solution Plate (90 to 130 15) 25 25 25 25 25 25 25 25 25 25 25 25 25	to the same of the			13								10
position Mold (Crane) 10 Range of flask volume 1440 to 2600 cu in. Range of cope volume 800 to 2600 cu in. Range of cope volume 800 to 2600 cu in. Range of cope volume 800 to 1800 cu in. Range of cope volume 800 to 1800 cu in. Depending on weight—no range reported. Range of flask volume—1440 to 3000 cu in.			10			15						10
sition Plate (90 to 130 Bange of cope volume 800 to 2600 cu in. Range of cope volume 800 to 1800 cu in. Range of cope volume 800 to 1800 cu in. Bello (Manually- Manually- Range of flask volume-1440 to 3000 cu in.			10					luma	1440 ** 0	600 cm in		
lb) (Manually-alone) 25 25 Range of cope volume 800 to 1800 cu in. b) (Manually- Bange of cope volume 800 to 1800 cu in. Depending on weight—no range reported. Range of flask volume—1440 to 3000 cu in.						10						
osition Plate (over 130 Depending on weight—no range reported. Range of flask volume—1440 to 3000 cu in.			or	0"								
b) (Manually- Range of flask volume-1440 to 3000 cu in.			25	25			remile or cobe					
transfe or many		r 130					Depending on we					
with help) 20 ° Up to 4600 cu in, flask volume.			-				remise or mean to				п	
osition Slinger Head to Mold 10 7 Over 4600 cu in. flask volume.							cp to 4000 cu in					

sh be tra

in

in

an

no

and

alle

fou

par

bec

star A care acce allo

mol nerv

sented in a paper entitled "Timestudy Techniques" in A.F.S. Transactions, vol. 55, p. 221 (1947). One plant reporting in answer to the questionaire stated that this method had been applied to several molding operations and had proven satisfactory, and was to be extended to other operations on molding and coremaking in their new ferrous foundry which had just gone into operation.

This method also involves the type of motion of the body and the weight lifted or force exerted. See Table 3.

TABLE 3

Part of body used and type of movement	0-10					Exerted 101-125	
Hands & Fullarm	1.01	1.03	1.09	1.19	1.34		
Hands, Fullarm & Shoulders	1.02	1.04	1.12	1.25	1.45		
Hands, Fullarm Shoulders & Back partly ber (Less than 45°)		1.05	1.14	1.28	1.49	1.75	2.05
Hands, Fullarm Shoulders & Back partly ber	1.04 nt	1.07	1.17	1.35	1.60	1.92	2.30
(More than 45° Carryout & Close	1.04	1.07	1.17	1.35	1.60	1.92	2.30
Carrying an Object Walking	1.02	1.04	1.13	1.27	1.47	1.70	1.95

5. Summary of Elemental Fatigue Allowances

Although the elemental method of applying fatigue allowances is no doubt the best and most accurate way of doing the job, it does involve a considerable amount of detailed work to analyse a job and set up the allowances initially in setting up standard data.

However, once the allowances have been set up for the various elements in any job they can be used again for similar elements in other jobs. To facilitate such a practice and to serve as a guide for anyone wishing to apply elemental fatigue allowances, the following compilation of fatigue allowances made on various elements by four reporting members is given in Tables 4, 5, 6 and 7.

TABLE 6-ELEMENTAL FATIGUE ALLOWANCES
IN GRINDING

Element St	and Grind	Swing Grind			
Blocking Pick up and Aside (hand	5 (0 to 40 lb) 10 (40 lb & over	7 5 (0 to 75 lb) 10 (75 lb & over)			
Pick up and Aside (hoist) Position (hand)	7	5 (75 lb & over 5 (0 to 75 lb) 10 (75 lb & over)			
Position (hoist) Grind	7	5 (75 lb & over)			

DISCUSSION

Chairman: M. T. Sell, Sterling Foundry Co., Wellington, Ohio.

Co-Chairman: H. R. WILLIAMS, Williams Management Engineering Co., Milwaukee.

HENRY FELTEN: 1 We did some work on the fatigue allow-

TABLE 5-ELEMENTAL FATIGUE ALLOWANCES
IN COREMAKING

Percent Elemental Fatigue Allowances Used in 4 Plants. Each value given indicates use by one plant.

Element	Bench	Floor
Bend Rods, Wires, etc.	9	10
Carry Tray Unfinished		
Cores	15	
Center Crane for Draw		5
Clamp Plate to Box		10
Clean & Oil (Dust) Box	5	5
Clean off Bedding Sand	5	
Clean Out Hook Holes Clean Out Nail Holes	5	10
or Rod Vents	5	
Draw	7	5 (crane)
	10 (24-lb bo	ox & core) 15 (manual)
	20 (55-lb bo	ox & core)
Dry Cores (Torch)	5	
File Cores	5	
Fill Joints after Pasting	10	
Get Empty Core Plate	5	
Level Sand in Core Box	5	15
Load Cores in Trav	10	
Paste Cores	5	
Patch	5	5
Plate to Rack	9	
Position Plate		5 (crane)
		15 (manual)
Rap Box		10
Ram-hand	10	10
1101111	10	15
Return Corebox to Bench		45
to beller	10 (261/2 lb)	
Remove Clamps	10 (20/9 10)	5
Roll Box & Plate		5 (crane)
Rub Cores to Paste	10	5 (cranc)
Sand in Core Box (hand)	5	
and in core box (initia)	10	
and in Core Box	10	
(Shovel)		10
(SHOVEI)		15
et Nails, Wire	5	5
et Hans, wife	7	9
et Rods	*	10
pray with Oil		5
trike off	5	10
trike oil	7	10
ent	5	
ent	7	
	9	
	10	
ie Wires	10	5
Vash Cores-Blacking or		3
Silica Wash	10	

TABLE 7-MISCELLANEOUS ELEMENTAL FATIGUE ALLOWANCES

Element	Allowance
Shovel Sand	20
Walk with weight up to 10 lb	5
" 20 to 39 lb	10
" 40 to 59 lb	15
" 60 to 75 lb	20
Push wheelbarrow with sand (350 lbs)	20
Push wheelbarrow with concrete	25

ances by the evaluation process on certain operations. On some of the highly repetitive operations, we found that our fatigue allowance is too high. It has been difficult for us to determine just what has gone wrong with our evaluation. We do not know if the evaluation on monotony or units handled is too high, or just what it is.

ial)

al)

ne

ne

ne

ot

Where did this evaluation come from? How do you arrive at these various points of evaluation?

Member: In regard to pace rating, if a man is operating a machine at approximately 160 pace rate, does that not change

the fatigue factor value?

MR. ANNICH: No, because assuming that you have set your standard, you have set a certain percentage of fatigue value. That percentage has been set on normal, or 100 per cent, so that when the operator works at 160 per cent of normal actually he is getting the extra percentage for fatigue, is he not?

MEMBER: That is correct, but in pace rating individuals on a moving line, why can the individual do his day's work in practically half a day. This makes it difficult for the foundry melting department to keep the lines supplied with metal.

MR. Annich: If the man can do his complete day's work in a half day, then there is something else wrong with your standards. MEMBER: Exactly what is the set up for determining actual pace rating? Some day the man works at a pace rate of 85 and another day at a pace rate on 120 and the standards are set on those pace rates.

MR. Annich: We have what are known as pace rating exercises. We went through a demonstration during the A.F.S. Convention in Detroit in 1947. For details see "Establishment and Use of Standard Data, Timestudy Techniques" by M. E. Annich,

A.F.A. TRANSACTIONS, vol. 55, pp. 221-240 (1947).

Also, the S.A.M. recently issued a set of films showing pace rating for 24 different operations. These films show certain jobs being performed at various paces or various levels of productivity and will give a timestudy man a good idea of what normal should be and how to rate accurately at various levels above and below normal. Then when a timestudy man is instructed and trained in rating these operations he should be able to rate accurately in almost any operation.

You can make films of several short-cycle operations in your plant to suit your specific purpose. You can take the film showing two or three cycles, whichever is necessary and make about a 10-ft loop using a film splicer. Keep running that loop through a variable speed projector. You can have your foreman seated in the projection room and have them pace rate the operation at various projection speeds. By doing this a number of times and plotting the results you can arrive at a good idea of what normal should be. If you prefer you can have your timestudy personnel do it. We did that with foremen in several of our plants and we have had very good success with it.

JOHN TAYLOR: 2 Did you say that these fatigue factors are being applied on an overall basis? They are not being applied

by elements?

Mr. Annich: Yes, that is correct. H. S. Thompson: About 10 years ago we used overall allowances. Today we are at least 5 or 6 per cent higher on personal allowance and supplementary fatigue. Do you attribute that to loosening up of standards or do men exhibit more fatigue today

than they did 10 years ago?

Mr. Annich: I got into this fatigue problem because we found out, after applying the overall fatigue allowances, that there were cases where some jobs were rated too rigidly compared with others. In some cases we could not apply standards at all. We could not get the standards accepted in the shop because we did not allow enough fatigue on some of the difficult jobs. Since then, wherever we made varying fatigue allowances, based on the amount of effort involved in a job, we found the standards more consistent and more acceptable to the shop man.

MR. FELTEN: I would like to clarify a point. We must be careful that we do not confuse fatigue allowances with rating accomplishment of the operator. Fatigue allowance is a time allowance given, depending upon the physical and nervous demands of the job, which make it possible for a workman to

perform it repeatedly all day long.

Let us suppose we are going to apply a fatigue allowance on two different types of jobs. An operator is producing a small mold at a high rate of performance; there is a low physical and nervous demand, so a small fatigue allowance is given.

On the other hand an operator works at a normal pace on an extremely large mold; the physical and nervous demands are high, so a high fatigue allowance is given.

Supt., Peoria Malleable Castings Co., Peoria, Ill.

One may produce 250 molds a day and the other only 12, but the earnings could be the same. In other words, we can maintain consistency in a man's earnings.

MR. ANNICH: I agree, but we are not using fatigue allowance to maintain a man's earnings, only when he is working at a

normal pace.

MR. FELTEN: I am thinking of an operation in our own foundry on which it was extremely difficult to apply fatigue allowances. It is the heat operation which involves shifting of weights, jackets, and molds for the iron pourer, and then dumping the molds into a windrow and hooking out the castings.

One shifter would have 200 molds containing about 5 lb of iron per mold. He was able to work at a high rate of perform-

ance at 10 per cent fatigue allowance.

Another shifter would have 120 molds containing about 40 lb of iron per mold. His rate of performance, naturally, was slower but there was a fatigue allowance of 55 per cent.

Consequently when we posted our heat performance reports there was a consistency of performance among the operators. It is not difficult at all to get a man to work on our heavy floors during heat operation, in fact some men prefer them.

MR. Annich: Thank you for supporting my view point. MEMBER: Fatigue values have a great deal to do with the physical makeup of the individual performing the job. Today. the smaller foundries are faced with the problem of upgrading men with, let us say, smaller men on these heavier jobs due to union regulations. Therefore, some of these fatigue factor standards cannot be applied.

MR. ANNICH: We all have that problem. I feel that if you must make arbitrary allowances over and above the normal allowances that you make in any timestudy program, you will

get into trouble.

MEMBER: If one man makes a mold in a certain length of time that is not necessarily a basis for setting the rate on that

MR. ANNICH: That is right. It takes in pace rating and all these other factors which we talked about in the past.

When we set a standard, we are trying to set a standard that can be attained by an average man. The fact is we are trying to set a standard where an average man can earn 25 per cent over his base rate.

MEMBER: There must be some definition as to an average man because of physical makeup. That is the problem. One man may produce 145 molds in a given time. The next man may be able to produce only 110 molds in the same time with the same effort. This is due to the physical makeup of the

MR. ANNICH: That is right. But regardless of that, we still have to set a standard if we are going to have incentive, or use our data for job costs.

MEMBER: Are there any figures correlating physical makeup

of a man to a certain type job in the foundry?

Mr. Annich: Not that I know of. The only thing I can refer you to are the various methods of pace rating.

MEMBER: A normal man is what we pick to determine what a normal rate, a normal incentive should be. A normal man is a trained man, qualified physically to do the job.

Pace rating is only one portion of the timestudy man's task. His effectiveness and his skill are requisites.

A timestudy man must recognize that an untrained man has not acquired what a normal operator on that job should be able to turn out.

MR. ANNICH: I think you are right and the difference in skill required on that job should be reflected in the base rate of the job. The man who works at normal rate is supposed to get his base rate.

HARRISON ATKINSON: 4 I am taking into account the logic of a normal man working and I am wondering if during 1934 and 1940 a normal man could turn out more than a normal man can today from the standards.

MEMBER: We are examining some 23,000 old rates which have been in the plant. The variance between those old rates and new ones, as being established through timestudy today, is extreme in many cases. Analysis of the job indicates that supervision allowed methods to change and, therefore, the skill and effort of the men have changed. That is reflected in your standards. I would not say that is rating nor would I say that it is fatiguing on the job.

² Vice-President, Lester B. Knight & Associates, Inc., Chicago.

Vice-President, American Boiler & Foundry Co., Milan, Mich.

⁴ Methods Engr., International Harvester Co., Hamilton, Ont., Canada.

MR. ANNICH: I think there is a compromise between the two view points. There is always a question about what is normal in different plants and in different parts of the country. What is considered normal in one plant will not be considered normal in another. One of the objectives we are striving for is to get a common denominator of what is considered normal, which will be used universally.

I will admit that possibly back in 1933 their concepts of normal were entirely different from what it is today. After all,

this pace rating idea is still in its infancy.

What we are trying to do is establish methods to determine what the normals are and then to have all parts of the country arrive some day at these normals and use them and keep them so that a normal pace now will be normal in 2000 A.D.

F. G. Olson: We had a four-month period during which

time we worked with M.T.M. We have a set of molding standards which I established some years ago. We checked these standards against M.T.M. and came out within 21/2 per cent

In 1946 we established blower standards in the core room

using M.T.M. and we came out exactly.

In another foundry of ours I established shifting and dumping standards in 1947. We spent 21/2 days going over the various motions and working up the standards. Through M.T.M. standards we would have spent \$391.70 and through the present standards we spent \$391.06.

I am a firm believer in M.T.M. and I have never seen a molding job on which you cannot use M.T.M. I am developing a set of standards on jolt-roll-over work using M.T.M. standards. You could even go so far as to develop iron pouring standards

with M.T.M.

I am referring to snap flask, jolt-roll-over and floor molding

methods. This does not include pit molding.

We also made fatigue allowances on each element based on weight and position of body. It has been my observation that when you go from an overall fatigue rating to an elemental fatigue rating, you have to train your people all over again as far as effort rating is concerned. The difference is considerable.

When we first started in timestudy we were trained what was then considered 100 per cent. That also included some allowance for fatigue. Now when you are going to divorce this fatigue from your effort rating, you will have to have a different concept of 100 per cent. We used to have flat-overall delays. Now that we are getting into elemental fatigues we have to forget about this 121/2 per cent and we must divorce our fatigue allowance in order that we can come up with approximately the same answer when we get through.

MR. ANNICH: I think you are right. We are starting to get refinements in our timestudy methods. We were possibly somewhat careless about timestudy in the past and did not worry too much whether we were lax 15 or 20 per cent. If we are to make fatigue allowances and make them as they should be, then we must refine our timestudy procedures. We are trying

to do just that with these pace rating methods too.

MR. TAYLOR: When applying fatigue by elements, it has been my experience that there are some companies who have followed the practice of using a leveling factor that also includes fatigue while other companies have used a factor that is a pure leveling factor and nothing else. Those companies that have always followed the practice of using a factor that is purely a leveling factor and nothing else, regardless of which way they go in fatigue, can make changes that will not disrupt their procedures very much. To those companies who have used a leveling factor that includes fatigue and supplemental as well as personal, it does mean a retraining of the timestudy per-

MR. ATKINSON: At present we are more or less going through that process where we leveled everything on the 100 per cent factor in trying to develop something more worthwhile and to be able to put a critical examination on timestudies. But we find ourselves with the biggest problem and that is to retrain our timestudy personnel and at the same time to retrain the employes being timestudied.

MR. Annich: In other words you are restudying the job with the idea that there has been a change of method.

H. D. HAMMOND: 6 The justification for timestudy is the fact that you want to give the men a fair day's pay for a fair day's work no matter whether they are working in New York or Kalamazoo. It is up to us to establish the right time standard on

the job and sell it to those whom it concerns as to what the allowance is going to be over and above a day rate rather than say, one man is 25 per cent over the standard, the next man 35 per cent over the standard, the next man 50 per cent over

the standard, for doing a fair days work.

MR. TAYLOR: I would like to explain an experience I had in connection with giving credit to the A.F.S. Timestudy and Methods Committee. On a recent job the question arose as to what we were doing about fatigue allowance. I replied "We will use the fatigue factors that have been developed by the A.F.S. Timestudy and Methods Committee. These figures are the concensus of approximately 54 foundries. Can you produce anything better?" Since no one could, I said "Unless you can show me averages from more than 54 companies, I think you will have to agree that this must be a pretty good figure." So, they decided to accept it as such.

MR. ANNICH: Thank you, Mr. Taylor, for the compliment to the work of the A.F.S. Timestudy and Methods Committee.

We are trying to set up something for management to work with in our studies on pace rating, fatigue allowances and anything else connected with timestudy and incentives. If, in their problems with labor, management changes values that we give them, they should still recognize that in these values we are trying to be as accurate as possible. If they have to change them they certainly should not expect us to change our method, our allowances, facts and figures just to suit the union or to sell the job. We should make each change as an extra allowance or a compromise to them, but not compromise our own figures and our methods.

E. A. Rehr: 7 Timestudy is a selling job and when we enter into a mutual agreement between labor and management, do we not destroy the fundamental principles of good basic timestudy?

Mr. Annich: I agree with you. I will give you an example. My assistant and I were working at a plant which was a machine shop and not a foundry. Management complained about the standards being too loose. They have a method of bookkeeping with their timestudies that every time they set a new standard, they put a new timestudy number on it. Each time there was a change in specifications or a change of the elements in a job, they would take an original timestudy that they had and to that sheet they would add whatever elements were put in or take out whatever other elements were taken out and come up with a new standard. Management complained that our standards were lax 25 to 40 per cent.

My assistant studied their timestudy files and in some cases spent as much as 2 hr on one standard alone to find out where the original timestudy came from and every time he got back to the original timestudy. At the bottom of the sheet was a note to the effect that the standard was "arbitrarily raised 25, 30 or 35 per cent" and initialed by the superintendent. And then they wondered why the standards were lax. That is the sort of thing we must get away from if we are to have good standards.

M. A. CUNNINGHAM: 8 Reference was made to using 5 per cent as a personal factor. Are there any figures to back

that up?

MR. ANNICH: In one of the reports that came to the Committee in its survey it was definitely stated that they had taken what we call activity studies over 70 days and they found that the personal allowance amouned to 4.8 per cent. It was felt that this backed up the 5 per cent which has been generally accepted and has never to my knowledge been questioned. If we did not have any more trouble with other allowances that we do with the 5 per cent personal allowance, then the timestudy man's life would be a lot less miserable.

MR. FELTEN: Anyone who has any experience taking all-day timestudies and carefully itemized the time that the operator takes out for personal allowance knows that it seldom exceeds

25 min per day. This is approximately 5 per cent.

CHAIRMAN SELL: There is one point that has not been stressed sufficiently in this discussion and that is that every timestudy engineer is a salesman and that perhaps we have one of the most difficult selling jobs there is. The timestudy man will not have a successful Standards Department or successful standards in his plant if he cannot sell them to the men.

⁸ Industrial Engr., Union Malleable Iron Works, E. Moline, Ill. ⁶ Chief, Costs & Methods Dept., Electro-Alloys Div., American Brake Shoe Co., Elyria, Ohio.

⁷ Industrial Engr., Oil Well Supply Co., Oil City, Pa. 8 Industrial Engr. Dept., Clark Bros. Co., Inc., Olean, N. Y.

SCAB DEFECT ON GRAY IRON CASTINGS

A Progress Report By
A.F.S. Committee on Physical Properties of Iron
Molding Materials at Elevated Temperatures*

Introduction

THE SCAB DEFECT is defined and characterized by an encrustation of sand and metal on the surface of a casting. Usually when the scab is removed, a depression will be found in the casting surface. The scab is caused by a piece of the mold face flaking or spalling off. This Committee undertook to learn something about the characteristics of scabbing and non-scabbing molding sands.

In a progress report published in the 1949 A.F.S. Transactions¹ complete details of the testing procedure are recorded. The Committee has formulated 18 sand mixtures most of which produced scabs on the test casting. Some 29 other sands were obtained from cooperating foundries. Most of these foundry-submitted sands did not produce scabs.

This report tabulates differences observed between

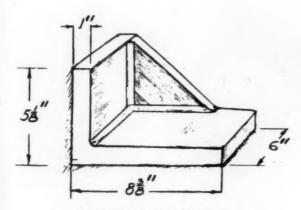


Fig. 1-Sketch of test pattern.

* Membership of the A.F.S. Committee on Physical Properties of Iron Molding Materials at Elevated Temperatures consists of the following: V. M. Rowell, Chairman, R. W. Bennett, J. E. Coon, H. W. Dietert, R. L. Doelman, H. H. Fairfield, J. A. Gitzen, John Grennan, H. J. Jameson, Roy Korpi, Henri Louette, R. W. Mason, C. S. Parsons, E. J. Passman, R. P. Schauss, W. A. Spindler, Michael Warchol, G. F. Watson, R. E. Wilke, J. T. Zayner, and E. C. Zirzow.

the scabbing and non-scabbing sands tested. It is hoped that the data contained herein proves useful as a starting point for foundrymen faced with the problem of eliminating scabs.

Procedure

The pattern used in this work is shown in Fig. 1. Molding procedure¹ was consistent for all castings, except 2g, 2h and 2i (Table 1) when mold hardness was varied. Three castings were made in each sand.

All castings were poured with gray iron at a temperature of 2650 F. Metal temperature was measured with a platinum-platinum rhodium thermocouple enclosed in a quartz tube. This device was described by W. A. Spindler in American Foundryman, April, 1947.

The castings were sandblasted and examined for defects. From each set of three castings a representative one was selected. The molding sands were tested in the laboratories of the H. W. Dietert Co.

Materials

Table 1 lists the data concerning the sands used. The preliminary work of the Committee showed that scabs could be produced consistently with a sand mixture containing about 20 per cent silica flour added to a coarse silica sand. This type of sand is listed under mixes No. 2 to 2i, 11 to 14, 29, 30 and 33. The other sands except No. 1 were obtained from cooperating foundries.

Approximate analysis of iron used was:

Total Carbon,	1/0	,			×			3.00
Silicon, %								2.30
Manganese, %								0.50
Sulphur, % .								0.10
Phosphorus, %								0.25

Test Results

Table 1 shows the results of a cross-section of the test data. Complete tests were made on 47 sands and the complete data are too unwieldy to publish in this paper. The first section of the table indicates the sand mixture or its source. Next are listed the green

the han han ver

to vill.

the uce can

So,

to ork ny-

ive are

em

nir

ell

OI

res

ter

we

lv?

le.

ne

he

ng

rd,

a

b.

at

ut

a

ere

ses

to

10

ey

ck

n

ad

at. en

wen

ds

ly

ne

ot

15

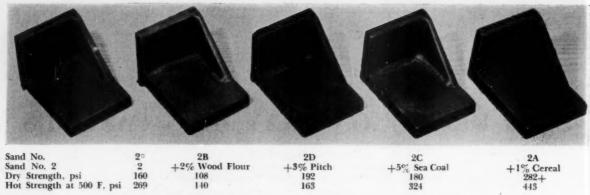


Fig. 2-Photo of castings made with Sands 2°, 2A, 2B, 2C and 2D.

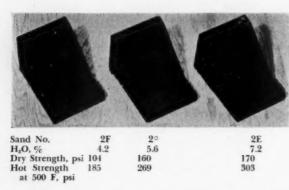


Fig. 3-Photo of castings made with Sands 2°, 2E and

Discussion

Members of the Committee advance the following theory to account for the formation of scabs.

Cause of Scabs

The formation of a scab on the surface of a castting is the final result of numerous reactions which take place within a few seconds of time. Granules of silica sand are compressed into a shape having some strength due to the adhesive properties of the materials used to coat the grains. The sudden application of liquid iron sets off a series of reactions. Moisture is converted to steam. Organic materials are decomposed. Clays and ceramic materials change structure and mechanical properties. The sand grains ex-

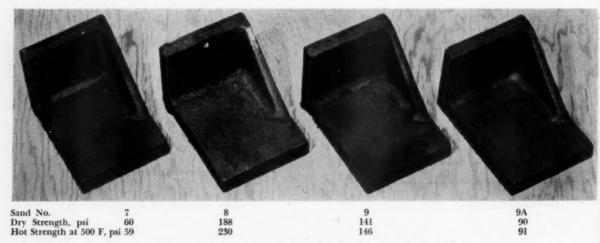


Fig. 4-Photo of castings made with Sands 7, 8, 9 and 9A.

properties of the sands. Dry strength and screen analyses are in the third section. Properties of the sands at elevated temperatures follow. At the bottom of the table the appearance of the test castings is described.

Photographs of representative castings are shown in Fig. 2 to 5.

pand and tend to cause a compressive stress in the surface layer of the mold. Each successive layer underlying the mold surface is at a different temperature and has a different set of physical properties. When the stresses set up due to expansion exceed the mechanical strength of the molding material, a piece of the mold flakes off.

h

of

16

n

eof

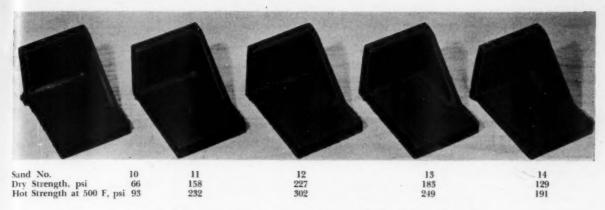


Fig. 5-Photo of castings made with Sands 10, 11, 12, 13 and 14.

Heat Shock Resistance

The problem of making a material to resist heat shock is not limited to molding sands. It would be useful to consider therefore some of the common practical knowledge concerning heat shock resistance.

Ordinary glass is liable to crack when exposed to heat. Pyrex glass does not crack on being heated, because it has a very low rate of thermal expansion. The superior spall-resisting characteristics of many refractories are attributed to their low thermal expansion rate.

Spall resistance of graphite and silicon carbide is partly due to high thermal conductivity. By heating up uniformly and quickly, stresses are minimized.

Ductility aids spall resistance in the case of metals.

Fig. 6-A scabbed casting, showing how the area of the scab was measured.

Gray iron must be heated carefully to avoid cracking. Soft, low-carbon steel absorbs the dimension changes due to heat without cracking.

Rammed plastic mixtures contain granular refractory which expands on heating, together with fireclay which shrinks on heating. By proper blending a spall resisting material is obtained.

Manufacturers of refractory bricks know that the tendency to spall is increased as the density of silica or fireclay brick is increased.

Silica (SiO₂) is the main component of molding sands, due to its availability and its high fusion point. The expansion characteristics of silica have been carefully studied.^{2, 8} Silica will expand more than 1 per cent in length when heated to 1500 F. The expansion characteristics of silica can be changed only by fusing the silica. This is obviously impractical for foundry sands. The only practical ways that expansion characteristics of molding sand can be changed is to alter the composition of the material surrounding the sand grains or to change the grain size distribution of the sand.

The Cure for Scabs

In view of the foregoing general information, in order to improve shock resistance of molding sand, a

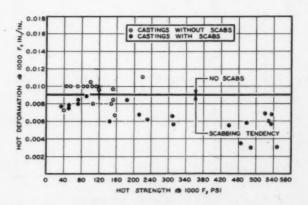


Fig. 7 - Curve showing relationship between not strength and hot deformation at 1000 F.

change in the ingredients of the sand mixture must be made. The significant property to control is not expansion, but rigidity, that is the capacity to absorb expansion without distorting or cracking. The Committee therefore advances the following hypothesis:

To cure the scab defect, reduce the rigidity of mold surface at high temperatures.

An examination of the test results will now be made in order to see if empirical test results support this theory. The scab is formed by physical forces in action. A.F.S. sand tests measure physical properties at rest. Therefore we cannot expect to get clear cut separation of good and bad sands.

The scabbed areas on the castings were measured by pressing aluminum foil over the defective areas and trimming around the edges. With the weight of the aluminum foil per square inch known, the area of the scab could be determined by weighing the ir-

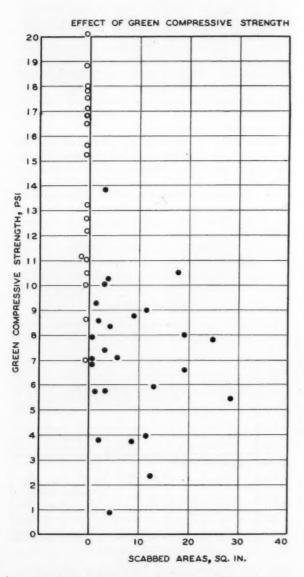
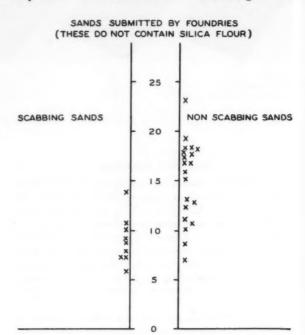


Fig. 8-Showing green strengths of all sands.

regular piece of foil which duplicated the scabbed areas (Fig. 6).

Hot Deformation at 1000 F

This temperature was selected because it is near the temperature at which the greatest stresses due to expansion are encountered. Table 1 and Fig. 7 show



GREEN COMPRESSIVE STRENGTH, PSI Fig. 9—Showing green strengths of sands submitted by foundries.

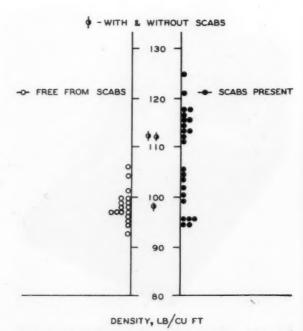
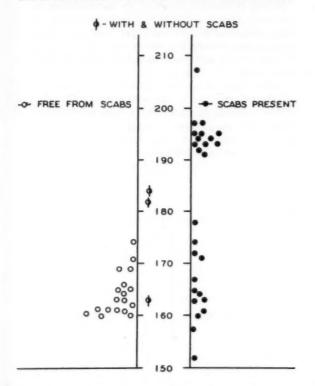


Fig. 10-Showing rammed density in lb/cu ft.

that the sands which deformed 0.009 in. or more did not produce scabs. The more rigid sands which deformed 0.007 in. or less, all produced scabs. The property of hot deformation is considered by the Committee to be the key to the scab forming tendency.

Green Compressive Strength

As Fig. 8 shows, the majority of the scabbing sands had a green compressive strength below 10 psi. At first it would appear that this is merely a coincidence due to the fact that all the silica flour sands were low



WEIGHT OF 2 IN. x 2 IN. SPECIMEN IN GRAMS

Fig. 11-Showing weight of 2-in. by 2-in. A.F.S. standard specimens

in green strength. Closer examination of the data shows (Fig. 9) that the foundry submitted sands which caused scabs had low green strengths also.

Since clay shrinks when heated, large amounts of clay would be expected to reduce the expansion stresses in molding sand.⁴ It is also suggested that if the subsurface layer of sand is tough and plastic, there is less tendency for pieces of sand to spall off when the sand is exposed to heat shock.

Rammed Density

The non-scabbing sands had a density below 110 lb per cu ft when rammed (Fig. 10). Half of the scabbing sands were heavier than 110 lb per cu ft. Also, as Fig. 11 shows, the non-scabbing sands weighed less than 175 grams when rammed into the standard A.F.S. 2-in. x 2-in. specimen.

It should be noted that most of the high density sands contained silica flour. The silica flour is a scab-forming ingredient which increases sand density. With silica flour, the sand is hard and brittle when hot. The addition of wood flour to the dense silica flour sands reduced the density, also the brittleness (increases hot deformation) and was helpful in preventing scab formation (compare sand No. 29 with sand No. 33 in Table 1).

There are other materials which might increase the density of molding sand without causing scabs. Graphite, silicon carbide flour and clay increase the density of sand, but they do not necessarily produce scabs. Whether or not high density is a factor in causing scabs therefore depends upon the type of fine materials in the sand. This investigation certainly shows that fines in the form of silica flour will cause scabs. Ramming a scabbing tendency sand to a higher density as indicated by higher mold hardness will increase its tendency to scab (Fig. 12).

Green Deformation

As Fig. 13 shows, most of the non-scabbing sands had green deformation values below 0.021 in./in.

The higher the deformation, the wetter the sand is, and therefore the tighter it will pack when rammed. High green deformation means higher dry and hot strength, and a more brittle sand at high temperature.

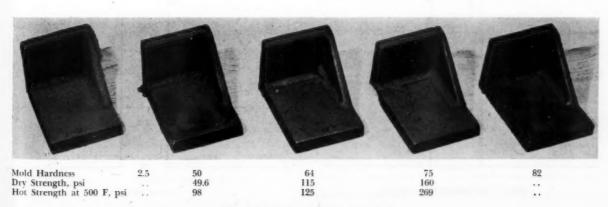


Fig. 12-Showing mold hardness

Consider the following results from Table 1. Sands 2E and 2F are identical in composition except for moisture content.

Property	Sand No. 2E	Sand No. 2F
Moisture, %	7.2	4.2
Density, lb/cu ft	125	113.6
Green Deformation, in./in.	0.025	0.012
Dry Strength, psi	170	104
Scab Area, sq in.	27.4	3.3

Wet sands pack tighter, have higher dry and hot strengths and therefore are more brittle. Green deformation should be considered as related to scab formation in that it indicates when the sand is too wet, and wet sands tend to cause scabs.

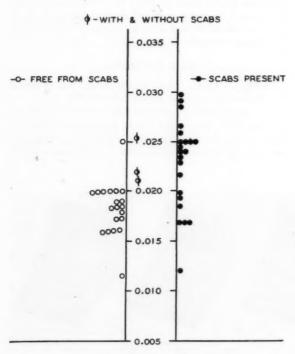
Flowability

There is a slight indication that high flowability is a characteristic of scabbing sands. It would seem logical that sands which pack tightly will be brittle at high temperatures.

Permeability

Sand No. 1 had a permeability of 278 and did not cause scabs. Adding 20 per cent silica flour produced sand No. 2. Sand No. 2 had a permeability of 65 and produced scabs on the castings. In this case the silica flour was the causative agent, and the permeability was one of its attributes.

When the foundry-submitted sands are considered separately (Fig. 14), it is apparent that permeability over 80 is found only in the non-scabbing sands. Low permeability indicates the presence of fines in the



GREEN DEFORMATION - IN./IN.

Fig. 13-Showing green deformation.

sand. Whether a low permeability sand will scab or not depends upon the type of material in the fines.

Dry Strength

The silica flour sands had higher dry strengths and produced scabs. Figure 15 shows that dry strengths over 150 psi, were associated with the scabbing sands.

Other materials might cause higher dry strength without causing scabs.

Silt Content

Figure 16 shows that two of the non-scabbing sands had more than 8 per cent silt. Fourteen of the scabbing sands had more than 8 per cent silt. This investigation certainly has shown that silt in the form of silica flour will make molding sand brittle and cause scabs.

Other forms of silt may be beneficial. Organic materials will show up in the screen analysis as silt. As Table 1 shows, the addition of organic materials reduced the amount of scabbing. Active clay which shrinks on heating is also a desirable constituent of molding sand. The high green strength sands did not scab, therefore, clay particles are not detrimental.

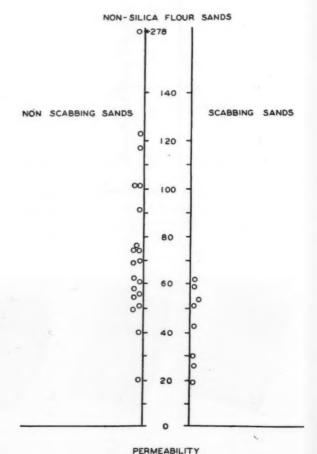


Fig. 14-Permeability of foundry-submitted sands.

11

d

Expansion of Molding Sand

The expansion tests performed on the sands described in this report did not show a definite correlation with scab defects. It is believed that the technique of testing was in error. No standardized test for expansion has been approved by the A.F.S. Further work on getting an approved standardized method is recommended. The tests were performed by submitting sands to 1500 F. It is probable that slow heating of the specimen would have given better results.

Hot Strength

The silica flour sands had higher strengths at elevated temperatures (Table 1) than did the foundry-submitted sands. When powdered silica is present in the sand, the sand is hard and brittle at high tempera-

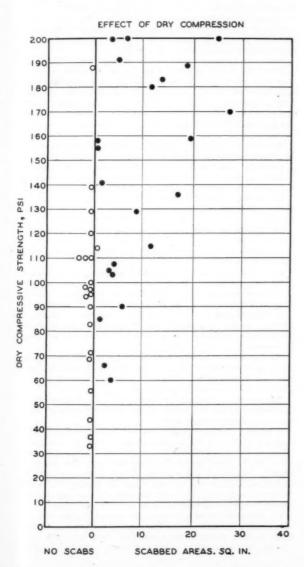


Fig. 15-Showing relationship between dry compressive strength and extent of scabbed areas.

tures and high stresses are set up due to the expansion of the silica flour.

With other types of fines in the sand, high hot strengths may not be harmful (Fig. 7).

ORGANIC MATERIALS ADDED TO MOLDING SAND

Sand Number	Materials Added	Scab Area sq in.
2	none	18.7
20	none	19.6
2A	1% Cereal	9.8
2B	2% Wood Flour	4.0
2C	5% Sea-coal	11.2
2D	3% Pitch	4.6
33	3% Wood Flour	0.0

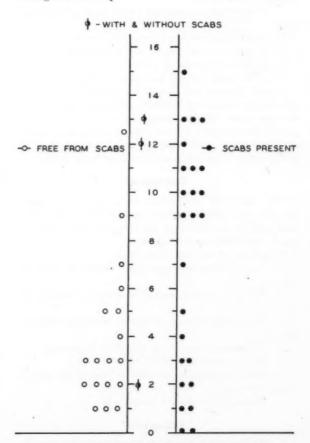
The beneficial effects of organic materials, especially wood flour, are quite apparent in the above table. Wood flour does reduce the brittleness of sand at high temperatures.

How to Correct a Scabbing Sand

To summarize the practical findings of this report, the following list of remedial measures is given as a guide for those who wish to correct a scabbing sand.

1. Try removing the fines from the sand, until permeability exceeds 80.

2. Add sufficient clay to get green compressive strength over 10 psi.



SCREEN ANALYSIS PERCENTAGE SILT (0 TO 50 MICRONS)

Fig. 16-Showing screen analysis percentage of silt.



Pouring Rate, sec 7 Gate Area, Sq In. 0.245

Fig. 17-Photo shows castings made using various pouring rates.

3. Blend bonding clays to get dry strength under

4. Control moisture to optimum temper by means of green deformation or other tests.

5. Add granular organic material to lower high temperature brittleness.

Foundry Conditions That Promote Scabs

Experienced foundrymen have observed that certain foundry conditions tend to cause scabs. When scabs are found on castings it would be wise to check the following foundry conditions before making any changes in the sand.

1. Is the mold rammed unevenly? Is it rammed too hard?

2. Is the sand too wet or unevenly tempered?

3. Is there excessive patching or slicking?

4. Is the mold properly vented?

5. Is the mold poured too slowly? (Fig. 17)

Certain geometrical shapes are prone to scabbing. An increase in casting thickness increases the tendency to form scabs. Abrupt changes in section size will increase the tendency to scab. Internal corners, ribbed or fluted shapes, and castings made with pockets of green sand in the drag are liable to form scabs.

Conclusions

The following conclusions as to optimum properties for non-scabbing green sands for gray iron castings refer to the 47 sets of test castings made with the sands described in this report.

Properties of non-scabbing sands:

1. Green compressive strength should be over 10 psi.

2. Rammed density should be below 110 lb per cu ft.

3. Green deformation should be less than 0.021 in.

.4. Silt content should be less than 8 per cent.

5. Dry strength should be under 150 psi.

6. Hot deformation at 1000 F should be greater than 0.009 in./in.

7. Permeability should be over 50.

Bibliography

1. "Scabbing Tendencies of Molding Sands," A.F.S. TRANSAC-TIONS, vol. 57, pp. 400-408 (1949).

2. H. H. Fairfield, "Expansion of Silica Sand," The Foundry, vol. 76, pp. 128-29, May, 1948.

3. R. B. Sosman, The Properties of Silica, Chemical Catalog Co., Inc., Rheinhold Publishing Co., N. Y. (1927)

4. Analysis of Casting Defects, pp. 71-2, American Foundry men's Society, Chicago (1947).

DISCUSSION

Chairman: C. A. SANDERS, American Colloid Co., Chicago. Co-Chairman: H. W. DIETERT, Harry W. Dietert Co., Detroit. MEMBER: In our foundry we pour castings with flat surfaces and the scabbing defect is one of our problems. Did the Com-

mittee do any work on heavy dry sand work?

MR. ROWELL: Experiments reported in this progress report involved using green sand molds. Experiments with dry sand will be made at some future time.

MEMBER: Did the Committee do any work on distribution of

grain sizes in the sand?

CO-CHAIRMAN DIETERT: The Committee made some tests along that line but the number of tests made was insufficient for conclusions to be drawn. Therefore, no data has been given on the effect of grain distribution and scabbing.

The Committee performed some tests using silica flour in the sand mixture. The compressive strength of this sand mixture was 7.0 psi which is not considered low green strength. We also found it did not have to be silica flour which imparted scabbing tendencies to the sand mixture. It could be a high fines content with high siliceous content. It is more a matter of how much fines you have in the sand. It is desirable to control the very

fine siliceous grains present in a sand.

NATHAN LEVINSOHN: 1 We make castings weighing from a fraction of a pound to about 1,500 lb. We had the scab defect with us a short time ago for a period of three days. What the cause of it was, we do not know. We did notice that our moisture content dropped as our permeability values went up. The moisture content dropped from 4.4 per cent to 3.0 per cent and the permeability increased from approximately 65 to approximately 85.

MR. ROWELL: That contradicts some of our data. There must have been some other controlling factor. We found that high permeability and low moisture tended to relieve the scabbing.

MEMBER: Sand that is undertempered at the time of pouring will cause that scabbing condition. These are expansion scabs.

Mr. ROWELL: With a high permeability and low moisture content it is possible those molds might have been rammed harder.

Mr. Levinsohn: Our normal ramming is 70 to 80 and these had tested 90.

MR. ROWELL: That is probably the cause. You have raised the hot strength.

E. C. ZUPPANN: 2 We pour chilled plowshares, which are some of the worst bucklers. These are 18-lb castings and take more than 20 sec to pour. We ran into the scabbing defect recently. I attributed the cause to the efficiency in our dust collection system. Maintenance men repaired the equipment, increasing its efficiency so that the wood flour disappeared, allowing higher permeability, requiring less moisture. The combination results

Mr. Levinsohn: Molding sand grain size tests are made at our plant at least three times per week. The grain distribution of our sand remained the same when we encountered the scab defect as well as when it disappeared.

Mr. Rowell: In some experiments we rammed the mold to a hardness of 25, 50, 64 and 82. As the mold was rammed harder

the scab was more aggravated.

S. CMUNT: 8 We are using a synthetic sand mixture in our foundry and on any flat surface casting we make a scab appears. Many of our molds should be rammed to a mold hardness of 85 to 90, but even at that hardness they will form scabs. We found a solution to the problem. If you place graphite flakes in carbon tetrachloride and spray the flat surfaces of the mold with this suspension, scabs will not form on these surfaces even though the mold be rammed to 80, 85 or 90 hardness.

MR. ROWELL: I do not know the reason for the way these sprays affect the sand. A few years ago we were having an occurrence of scabs on transmission housings. This did not harm the quality of the casting but it did not look well, so we rammed

¹ Foundry Supt., Minneapolis-Moline Power Implement Co., Minneapolis.

⁸ Foundry Engr., The Oliver Corp., South Bend, Ind. ² Molding Foreman, Superior Foundry, Inc., Cleveland.

the mold softer. Then we had to lower the hot strength of the sand mixture and we eliminated the scabbing tendency.

M. E. Roe: 4 Mr. Levinsohn stated that as the moisture content in his sand mixture went down, the permeability went up. If that is the case, then the sand mixture must be on the plus side of temper and the flowability must be down. If you were to plot permeability as rising, flowability would go in the opposite direction.

MR. ROWELL: That can go either way. You can bond the sand with all bentonite to attain strength and your moisture or temper requirements are much lower. That is a little hard to predict. This green deformation test might have application.

We have advocated several suggestions in our conclusions. The green compressive strength should be over 10 psi. Ram density should be at least 10 lb per cu ft. Green deformation should be less than 120. Dry compression should be less than 150 psi.

We have been unable to find any one or two or three tests that always hold true. In every chart in the paper there are exceptions. We should do more work on these exceptions and correlate them with the other properties to see if they fall in line. I suspect they do, although I expect what is an exception on one property more than conforms on the other.

CO-CHAIRMAN DIETERT: Here is a point we can make. Remember, the reason why the sand loses its moisture was because of the possibility that the suction system pulled out some of the cushioning material, for example, wood flour. In a case like that permeability would increase while the moisture content would decrease.

W. H. Moore: ⁵ If you lost colloidal clay, the permeability would increase and hot toughness would decrease.

H. LOUETTE: ⁶ When we conducted tests for the purpose of eliminating scabs from flat surfaces on our castings we introduced wood flour (130 fineness) into our sand mixture. Since then we have practically eliminated all scabbing. We also sprayed wood flour on the pattern instead of the parting. This also improved the condition.

Another foundryman tried spraying the wood flour on the

pattern but he still had the scab defect. Then he sprayed it on the mold and the scab defect vanished. Sometimes it works on one job and not on another. I would suggest the use of wood flour either on the pattern or the mold and also the regular way.

CO-CHAIRMAN DIETERT: With our present day information in the literature and in this report, the scabbing tendencies of sand may be reduced when we have a number of sand properties at specified values.

J. A. RASSENFOSS: ⁷ Several foundrymen made the statement: "We rammed to a certain mold hardness and we did not get so and so." I think something has been forgotten. Each of you probably has a sand which has a different green strength. It is unlikely any or all of you had the same strength. It will range from 6 to 15 psi green compressive strength.

Now each one of those sands is not rammed to the same degree if you bring it to the same hardness. It is quite unlikely that the low sands could ever be brought up to the hardness of the high sands. Consider a 3 per cent bentonite sand. You can pound on it all day long, but the mold hardness will not go above, say, 75. But, if you increase the bentonite content of that sand to 7 per cent you are capable of getting a mold hardness of the order of 90.

If you take a given sand and ram it to get the mold hardness, you know how well rammed that sand it. You do not know whether it is rammed as well as some other sand which may have a different characteristic hardness. If a foundryman is used to getting a certain mold hardness with his sand and his sand has actually changed in composition, an attempt to maintain that hardness may result in over-ramming. The increase in density may be out of proportion to the increase in hardness number.

⁴ Technical Repr., Archer-Daniels-Midland Co., Cleveland.

⁵ Foundry Engr., Meehanite Metal Corp., New Rochelle, N. Y.

⁶ Asst. Supt., Warden King Ltd., Montreal, Que., Canada.

Research Metallurgist, American Steel Foundries, E. Chicago, Ind.

METAL PENETRATION

Research Progress Report

By

S. L. Gertsman and A. E. Murton *

This report presents a statement of the progress which has been made in an investigation of metal penetration of green sand cores. The decision to undertake this work was made at a meeting of the Mold Surface Committee of the A.F.S. Sand Division held at Cleveland, in May, 1950. At that time it was agreed that the standard casting, developed to evaluate oil sand cores, was to be employed for the tests. The preliminary tests were to be made at the Mines Branch Laboratories, Ottawa, Canada.

Procedure

Close control of the moisture content of the test cores is essential if significant results are to be obtained. Such control was established by means of the following procedure:

- 1. Controlling the moisture in the core mixtures,
- 2. Storing the prepared cores in a humid atmosphere until they were used,
 - 3. Using a cold mold,

4. Installing the cores and closing the mold at the latest possible moment before pouring.

Constant storage humidity was ensured by means of a desiccator which contained water instead of desiccant. Under these conditions, where the relative humidity was 100 per cent, it was found that the moisture content was constant over at least a 2-hr period, which was more than sufficient time for the test. The moisture contents of green sand cores after storage at different humidities are shown in Table 1. The results for relative humidities of 25, 50 and 75 per cent were obtained by using constant humidity solutions of sulphuric acid and water in the bottom of the desiccators.

A means of holding the cores in the sand cavity had to be devised. The method selected was to fasten each

TABLE 1-MOISTURE CONTENT OF STORED GREEN SAND CORES IN PER CENT

Relative Humidity		Storage Tin	ne, Hours	
Per Cent	0	. 1/2	1	2
25	3.45	2.6	2.4	2.1
50	3.45	2.7	2.5	2.2
75	3.45	3.0	2.8	2.4
100	3.45	3.45	3.45	3.45

core to the base core by a wire. The wire was rammed up in the test core in a double-end rammer, and reached to within ½ in. of the top of the rammed core (see Fig. 1). In preparing the test core the wire was passed through a hole in the stripping post, and out a

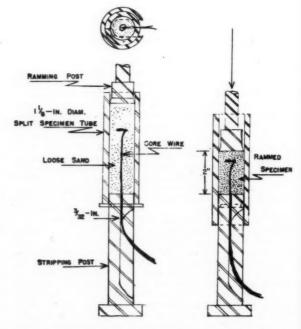


Fig. 1-Sketch shows setup for preparation of green sand cores for metal penetration test.

^{*} Physical Metallurgist and Metallurgist, respectively, Physical Metallurgy Division, Department of Mines and Technical Surveys, Mines Branch, Ottawa, Ontario, Canada.

This report is sponsored by the Mold Surface Committee, Sand Division, American Foundrymen's Society. Members of this Committee are as follows: S. L. Gertsman, Chairman, C. C. Sigerfoos, Vice-Chairman, D. C. Williams, C. Locke, W. T. Shute, J. B. Caine, A. J. Kiesler and H. H. Fairfield.

slot in the side. This allowed free vertical movement of the wire as the core was rammed. In the specimen tube, three-quarters of the weight of the core sand was placed below the loop at the top of the wire, and the rest above.

In order to tie the wires it was necessary to remove the supporting molding sand below the base core. The core thickness had to be increased over that used in oil sand work in order to prevent the metal from breaking through the core below the runner when high metal heads were used. The wires were tied to steel rods. The method of assembling the mold is shown in Fig. 2, and the core assembly in Fig. 3.

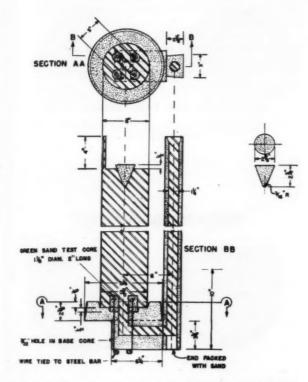


Fig. 2—Sketch shows metal penetration test casting for green sand cores.

Effect of Metal Head and Temperature

A series of tests was carried out to determine the temperatures and pressures required to produce metal penetration in green sand cores. The following mixture (No. 5 of the Appendices) was used:

New Jersey No. 57 Sand, gm 2000 Western Bentonite, gm 100 Moisture, % 31/2

Figure 4 is a photograph of a section which illustrates the type of penetration obtained on green sand cores. This sample was cut vertically through the core taken from a 30-in. casting poured at 2900 F. It will be seen that the core was partially penetrated, with the greatest penetration occurring on the side nearest the ingate (right side of picture).

Figure 5 shows the results obtained with 5, 24 and 30-in. high castings. The following will be observed:

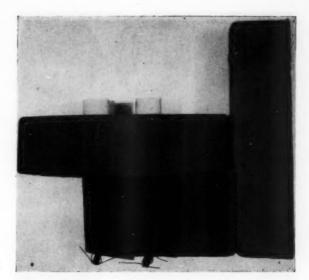


Fig. 3—Core assembly for testing green sand 1-1/8-in. diam by 2-in. test cores.

- 1. The higher the casting (ferrostatic pressure) the lower the pouring temperature at which complete penetration is obtained. For example, in comparing the last vertical row of castings in Fig. 5, the 5-in. shows bad burn-on at 3100 F. The 24-in. shows complete penetration at 3050 F, whereas the 30-in. casting is completely penetrated at 2950 F.
- 2. For the same temperature, 2800 F, (first vertical row in Fig. 5), increasing the height of the casting increases the metal penetration. The 5-in. is not penetrated. The 24-in. has about 1/16 in. penetration; and the 30-in. casting has ½ in. of penetrated metal.
- 3. For each height of casting there is a temperature below which further reduction of temperature has little effect on penetration. Raising the pouring temperature above this temperature, however, produces in-

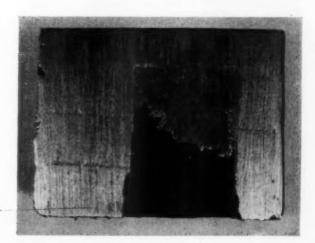


Fig. 4-Section of penetrated green sand core; 30-in. casting, poured at 2900 F. (Outside edge of casting at left, ingate side at right.)



Fig. 5-Photo shows effect of metal head and temperature. Green sand cores.

creased metal penetration. Thus the degree of penetration of the first and second vertical rows of castings in Fig. 5 is about the same, despite the difference in pouring temperature. However, the 50-degree increase between the second and third rows increases the penetration noticeably.

4. Within the temperature range of 2800 to 3100 F, and the pressure range obtained between heights of 5 to 30 in., it would appear that pressure is the more potent factor causing metal penetration. The 5-in. casting has to be poured at 3100 F to produce burn-on, whereas at 2800 F increasing the height of the casting from 5 to 30 in. is enough to cause approximately ½-in. penetration.

Effect of Core Composition

In this series, casting heights of 24 and 30 in. were used. These two heights permitted full comparison of the cores tested at 2850 F. Mixtures were made with New Jersey No. 57 sand. Some of the cores were coated with a silica flour wash; the rest of the cores were used in the unwashed condition. The properties of these cores are shown in Appendix III.

Seven different core types were used:

- 1. Green Sand, (Mix No. 5)
- 2. Oil sand unwashed, (Mix No. 7)
- 3. Oil sand washed (Mix No. 7)
- 4. Dry sand 5% Bentonite unwashed, (Mix No. 5)
- 5. Dry sand 5% Bentonite washed (Mix No. 5)
- 6. No. 57 sand plus silica sol binder unwashed, (Mix No. 6)
- 7. No. 57 sand plus silica sol binder washed, (Mix No. 6)

The cores bonded with silica sol were prepared to determine the behavior of cores made entirely of silica and no other additions. The method of preparing the cores, which is not recommended for industrial use, is shown in Appendix II.

The cores used with the 24-in. casting (first vertical row left in Fig. 6) can be arranged in the following order of merit:

- 1. Washed silica sol binder,
- 2. Unwashed silica sol binder, washed dry sand 5 per cent bentonite,
 - 3. Washed oil sand,
 - 4. Green sand,
 - 5. Unwashed dry sand 5 per cent bentonite,
 - 6. Unwashed oil sand.

With the 30-in. casting (second vertical row right in Fig. 5) the washed silica sol core gave the best results. The green sand cores were next best, but all the other penetrated badly.

It is interesting to note that the green sand was better than the unwashed oil sand and dry sand cores. Furthermore, although the wash is effective at lower ferrostatic pressures, it loses most of its effectiveness as the pressure is increased.

Effect of Silica Flour Additions

In order to determine the effect of silica flour addi-

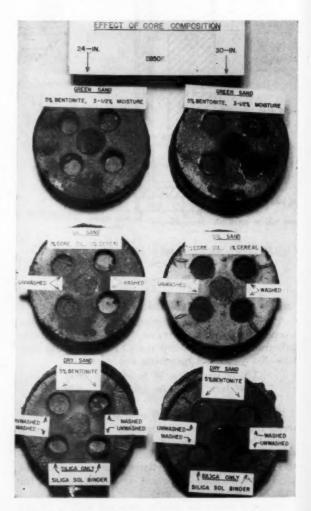


Fig. 6-Photo shows effect of core composition.

tions in green and dry sand mixtures, 36-in. castings were employed at a pouring temperature of 2850 F. New Jersey No. 57 sand, bonded with 5 per cent bentonite, was used in preparing the following core mixtures:

No. 1-10 per cent silica flour,

ıl

g

5

No. 2-25 per cent silica flour,

No. 3-50 per cent silica flour,

No. 4-10 per cent silica flour plus 1.0 per cent iron oxide.

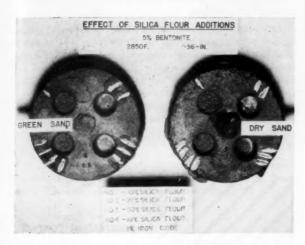


Fig. 7-Photo shows effect of silica flour additions.

Figure 7 illustrates the improvement in the resistance to penetration obtained with No. 2 and No. 3 cores containing 25 and 50 per cent silica flour respectively in both the green and dry sand cores. The No. 2 green sand core is better than the No. 2 dry sand core. The No. 3 cores having 50 per cent silica flour gave a very smooth surface in both the green and dry sand condition.

Metallographic Examination

Samples were cut vertically through the center of the cores along the center line of the casting. These samples, which were taken from the side adjacent to the ingate, were mounted in bakelite and polished. The parent metal and the metal penetration compact (if any) are shown in the photomicrographs (Fig. 8 to 19 inclusive, except Fig. 12) taken at 75 diameters.

Figure 8 illustrates the sand, metal and slag of the metal penetration compact at the air surface for the oil sand cores poured at 2800 F in a 30-in. casting. Severe penetration has occurred. Figure 9 is from the same sample, but at the metal surface. It will be observed that at the metal surface no slag is present, only sand and metal. Figures 10 and 11 are views of the penetrated compact of a green sand core. This core was used in a 30-in. casting poured at 2800 F, and was penetrated about ½ in. Near the air surface of the penetrated compact (Fig. 10) slag is present as well as sand and metal; but near the metal surface (Fig. 11) only sand (fused probably due to bentonite) and metal are present.

The slag occurrence in all the penetrated samples



Fig. 8—Photomicrograph of air surface of penetrated compact of 30-in. casting poured at 2800 F. The metal is white, slag is light gray, sand is slate gray and voids are black. Oil sand. 75X.

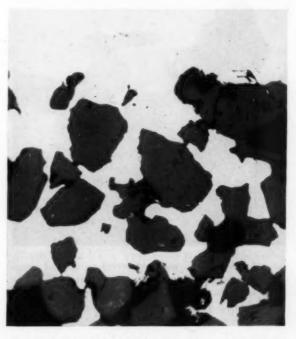


Fig. 9—Photomicrograph of metal surface of penetrated compact of 30-in. casting poured at 2800 F. The metal is white, sand is slate gray and voids are black.

Oil sand. 75X.

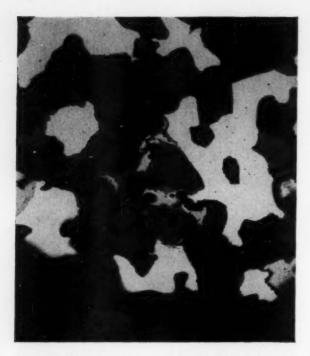


Fig. 10-Photomicrograph of air surface of penetrated compact of 30-in. casting poured at 2800 F. The metal is white, slag is light gray, sand is slate gray and voids are black. Green sand. 75X.

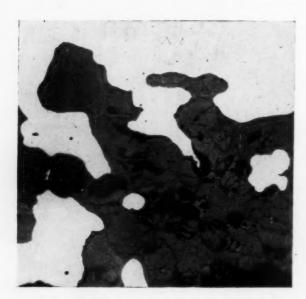


Fig. 11—Photomicrograph of metal surface of penetrated compact of 30-in. casting poured at 2800 F. The metal is white, sand is slate gray and voids are black.

Green sand. 75X.

examined was similar to that observed above. This was true of oil sand, green sand, and dry sand in the pouring range of 2800 to 3100 F. In all cases slag occurred only near the air surfaces, either of the metal itself

(where no penetration occurred) or of the penetrated compacts. No slag was observed near the surface of the metal when it was covered with a mass of penetrated sand. Figure 12 is a sketch illustrating the above.

These observations suggest the following possibilities:

- The slag was formed and built up ahead of the metal while it was penetrating.
- The slag was formed by a reaction between the oxidizing fingers of metal and the sand after the penetration had occurred.

The influence, if any, of the slag (probably fayalite) on the penetrating action of the metal is not known. The work of Hoar and Atterton² appears to indicate that the slag, if it is considered to be fayalite, has at most a minor influence on inducing penetration, as they found that metal will not wet either sand or fayalite. In fact, these investigations² have shown evidence that slag helps to prevent penetration by plugging the voids.

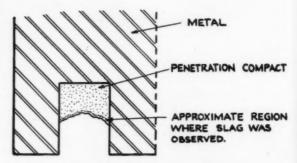


Fig. 12-Position of slag occurrence.



Fig. 13—Photomicrograph of metal surface of penetrated compact of 30-in. casting poured at 3050 F. The metal is white, sand is slate gray and voids are black.

Oil sand. 75X.



Fig. 14—Photomicrograph of metal surface of penetrated compact of 24-in. casting poured at 3050 F. The metal is white, sand is slate gray and voids are black.

Green sand, 75X.



Fig. 15—Photomicrograph of mold-metal interface of 5-in. casting poured at 3100 F. The metal is white, slag is light gray, sand is slate gray, voids are black and bakelite is mottled gray. Oil sand. 75X.

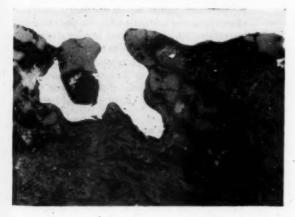


Fig. 16—Photomicrograph of mold-metal interface of 5-in. casting poured at 3100 F. The metal is white, sand is slate gray and bakelite is mottled gray.

Green sand. 75X.



Fig. 17—Photomicrograph of mold-metal interface of 5-in. casting poured at 2800 F. Metal is white, slag is light gray, sand is slate gray, voids are black and bakelite is mottled gray. Oil sand. 75X.



Fig. 18—Photomicrograph of mold-metal interface of 5-in. casting poured at 2800 F. Metal is white, sand is slate gray and bakelite is mottled gray.

Green sand. 75X.



Fig. 19-Photomicrograph of mold-metal interface of 5-in. casting poured at 2800 F. Metal is white, slag is light gray, sand is slate gray, voids are black and bakelite is mottled gray. Dry sand. 75X.

Figure 13 (oil sand 30-in. poured at 3050 F) and Fig. 14 (green sand 24-in. poured at 3050 F) show the fused sand adjacent to the parent metal. This fusion is caused by the high pouring temperature. There was no such continuous fused layer at the surface when the same cores were penetrated at a lower temperature (see Fig. 9 and Fig. 11). It should be stated that despite this

fused surface layer more severe penetration occurred at the higher temperature. The possible relationship of this fused layer to penetration and peel is discussed

Figures 15 and 16 compare the oil sand and green sand cores in 5-in. castings poured at 3100 F. It will be noted that under these conditions the green sand core was worse than the oil sand.

Figures 17, 18 and 19 compare the results from 5-in. castings poured at 2800 F using oil sand, green sand, and dry sand cores respectively. It will be observed that there is a continuous layer of slag on the metal adjacent to the dry sand core. The oil sand core showed a discontinuous and porous slag layer. Little slag was observed with the green sand cores. Of the three cores, the dry sand core appeared to give the smoothest contour.

Possible Mechanism of Metal Penetration

Based on this work and that of other investigators a possible mechanism of metal penetration may be postulated. What happens when the metal comes in contact with the sand? It is believed that the mechanism may vary somewhat under different conditions as shown below:

1. Take the case where the metal is cold and there is no appreciable head on the casting. The probability is that a smooth surface will result due to the fact that the metal freezes immediately as it comes into contact with the cold sand. It is possible for fayalite or slag to form on interaction between the metal and the sand, but the fayalite peels off since there are no metal prongs

holding it to the casting surface.

2. Suppose hot metal is poured into a casting having no appreciable head. As the metal makes contact with the sand, some sand fusion occurs. This will cause some voids to appear in the sand. The hot metal will remelt any skin that is formed and then will penetrate into the sand voids. The extent of the metal penetration will depend on the size of the voids, and the depth to which the sand is heated above the freezing point of the metal, and possibly on the metal surface tension²

3. If cold metal and a high ferrostatic head is used, the pressure can force the metal into the interstices of the sand core without any fusion occurring. Here a finer sand should help to prevent penetration.

4. If hot metal and high ferrostatic heads are used the voids created, due to sintering or incomplete fusion of the sand, together with the pressure, will produce metal penetration.

Should a completely fused sand face be produced by high pouring temperatures then it is reasonable to expect no penetration. If there were any discontinuities whatsoever then penetration would probably be more severe than at lower pouring temperatures. The following factors would tend to increase the penetration:

1. Possible lower surface tension of the hot metal, 2. The metal being fluid for a longer period of

3. The possibility of a zone of larger voids in the sand, where agglomeration due to partial sintering has occurred behind the fused layer.

In cases where metal has penetrated through a con-

tinuously fused layer of sand, as in Fig. 13 and 14, the penetrated compact might reasonably be expected to separate easily from the metal, except in internal cavities where it would lock itself in. This view is supported by the fact that metal has been observed in compacts which were easily peeled from outside casting surfaces.

Hoar and Atterton² found that for three different sands the amount of pressure necessary to cause penetration by tin decreased with increasing pouring temperatures up to 1650 C (3000F). Above this temperature, the pressure required to cause penetration increased sharply as complete fusion of the sand formed a glassy surface. These results have not been confirmed in this investigation (see Fig. 5). Three possible explanations may be advanced to account for difference:

1. The larger surface area exposed by our core might make it more difficult to produce a continuous

fused surface.

2. The techniques of applying temperature and pressure differ for the two methods. The method used by Hoar and Atterton might tend to present a fused surface to the metal before pressure is applied.

3. Perhaps with the sands used in this investigation the maximum temperature tested was not high enough

to produce a completely fused layer.

Summary

- 1. Green sand cores can be stored at least 2 hr at 100 per cent relative humidity without appreciable change in the moisture content.
- 2. A wire was rammed into each of the 11/8-in. diameter by 2-in. green sand cores in order to tie the cores down in the mold cavity.
- 3. As an added safety precaution, a thicker base core was used than was employed in the previous oil sand work.1
 - 4. For green sand cores it was found that:
 - a. The higher the ferrostatic pressure the lower the pouring temperature at which complete penetration was obtained.
 - For the same temperature increasing the casting height increased the metal penetration.
 - For each casting height there appears to be a temperature below which further reduction of temperature has little effect on penetration, but above which increased temperature produces increased metal penetration.

d. Within the ranges of temperature (2800 to 3100 F) and pressures (heights of 5 to 30 in.) tested, it would appear that pressure is the more potent factor causing metal penetra-

sh

ca

ha

me

kn

ne

me

ve

vol

tion.

5. a. With the limited number of tests carried out the washed silica sol core used gave good results up to 30 in. This core penetrated completely at 36 in.

b. Green sand cores were better than unwashed

oil sand and dry sand cores.

- c. Although the wash was effective on the oil sand and dry sand cores at lower ferrostatic pressures, it lost most of its effectiveness as the pressure was increased.
- 6. The addition of 50 per cent silica flour gave a

very smooth surface in both the green and dry sand condition at a height of 36 in.

 Metallographic examination revealed slag at the air surface of the penetrated compact but not at its metal surface.

8. There was a fused layer of sand adjacent to the metal when penetration occurred at higher temperature, but none at lower temperature. This fused layer did not prevent the metal penetration from being greater at the higher temperatures.

References

1. S. L. Gertsman and A. E. Murton, "An Investigation of Metal Penetration of Steel Sand Cores," A.F.S. Transactions, vol. 58, pp. 595 to 603 (1950).

2. T. P. Hoar and D. V. Atterton, "Penetration of Molten Metal into Compacted Sand," *Journal of the Iron and Steel Institute*, vol. 166, part I, pp. 1 to 17, September, 1950.

Appendix I—Screen Distribution of New Jersey No. 57 Sand

U. S. Screen No.	Per Cent Retained
30	0.5
40	2.2
50	11.7
70	39.5
100	38.8
140	6.0
200	0.8
270	0.3
Pan	0.1
A.F.S. Clay	Nil

APPENDIX II-MIXTURES USED IN PENETRATION TESTS

Mix No.	Moisture Per Cent	Sand	Bent- onite	Cereal	Oil	Silica Flour	Iron Oxide	18% Silica Sol
1	4.6	1800	100	**		200		
2	5.0	1500	100			500		
3	7.0	1000	100			1000		
4	5.0	1800	100			200	20	
5	4.0	2000	100					
6		2000		4.				See note
7	4.0	2000		20	20		* *	

Note: The cores bonded with silica sol were prepared to determine the behavior of cores made entirely of silica with no other addition. The method of preparing the cores, which is not recommended for industrial practice, was as follows:

Test cores were prepared from sharp sand tempered to 4 per cent moisture with a solution which contained 18 per cent silica by weight. The cores were dried and resaturated with a silica sol solution to which had been added an accelerator consisting of a 10 per cent solution of ammonium chloride added to the silica sol in the ratio of 9 parts of sol to 1 part of ammonium chloride. The accelerator was added to prevent the silica from migrating to the surface of the cores on drying. After the gel had set the cores were again dried.

C. E. Wenninger: 2 I would like to ask the author whether he tried to counteract pressure with pressure. Did he try to increase moisture to see if it would create a little more steam head with possibly a little more interior pressure being exerted against the metal?

MR. GERTSMAN: No, we have not tried that at all, but that brings up Mr. Locke's paper in regard to mold gases. (Charles

APPENDIX III-PROPERTIES OF CORE MIXTURES

Mix	Mois-	Green Permea-	Green Compres- sive Strength,	Dry Compressive Strength,	Flowa- bility Dietert	p	rength, si min)	30-lb Collapsi- bility 2000 F.	Weight of 11/8 in. Dry Specimen.
No.	%	bility	psi	psi	No.	1000 F	2000 F	sec	grams
1	4.6	99	4.8	112	83	238	220		49
2	5.0	39	8.7	182	75	485	830		53
3	7.0	7.7	12.2	208	72	648	985		55
4	5.0	80	6.3	148	85	296	562		50
5	4.0	138	4.2	93	87	128	502	* *	49
6		178	0.43	10	90	24	80		48
7 (a)	4.0	88	0.59	(b)	92	32		18	49
	nd baked 450	F for 40 min.		. ,	(b) Baked	tensile streng	th 185 psi.		

DISCUSSION

Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

J. B. CAINE: ¹ Mr. Gertsman stated that the test is a very severe test. I think there are many castings we are having trouble with today in production in which the penetrating conditions are more severe than in this test. Mr. Gertsman has shown when the sand-metal ratio is low, pressure is the major cause of penetration. This checks out in practice where you have very small cores surrounded by let us say, 10 or 12 in. of metal with easily 6 ft of hydrostatic head on large castings. We know that such cores will penetrate completely. Perhaps we need to study the resistance of sand to penetration under even more severe conditions.

MR. GERTSMAN: The only reason I mentioned that we have a very severe test is that in most cases where you are dealing with green sand you have a large surface of green sand facing the metal. Here we have a small core with very little surface or volume, surrounded completely by metal. Of course, when you are dealing with re-entrant angles you are correct, you have a very severe condition whether it is green sand or not.

Locke and R. L. Ashbrook, "Nature of Mold Cavity Gases," A.F.S. TRANSACTIONS, vol. 58, pp. 584-594 (1950). I believe he showed in one of his charts that the counteracting pressure due to gas evolved from the mold was in the neighborhood of about 6 in. of steel head. Your suggestion however is a good one and should be tried.

V. E. ZANG; ^a I have used this particular test to evaluate core mixes and found it applicable to shop practice. We went one step further by inserting a test core into the runner and into the down sprue to test erosion on those core mixes. We found that we get very good cores in regard to penetration but we can erode them very fast.

MR. GERTSMAN: This casting as it was designed was set up solely for metal penetration. You have put in an improvement which I think is a very good one. You are going to be able to determine whether you have a core mix that is satisfactory for metal penetration and also whether it is satisfactory for

¹ Foundry Consultant, Wyoming, Ohio.

² Professor, University of Kentucky, Lexington, Ky.

⁸ Vice President, Unitcast Corp., Toledo, Ohio.

t

erosion, this latter being indicated by the core which you have put in across the runner. You then know whether you can use that core mix at a point in the casting where erosion conditions occur, or whether you can only use that core mix in a casting where erosive conditions are not present.

T. W. CURRY: 4 Recently we have been doing some experimental work on gating and penetration has come into the investigation very strongly. We have been doing some work with the gating procedures that Mr. Dietert prepared in 1937 or 1938. We find that in our gray iron foundry, and we have reason to believe that it would be possible to determine this in any other foundry, that the rate of pour, i.e., the pounds per second of pouring rate for a given metal is very important from the standpoint of penetration. I would like to add that as a suggestion for further work on this study of penetration.

Since October, 1950, we have been able to place about 100 different castings on our gating and risering chart. In no instance have we found an increase in cleaning time. We have as high as 36 per cent reduction in cleaning time, and the reduction has been the result of reduced penetration.

I am referring in one case to a 170-lb drum with a green sand core 14 in. in diam. In this particular case it was necessary for us to build a new pattern because we had eliminated all swell and burned-on sand. When we adjusted the pouring rate in pounds per second, penetration almost completely disappeared on production runs of 5,000 castings. As far as perceptible swell prior to gate change that you could determine as a lump or a knot or some such flaw on the castings, there were none present. There was a perceptible overall swell to bring the casting to the proper dimensions and the swell was accompanied by penetration.

We have had several other castings where we have 14 to 16-in. high walls and only 1/4-in. sections-I am speaking of the typical motor-frame type of castings-and we had penetration to great extent even in the 1/4-in. sections. They have been greatly improved by gating but penetration has not been entirely

We believe that penetration can be greatly reduced if a study is made of the fluidity characteristics of the metal you are working with and if you will determine the proper rate of flow into the mold.

MR. GERTSMAN: As far as rate of flow is concerned, I think that might have something to do with the corrosive action. In regard to fluidity, Hoar and Atterton in England, dealing with steel, have shown that fluidity does not have much effect on metal penetration. It should be stated however, that this was shown by a mathematical attack and as with all mathematical attacks, it should be taken with some reservation until more facts are established.

MR. WENNINGER: If you reduced the gate area it would take a longer time to fill your casting. Do you not believe you would probably show more penetration due to having a longer time of heat transfer?

MR. GERTSMAN: If there is a longer time for heat transfer and there is sufficient pressure available to cause the metal to pene trate, then if the core heats up so that the thickness of the core which is above the melting point of the metal is increased. then you would expect the metal to travel greater distances within the pores of the sand and the penetration compact to be greatly increased.

CHAIRMAN RASSENFOSS: There is one other possibility too, as Mr. Curry mentioned, the kinetic energy of the molten metal. 1 would be interested in knowing what were the differences, Mr. Curry, between good and bad pouring rates on the particular casting you mentioned.

Mr. Curry: The 170-1b drum was poured in about 20 sec and we increased the pouring time to 31 sec. Insofar as penetration is concerned, in reply to the last remark, I cannot visualize taking a metal with a melting point as high as steel and cast iron would have and placing it against a mold face without getting some thin chilling effect on that surface. I think that is where our results have been centered, because we have been able to improve the finish from green sand cores and mold walls by the rate of flow, and drastically so. The temperature of the metal in the mold cavity when you restrict the pouring rate down to that which would be compatible with its fluidity characteristics is lower. I am sure that the metal going into the mold from the gating system would have a lower temperature in the mold than what it would be with the very rapid type of pour. I do not believe that a skin chill, if a very thin one exists, would be as heavy in the case of a rapid pour as that of a relatively slow pour within the range of the fluidity of the metal.

MR. GERTSMAN: I want to point out that if a skin forms and gets thick enough so it does not remelt, then you cannot have metal penetration, i.e., if it is thick enough to withstand the particular effective pressure. Insofar as gray iron is concerned, I might say that we carried out a series of tests on the effect of pouring temperatures at different heights and we found that there is very little effect of pouring temperature on the penetration of gray iron at different heights. We found, however, that for the same temperature, increasing the height, produced metal penetration when we used coarse sands. When fine sands (175 A.F.S.) were used, we went up to 44 in. in height, which is the limit of our facilities and no penetration occurred. Temperature did not have much effect on the penetration characteristics of the cores which we tested for gray iron.

Pressure did have an effect similar to that observed with steel, but we had to use coarse grain sands in order to produce penetration for the maximum height which our facilities allowed

MR. CAINE: I think Mr. Curry has expressed an important misconception of metal flow regarding this formation of a skin. There is abundant evidence today from all the motion pictures of metal flow that a skin does not form immediately, while metal is in motion. It has been checked and rechecked that a skin does not form on discontinuous flow because you can see the sand when it is uncovered, or the metal runs off of it.

⁴ Metallurgist, Lynchburg Foundry Co., Lynchburg, Va.

SIEVE ANALYSES OF SILICA SANDS

By

A. I. Krynitsky and F. W. Raring*

A STUDY OF SIEVE ANALYSES of silica sand was undertaken to determine the precision of the American Foundrymen's Society fineness test for foundry sands. The primary purpose of the investigation was to determine whether the type of sieve shaker used has any significant effect on the analytical results for a given sample, and the reproducibility of a sampling technique.

The effect of various factors on the results of particle size determinations made by means of sieving tests has been investigated and the results described in several publications. Shergold¹ investigated the effect of sample size and has made recommendations which were adopted in the most recent British standard method for sieve analysis.

Mörtsell² has reported on a series of experiments involving types of sieves, sample size and sieving time effects, comparative tests with different types of sieve shakers, and reproducibility of sieve analyses. Fahrenwald and Stockdale³ investigated the relative efficiencies of several types of sieve shakers, and Swinford⁴ has presented a statistical treatment of data obtained with vibratory and rotary type sieve shakers. For further details see Foundry Sand Testing Handbook, 1944 ed., published by American Foundrymen's Society, 616 S. Michigan Ave., Chicago 5, Ill.

Materials and Testing Techniques

The sand used in the experiments was a washed New Jersey silica sand with an A.F.S. fineness of about 70 and with a particle size distribution (average percentage retained on different sieves) as indicated in Table 1.

The A.F.S. fineness number is obtained by using the standard A.F.S. set of 11 sieves. It is computed by multiplying each weight fraction percentage by a factor, adding the products and dividing the total product by the total percentage of retained sand particles on the different sieves. The multiplying factors are as follows:

c as follows	3.		
U. S. Series		U. S. Series	
Sieve No.	Multiplier	Sieve No.	Multiplier
6	3	70	50
12	5	100	70
20	10	140	100
30	20	200	140
40	30	270	200
50	40	Pan	300

A 5000-gram batch of this sand was prepared by drying at room temperature for 18 hr, passing through a No. 8 sieve six times and through a riffle eight times in order to mix the sample thoroughly. The sand was then spread on a sheet of glossy paper to a thickness of about one inch and covered to prevent dust

accumulation. Samples were drawn with a spatula in such a way that each sample represented many sections of the heap. Sample size was fixed at 50 grams and all samples were dried at 105 C prior to analyzing. Although this sampling procedure was used for the convenience of the authors it is recognized that it may not be the most accurate reducing procedure.

One set of sieves was used in all cases, sieving time was fixed at 15 min, and techniques followed in collecting sieve fractions were kept constant throughout. One extra precaution was to place the empty sieves in the sieve shaker prior to every analysis, subject them to a 15-min period of shaking, and remove any particles knocked loose. It was reasoned that this precaution would minimize sample contamination from partially blinded sieves.

Two commonly-used types of sieve shakers were used in the experiments: type (A) sieve shaker accomplishes sieving action by eccentric rotation and tapping. The nest of sieves is rotated at the rate of 285 rpm and tapped at the rate of 150 blows per min. This apparatus is driven by a ½-hp motor to which both the tapper and rotating shaft are geared, and is controlled by an automatic time switch. Type (B) shaker (Fig. 1) consists of a suspended wooden framework into which the nest of sieves is clamped, and with a 1/6-hp motor mounted on top. The motor drives an unbalanced wheel at the rate of 860 rpm, which causes oscillation of the entire assembly and sets in motion tappers mounted on the sides.

Difficulty was at first encountered with this apparatus in that the individual sieves tended to rotate relative to each other while the shaker was in operation. The result was that metallic particles were ground from the contacting surfaces of the sieves; these particles entered the sample and, being very fine, were retained on the No. 270 sieve or the pan. This source of error was eliminated by bracing the nest of sieves with an aluminum strip (Fig. 2) which did not interfere with the operation of the apparatus but held the sieves in the same relative positions.

Procedure and Results

The reproducibility of the sampling technique employed was determined by analyzing 16 samples and applying statistical methods to the interpretation of the data. These samples were drawn one at a time without mixing the batch between drawings. The analyses were made with a type (A) sieve shaker and the data are tabulated (Table 1) in terms of both A.F.S. fineness numbers and percentages retained on the individual sieves.

The A.F.S. fineness number system of describing

^{*} National Bureau of Standards, Washington, D. C.

the particle sizes present in a said has been criticized because of the wide latitude of particle size distributions possible with any one specific fineness number. This shortcoming is of no importance in this investigation, however, because all samples were of the same type of sand and were as nearly alike as the sampling technique permitted. The A.F.S. fineness number system is used in this study because it is a convenient method of epitomizing groups of data and simplifying their interpretation.

The method used in the interpretation of the data can be briefly explained as follows: Indeterminate errors are of varying magnitude for a series of observations. The positive errors, reckoned from the mean or average for the series, must balance the negative errors and the algebraic sum of all deviations from the average is zero. The limits of these errors and the frequency of occurrence of an error of given magnitude can be determined with the data from a few observations in terms of the standard deviation of the entire series from the mean or average value. The standard deviation can be expressed mathematically as follows:

$$\sigma = \sqrt{\frac{(x_1 - \bar{x})^2 + (x_2 - \bar{x})^2 + \dots + (x_n - \bar{x})^2}{N - 1}}$$

where σ = standard deviation

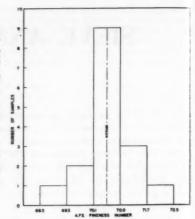
x = average or mean value of all observations

 x_1, x_2, x_3, x_n = values of individual observations

N = number of observations.

According to accepted statistical methods the limits of a large number of subsequent observations, made in the same way, in terms of the standard deviation are $x \pm \sigma$ for 68.3 per cent of all observations, $\bar{x} \pm 2 \sigma$ for 95.6 per cent of all observations, and $\bar{x} \pm 3 \sigma$ for 99.7 per cent of all observations. A complete explanation of the method used here can be found in a standard text on statistics such as Simon⁵.

Fig. 3-Histogram showing the grouping of A.F.S. fineness number determination experimental data about their average. Analyses were made with type (A) sieve shaker.



Data obtained in the experiments (Table 1) yield an average A.F.S. fineness number of 70.5. The standard deviation from this average calculated from the formula above is 0.8.

By the process outlined in the foregoing the limits of a large number of subsequent analyses made by the same procedure are as follows:

Limits	Percentage of Tota
$\bar{x} \pm \sigma \text{ or } 70.5 \pm 0.8$	68.3
$\bar{x} \pm 2\sigma$ or 70.5 ± 1.6	95.6
$\bar{x} + 3a$ or $70.5 + 2.4$	99.7

For practical purposes the largest error which can be expected is therefore \pm 2.4 A.F.S. fineness numbers or \pm 3.4 per cent. This error is due to both sampling and sieving. It should be emphasized that these limits apply when sigma is estimated from a large number of observations and strict adherence to them cannot be expected. The histogram (Fig. 3) of A.F.S. fineness number versus frequency shows the grouping of the experimental data (type A shaker) about its average.

Table 1 -- Data on 16 Samples Analyzed With a Type "A" Sieve Shaker

	_		R	etained	on U.	S. Sieve	e, per c	ent			A.F.S.
Sample No.	No. 20	No. 30	No. 40	No. 50	No. 70	No. 100	No. 140	No. 200	No. 270	Pan	Finenes: Number
18	0.0	0.4	2.0	8.5	34.8	33.3	13.8	4.5	1.3	0.9	70.5
19	0.1	0, 3	1.9	8.6	34.8	33.8	14.1	4.3	1.4	0.9	70.6
20	0.0	0.3	2.1	8.6	34.2	33.6	14. 1	4.6	1.5	1.1	71.5
21	0.1	0.3	2.0	8.5	33.4	34.7	14.4	4.5	1.3	1.0	71.1
22	0.1	0.3	2.0	8.9	34.1	34. 1	13.9	4.2	1.2	0.9	70.2
23	0.0	0.4	2. 3	9.6	35.8	33.0	13.2	3.9	1.2	0.8	69.0
24	0.1	0.3	1.7	7.6	33.0	34.7	15.4	4.8	1.5	1.0	72.5
25	0, 1	0.3	1.8	8.4	33.8	34. 2	14.3	4.6	1.4	1.0	71.4
26	0, 1	0.4	2.1	8.6	34.5	33.7	14. 1	4.2	1.3	0.9	70.4
27	0.1	0.4	2.0	8.7	35.0	33.6	13.8	4.3	1.3	0.9	70.3
28	0.1	0.4	2. 1	8.9	34.5	33.8	14.0	4.1	1.3	0.9	70.2
29	0.1	0.4	2.1	8.7	34.2	34. 1	14. 1	4.2	1.2	0.9	70.2
30	0, 1	0.3	1.9	8.7	34.3	34. 1	14. 1	4.4	1.3	0.9	70.6
31	0.1	0.3	2.0	8.9	35.3	33.8	13.4	4.2	1.2	0.8	69.6
32	0.1	0.3	2.0	9.0	34.9	33.6	13.7	4.3	1. 3	0.9	70.2
33	0.0	0.4	2.1	9.0	35.4	33.4	13.4	4.1	1.3	0.8	69.6

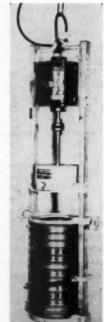


Fig. 1 (left)—Type (B) sieve shaker showing method of suspension and set of sieves in place. Fig. 2 (below)—Nest of half-height sieves and the bracing arrangement used to prevent turning of sieves during operation.

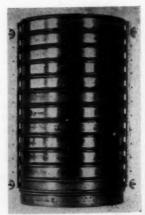


Table 2 -- Data on Repeated Analyses of One Sample

U. S. Sieve	Analysis Number								
No.	1	2 Per	4 Cent Ret	5 ained	6				
20	0.0	0.0	0.0	0.0	0.0				
30	0.4	0.3	0.3	0.3	0.3				
40	2.0	1.9	2.1	1.9	2.0				
50	8.5	8.5	8.6	8.6	8.5				
70	34.8	34.6	33.8	34.0	33.7				
100	33.3	34. 1	34.2	33.9	34.3				
140	13.8	13.8	14.3	14.5	14.4				
200	4.5	4.4	4.4	4.4	4.4				
270	1.3	1.4	1.4	1.4	1.4				
Pan	0.9	0.9	0.9	0.9	0.9				
A.F.S. Fin	eness N	70.7	70.9	71.0	71.1				

d-

ie

ie al

g

S

e

NOTE: There was reason to believe that the results of analysis No. 3 were erratic. Therefore, they were not taken into consideration.

The process followed in checking the error introduced by the sieve shaker was to make repeated analyses of one sample and check the reproducibility. The data of this test are presented in Table 2. One sample was analyzed six times using the type (A) sieve shaker. The results obtained give some indication that there is a very slight increase in the fineness number of the sample with each analysis. When this relation was plotted (A.F.S. fineness number versus analysis number) and the displacements of the points from this curve were used to compute an approximate standard deviation it was found that an approximate maximum error of the type (A) apparatus is \pm 0.4 per cent.

This error may be subtracted from the error due to both sampling and sieving to obtain the error due to sampling alone. According to statistical theory the squares of the deviations are additive; therefore, the error due to sampling alone equals $\pm \sqrt{3.4^2 - 0.4^2} = \pm 3.4$ per cent. The sampling error is thus seen to be by far the largest error for these particular test conditions.

The calculation of the standard deviation in this case is made on the basis of statistical theory described by G. W. Snedecor⁶:

$$\sigma = \sqrt{\frac{\sum (x - x^1)^2}{N - 2}}$$

where $\sigma =$ standard deviation

x = actual fineness number determined

x¹ = assumed true fineness number as determined from the curve.

N = number of determinations.

This method is employed because of the added variable of prolonged sieving. The summation of the squares is divided by (N-2) rather than (N-1) for

reasons explained in the reference. The standard deviation as calculated from the data is 0.1 A.F.S. fineness numbers. It has already been mentioned that according to statistical methods for practical purposes the largest error of determinations may be accepted as being equal to $\pm 3\sigma$. Accordingly this error of type (A) apparatus is $\pm 3 \times 0.1 = \pm 0.3$ fineness numbers, i.e., approximately ± 0.4 per cent with reference to the 70.5 A.F.S. fineness number (Table 2).

Compare Test Machines

The procedure followed in checking the correlation of results of analyses between the two types of sieve shakers was to make two analyses of nine samples. For all nine samples, the first analysis was made with a type (A) sieve shaker, and the second analysis was made with a type (B) sieving machine. The data pertaining to these experiments are given in Table 3 and presented graphically in Figs. 4 and 5. These data

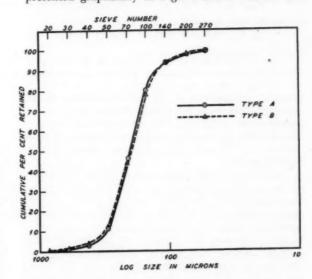


Fig. 4—Cumulative particle size distribution curves for the average values of nine samples analyzed with two types of sieve shakers (see Table 3 and Fig. 5)

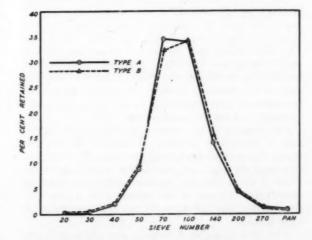


Fig. 5-Data from Fig. 4 are plotted as percentage retained instead of cumulative percentage retained.

Table 3 -- Data on Comparative Analyses of Nine Samples with Two Types (A and B) of Sieve Shakers

								Per	Cent R	etaine	on Si	eve									F.S.
-	20		30		40		50		70	1	00	1	40	20	00	2	70	F	Pan		lo.
A	В	A	В	Ā	В	A	В	A	В	A	В	A	В	A	В	A	В	A	В	A	В
0.0	0.0	0.4	0.4	2.3	2.4	9.6	10.3	35.8	32.8	33.0	33.7	13.2	14.7	3.9	4.0	1.2	1.1	0.8	0.6	69.0	69.
0.1	0.1	0.3	0.3	1.7	1.8	7.6	8.6	33.0	31.5	34.7	34.4	15.4	16.4	4.8	4.8	1.5	1.4	1.0	0.8	72.5	72.
0.1	0.1	0.3	0.3	1.8	2.0	8.4	9.4	33.8	33.3	34.2	33.6	14.3	14.6	4.6	4.6	1.4	1.4	1.0	0.8	71.4	70.
0.1	0.1	0.4	0.4	2.1	2.0	8.9	9.3	34.5	32.2	33.8	34.4	14.0	15.8	4.1	4.2	1.3	1.2	0.9	0.6	70.2	70.
0.1	0.1	0.4	0.4	2.1	2.1	8.7	9.9	34.2	33.3	34.1	34.4	14.1	15.8	4.2	4.2	1.2	1.2	0.9	0.7	70.2	70.
0.1	0.1	0.3	0.3	1.9	2.1	8.7	9.0	34.3	31 3	34.1	34.7	14.1	16.1	4.4	4.5	1.3	1.3	0.9	0.7	70.6	71.
0.1	0.0	0.3	0.3	2.0	2.1	8.9	9.3	35.3	32.0	33.8	34.5	13.4	15.6	4.2	4.2	1.2	1.2	0.8	0.7	69.6	70.
0.1	0.1	0.3	0.4	2.0	2.1	9.0	9.2	34.9	31.0	33.6	34.7	13.7	16.1	4.3	4.4	1.3	1.4	0.9	0.7	70.2	71.
0.0	0.1	0.4	0.4	2.1	2.1	9.0	9.8	35.4	32.2	33.4	34.2	13.4	15.1	4.1	4.2	1.3	1.2	0.8	0.6	69.6	69.
				2.0	2.1	8.8	9.4	34.6	32.2	33.9	34.1	13.9	15.6	4.3	4.3	0.3	1.3	0.9	0.6		
								4= 6													
	A 0.00 0.1 0.1 0.1 0.1 0.1 0.0 Per 0.1 ve A	0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.1 0.0 0.1 0.1 0.0 0.1 0.0 0.1 Per Cent 0.1 0.1 ve Avera	A B A 0.0 0.0 0.4 0.1 0.1 0.3 0.1 0.1 0.4 0.1 0.1 0.4 0.1 0.1 0.3 0.1 0.1 0.3 0.1 0.1 0.4 0.1 0.1 0.3 0.1 0.0 0.3 0.1 0.0 0.3 0.1 0.1 0.3 0.0 0.1 0.4 Per Cent Retai 0.1 0.1 0.3 ve Average Pe	A B A B 0.0 0.0 0.4 0.4 0.1 0.1 0.3 0.3 0.1 0.1 0.4 0.4 0.1 0.1 0.4 0.4 0.1 0.1 0.3 0.3 0.1 0.1 0.4 0.4 0.1 0.1 0.3 0.3 0.1 0.1 0.4 0.4 0.1 0.1 0.3 0.3 0.1 0.0 0.3 0.3 0.1 0.1 0.4 0.4 0.0 0.1 0.4 0.4 0.0 0.1 0.4 0.4 0.0 0.1 0.4 0.4 0.0 0.1 0.4 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.4 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4 0.0 0.1 0.3 0.4	A B A B A 0.0 0.0 0.4 0.4 2.3 0.1 0.1 0.3 0.3 1.7 0.1 0.1 0.3 0.3 1.8 0.1 0.1 0.4 0.4 2.1 0.1 0.1 0.4 0.4 2.1 0.1 0.1 0.3 0.3 1.9 0.1 0.0 0.3 0.3 2.0 0.1 0.1 0.3 0.4 2.0 0.0 0.1 0.4 0.4 2.1 Per Cent Retained 0.1 0.1 0.3 0.4 2.0 ve Average Per Cent Ret	A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 0.1 0.1 0.3 0.3 1.7 1.8 0.1 0.1 0.3 0.3 1.8 2.0 0.1 0.1 0.4 0.4 2.1 2.0 0.1 0.1 0.4 0.4 2.1 2.1 0.1 0.1 0.3 0.3 1.9 2.1 0.1 0.0 0.3 0.3 2.0 2.1 0.1 0.1 0.3 0.4 2.0 2.1 0.0 0.1 0.4 0.4 2.1 2.1 0.0 0.1 0.4 0.4 2.1 2.1 0.0 0.1 0.4 0.4 2.1 2.1 0.0 0.1 0.3 0.4 2.0 2.1 0.0 0.1 0.3 0.4 2.0 2.1 0.0 </td <td>A B A 9.6 0.1 0.1 0.1 0.3 0.3 1.8 2.0 8.4 7.6 8.4 9.0</td> <td>A B A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.0 0.1 0.1 0.3 0.3 2.0 2.1 8.9 9.3 0.1 0.1 0.3 0.3 2.0 2.1 8.9 9.3 0.1 0.1 0.3 0.4 2.0 2.1 9.0 9.2 0.0 0.1 0.4 0.4 2.1 2.1 9.0 9.8</td> <td>20 30 40 50 A B A B A A B A B A B A B A B A B A B</td> <td>20 30 40 50 70 A B A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 35.8 32.8 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 33.0 31.5 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 33.8 33.3 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 34.5 32.2 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 34.2 33.3 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.9 34.2 33.3 0.1 0.1 0.3 0.3 2.0 2.1 8.7 9.9 34.3 31.3 0.1 0.1 0.3 0.3 2.0 2.1</td> <td>20 30 40 50 70 1 A B A B A B A B A B A B A B A 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 35.8 32.8 33.0 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 33.0 31.5 34.7 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 33.8 33.3 34.2 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 34.5 32.2 33.8 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 34.2 33.3 34.1 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.0 34.3 31.3 34.1 0.1 0.0 0.3 0.3 2.0 2.1 8.9 9.3 35.3 32.0 33.8 0.1 0.1 0.1 0.3 0.4 2.0 2.1 9.0 9.2 34.9 31.0 33.6 0.0 0.1 0.4 0.4 2.1 2.1 9.0 9.8 35.4 32.2 33.4 Per Cent Retained 0.1 0.1 0.3 0.4 2.0 2.1 8.8 9.4 34.6 32.2 33.9 The Average Per Cent Retained</td> <td>20 30 40 50 70 100 A B A B A B A B A B A B A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 35.8 32.8 33.0 33.7 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 33.0 31.5 34.7 34.4 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 33.8 33.3 34.2 33.6 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 34.5 32.2 33.8 34.4 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 34.2 33.3 34.1 34.4 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.0 34.3 31.3 34.1 34.7 0.1 0.0 0.3 0.3 2.0 2.1 8.9 9.3 35.3 32.0 33.8 34.5 0.1 0.1 0.1 0.3 0.4 2.0 2.1 9.0 9.2 34.9 31.0 33.6 34.7 0.0 0.1 0.4 0.4 2.1 2.1 9.0 9.8 35.4 32.2 33.4 34.2 Per Cent Retained 0.1 0.1 0.3 0.4 2.0 2.1 8.8 9.4 34.6 32.2 33.9 34.1 Towe Average Per Cent Retained</td> <td>A B A</td> <td>20 30 40 50 70 100 140 A B A B A B A B A B A B A B A B A B A</td> <td>20 30 40 50 70 100 140 20 A B A B A B A B A B A B A B A B A B A</td> <td>20 30 40 50 70 100 140 200 A B A B A B A B A B A B A B A B A B A</td> <td>20 30 40 50 70 100 140 200 2 A B A B A B A B A B A B A B A B A B A</td> <td>20 30 40 50 70 100 140 200 270 A B A B A B A B A B A B A B A B A B A</td> <td>20 30 40 50 70 100 140 200 270 E A B A B A B A B A B A B A B A B A B A</td> <td>20 30 40 50 70 100 140 200 270 Pan A B A B A B A B A B A B A B A B A B A</td> <td>The contract of the state of th</td>	A B A 9.6 0.1 0.1 0.1 0.3 0.3 1.8 2.0 8.4 7.6 8.4 9.0	A B A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.0 0.1 0.1 0.3 0.3 2.0 2.1 8.9 9.3 0.1 0.1 0.3 0.3 2.0 2.1 8.9 9.3 0.1 0.1 0.3 0.4 2.0 2.1 9.0 9.2 0.0 0.1 0.4 0.4 2.1 2.1 9.0 9.8	20 30 40 50 A B A B A A B A B A B A B A B A B A B	20 30 40 50 70 A B A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 35.8 32.8 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 33.0 31.5 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 33.8 33.3 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 34.5 32.2 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 34.2 33.3 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.9 34.2 33.3 0.1 0.1 0.3 0.3 2.0 2.1 8.7 9.9 34.3 31.3 0.1 0.1 0.3 0.3 2.0 2.1	20 30 40 50 70 1 A B A B A B A B A B A B A B A 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 35.8 32.8 33.0 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 33.0 31.5 34.7 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 33.8 33.3 34.2 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 34.5 32.2 33.8 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 34.2 33.3 34.1 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.0 34.3 31.3 34.1 0.1 0.0 0.3 0.3 2.0 2.1 8.9 9.3 35.3 32.0 33.8 0.1 0.1 0.1 0.3 0.4 2.0 2.1 9.0 9.2 34.9 31.0 33.6 0.0 0.1 0.4 0.4 2.1 2.1 9.0 9.8 35.4 32.2 33.4 Per Cent Retained 0.1 0.1 0.3 0.4 2.0 2.1 8.8 9.4 34.6 32.2 33.9 The Average Per Cent Retained	20 30 40 50 70 100 A B A B A B A B A B A B A B A B A B 0.0 0.0 0.4 0.4 2.3 2.4 9.6 10.3 35.8 32.8 33.0 33.7 0.1 0.1 0.3 0.3 1.7 1.8 7.6 8.6 33.0 31.5 34.7 34.4 0.1 0.1 0.3 0.3 1.8 2.0 8.4 9.4 33.8 33.3 34.2 33.6 0.1 0.1 0.4 0.4 2.1 2.0 8.9 9.3 34.5 32.2 33.8 34.4 0.1 0.1 0.4 0.4 2.1 2.1 8.7 9.9 34.2 33.3 34.1 34.4 0.1 0.1 0.3 0.3 1.9 2.1 8.7 9.0 34.3 31.3 34.1 34.7 0.1 0.0 0.3 0.3 2.0 2.1 8.9 9.3 35.3 32.0 33.8 34.5 0.1 0.1 0.1 0.3 0.4 2.0 2.1 9.0 9.2 34.9 31.0 33.6 34.7 0.0 0.1 0.4 0.4 2.1 2.1 9.0 9.8 35.4 32.2 33.4 34.2 Per Cent Retained 0.1 0.1 0.3 0.4 2.0 2.1 8.8 9.4 34.6 32.2 33.9 34.1 Towe Average Per Cent Retained	A B A	20 30 40 50 70 100 140 A B A B A B A B A B A B A B A B A B A	20 30 40 50 70 100 140 20 A B A B A B A B A B A B A B A B A B A	20 30 40 50 70 100 140 200 A B A B A B A B A B A B A B A B A B A	20 30 40 50 70 100 140 200 2 A B A B A B A B A B A B A B A B A B A	20 30 40 50 70 100 140 200 270 A B A B A B A B A B A B A B A B A B A	20 30 40 50 70 100 140 200 270 E A B A B A B A B A B A B A B A B A B A	20 30 40 50 70 100 140 200 270 Pan A B A B A B A B A B A B A B A B A B A	The contract of the state of th

present a comparison of the relative abilities of the two sieve shakers to separate a sample into its component sieve fractions. The comparison is relative, not absolute, because the two pieces of apparatus were checked against each other, and not against a fully reliable standard.

The data in Table 3 reveals that the average values of the cumulative percentages retained on different sieves are nearly the same for both types of sieve shakers (Fig. 4). If average percentages retained are considered (Fig. 5), it is apparent that a somewhat higher percentage was retained on the No. 70 sieve in those analyses made with the type (A) sieve shaker than in the analyses made with the type (B) sifter. On the No. 140 sieve the opposite effect occurred, while on all other sieves the differences were slight.

Test Apparatus Variables

One would probably suggest that the differences between the two analyses of each sample were in part due to the effect of prolonged sieving because the first 15-min sieving on each sample was made with a type (A) sieve shaker and the second 15-min sieving was made with a type (B) sifter. However, a comparison between the results given in Tables 3 and 2 reveals that most of the difference between the analyses made with two different sieve shakers was caused by the apparatus variables and not the sieving time effect.

For example, Table 3 reveals that the average difference in percentage retained on the No. 70 sieve between the analysis made with the type (B) sifter and that with the type (A) sieve shaker was 2.4 per cent. Table 2 shows that the maximum difference between consecutive analyses on one sample with a type (A) sieve shaker was less than 1.0 per cent for the same sieve fraction.

The data in Table 3 indicate that there are some differences in results between the two sieve shakers when percentages retained are considered as is shown in Fig. 5, but these differences nearly cancel each other if data are plotted as cumulative curves (Fig. 4). The average A.F.S. fineness numbers for the nine samples were found to be 70.4 for the rotary (type A) shaker, and 70.6 for the gyratory (type B) shaker.

Conclusions

Under the conditions of the analyses made in this study the vibratory and rotary type sieve shakers will yield approximately the same results on a given sample with the same set of sieves. Of nine samples analyzed with both types of apparatus, the average A.F.S. fineness number for the analyses made with the rotary sieve shaker was 70.4, while the average for the analyses made with the vibratory sieve shaker was 70.6.

The maximum sampling error which can be expected for the sampling technique used and for the conditions of the experiments is about \pm 3.4 per cent in terms of A.F.S. fineness numbers.

Acknowledgment

Grateful acknowledgment is made to Joseph M. Cameron, National Bureau of Standards, and Dr. Frank G. Carpenter, Research Associate, National Bureau of Standards, for their helpful suggestions.

References

- 1. F. A. Shergold, "The Effect of Sieve Loading on the Results of Sieve Analyses of Natural Sands," *Journal*, Society of Chemical Industry, vol. 65 (1946), pp. 245-249.
- Industry, vol. 65 (1946), pp. 245-249.
 2. Sture Mörtsell, "On Accuracy of Sieve Analyses Made by Means of Sieving Machines," Kungl. Tekniska Höglans Handlingar, Transactions, Royal Institute of Technology, Stockholm, no. 17 (1948).
- 3. A. W. Fahrenwald and S. W. Stockdale, "Effect of Sieve Motion on Screening Efficiency," U. S. Bureau of Mines Technical Report 2933 (1929).
- 4. A. S. Swinford and Francis Swinford, "A Comparison of Three Sieve Shakers," *Journal of Sedimentary Petrology*, vol. 13, pp. 3-13 (1946).
- 5. L. E. Simon, An Engineers Manual of Statistical Methods, John Wiley & Sons, New York (1941).
- 6. G. W. Snedecor, Statistical Methods, fourth edition (1946) chapter 6, Iowa State College Press, Ames, Iowa.

OXIDATION - REDUCTION PRINCIPLES CONTROLLING THE COMPOSITION OF MOLTEN CAST IRONS

By R. W. Heine *

PART I

Control of Chemical composition of cast irons during melting operations has been considered mainly an oxidation problem. Silicon and manganese react with oxygen to form oxides that enter the slag. Carbon may react with oxygen and leave the iron as a gas, or carbon may be dissolved in the iron from carbonaceous materials such as coke. Chemical composition control within limits has been obtained by the use of arbitrary correction factors, devised by experienced foundrymen. However, the simplified viewpoint of oxidation as the only important chemical process predominant in the melting of cast irons leaves unexplained certain exceptions to the rule commonly experienced in the foundry.

Silicon pickup, a reversal of the oxidation process, is a well-known phenomenon often reported. Silica reduction rather than silicon oxidation is most frequently encountered in malleable and steel foundries. Since silica and other oxides can be reduced by carbon, it appears that reduction as well as oxidation is a factor in the composition changes which may occur in the melting of ferrous alloys. The principles involved in chemical composition changes related to oxidation or reduction reactions in molten cast irons are considered in this paper.

Theoretical Basis

The principle oxidation reactions in melting may be written as follows:

1.
$$\underline{C} + \underline{O} = CO \text{ (gas)}$$

2. $\underline{Si} + \underline{2} \underline{O} = SiO_2 \text{ (solid)}$
3. $\overline{Mn} + \overline{O} = MnO \text{ (liquid)}$

When underlined, the symbol denotes the particular element dissolved in iron. The above reactions are well known.

Some possible reduction reactions are the following:

4.
$$SiO_2$$
 (s) $+ 2 C = Si + 2 CO$ (g)
5. MnO (l) $+ C = Mn + CO$ (g)

6.
$$Al_2O_3$$
 (s) + $3\overline{C} = \overline{2}Al + 3CO$ (g)

7.
$$TiO_2$$
 (s) $+ 2\overline{C} = Ti + 2CO$ (g)

Reduction as in Equations 4 through 7 requires contact of the molten iron with an oxide-rich material to permit the reactions to go to the right to any appreciable extent. Because of the siliceous nature of many slags and refractories in cast iron melting, Reaction 4 is the one of prime importance. Accordingly, the phenomenon of silica reduction, Equation 4, will be considered first as an example.

Oxidation Reactions 1, 2, and 3 are usually considered to progress more readily as temperature increases. However, it is also true that reduction, as in Equations 4 through 7, begins and progresses more readily as temperature increases. Actually both oxidation and reduction may occur simultaneously. Thus the predominating reactions causing composition changes are affected by temperature especially, and by concentration. As a starting point, the temperatures and concentrations which favor Reactions 1, 2, and 3 or Reactions 4 and 5 may be calculated. Rehder¹ has described some of the factors involved in Equation 4, silica reduction, applied to air furnace malleable iron melting. These are reviewed and some additional principles discussed in the following paragraphs.

The conditions controlling the reaction: SiO_2 (s) + 2C = Si + 2 CO (g) may be analyzed with the help of thermodynamic data.** The equations and energy changes necessary to determine the possibility of this reaction are tabulated below:

Equation
$$\Delta F^{\circ} = H - \Delta S^{\dagger} T^{\bullet}$$

1. $C + O = CO$ (g) $\Delta F^{\circ} = -8510$ $-7.52T$
2. $\overline{Si} + \overline{2O} = SiO_2$ (s) $\Delta F^{\circ} = -119180 + 43.7T$

When Equation 2 is subtracted from twice Equation 1, and the resulting equation is rearranged the following expression is obtained:

115

. 3

. 3

^{*} Asst. Professor of Metallurgical Engineering, Dept. of Mining and Metallurgy, University of Wisconsin, Madison, Wis.

^{**} The data assembled by J. Chipman are used. "Chemistry at 1600," ASM, Transactions, vol. XXX (1942).

^{*} Values in calories per mol, temperatures in degrees K, data at standard state of 1600 C, 1% C, 1% Si in iron.

4.
$$SiO_2$$
 (s) $+ 2C = Si + 2 CO$ (g) $\triangle F^{\circ} = 102160 -58.74T$

The symbol $\triangle F^{\circ}$ indicates the free energy change occurring with this reaction at a standard state of concentration and temperature usually employed in thermodynamic calculations. At 1600 C (2912 F, 1873 K), $\triangle F = -7860$. The thermodynamic principles affirm that with a negative $\triangle F$ value, reactions should progress to the right spontaneously at 1600 C under the proper conditions of concentration, i.e., the standard state of 1 per cent carbon and 1 per cent silicon in the iron in contact with solid silica and one atmosphere pressure of CO. Under these standard conditions an increase of percentage silicon and a decrease in per cent carbon in the melt is thus anticipated with the reaction being accompanied by a CO evolution.

The equilibrium temperature for Equation 4 is important since the reaction will progress neither to right or left at that temperature. Equilibrium obtains when $\triangle F_T = 0$. Assuming the effect of temperature on $\triangle H$ and $\triangle S$ to be small, then for Equation 4 when $\triangle F_T = 0$,

$$102160 - 58.74 \text{ T} = 0$$

 $T = 1739^{\circ} \text{ K, or}$
 $T = 1466^{\circ} \text{ C, or}$
 $T = 2671^{\circ} \text{ F.}$

Thus at 2671° F the reaction should be in equilibrium when standard concentration of 1 per cent carbon and 1 per cent silicon are present. This temperature was found to agree well with experimental evidence obtained by E. A. Lange⁷ under the author's direction in the University of Wisconsin Metallurgical laboratories.

Under laboratory melting conditions, the effect of molten metal temperatures on the oxidation of carbon and silicon, or the reduction of silica was studied. For this purpose, 8-lb heats of gray cast iron scrap were melted in an induction furnace having a silica crucible. The molten metal was oxidized by holding it in contact with air for periods up to 2 hr. Composition changes under these conditions were studied at a number of different temperatures. The results obtained are shown in Fig. 1. These curves well illustrate the temperature dependence of silicon and carbon oxidation. At 2680 F, no oxidation of silicon occurred over a period of 1 hr. Below that temperature some oxidation occurred while above that temperature silicon pickup, or reduction of silica from the crucible occurred.

Thus it is seen that the calculated equilibrium temperature for Equation 4 even with its limiting conditions seems to apply closely to that obtained with a commercial gray iron melted in air. The limited amount of oxidation of silicon and carbon shown in Fig. 1 at a temperature of 2375 F could be due to a skin of solid oxides which persists over the melt and acts as a protective cover preventing further oxidation. When the skin is skimmed it immediately reforms. This skin dissolves on heating at about 2550 F and above that temperature the melt surface is clear. Once the surface of the melt is clear, oxidation of carbon may begin at the surface and carbon losses

as shown in Fig. 1 occur. At temperatures of 2600 to 2700 F close to the equilibrium for Reaction 4, carbon oxidation may proceed independently of silicon and in fact reduction of silica at higher temperatures results in the melt increasing in silicon simultaneously with the oxidation of carbon.

F

th

ta

Va

no

va

pe

ph

cei

ass

tic

use

the

din

wh

the

tha

rea

cer

per

sili

sili

eni

en

mi

ins

the

the

Since Fig. 1 shows that carbon may be oxidized to well below 2 per cent and yet be accompanied by an increase in silicon at temperatures where silica reduction will occur, it is evident that the concentration of the carbon in the iron has a vital bearing on the extent of the reaction: SiO_2 (s) + 2 C = Si + 2 CO (g). High carbon contents would be expected to favor reduction while low carbon content might permit the reaction to go from right to left or at least cause no increase in percentage silicon. The influence of various concentrations of carbon and silicon on the direction of Reaction 4 may be indicated by means of the equilibrium constant. Thermodynamic data were employed to calculate the equilibrium constant. The constant was obtained as follows:

8.
$$K = \frac{\text{Si} \times (\text{CO})^2}{\text{SiO}_2 \text{ (s)} \times \underline{C}^2}$$

$$\ln K = \frac{-\triangle F^{\circ}}{RT}$$

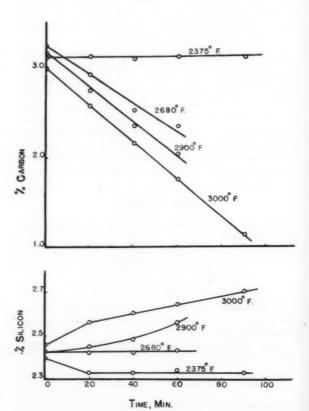


Fig. 1—Curves showing changes of carbon and silicon percentages in a molten cast iron held in a silica crucible in air and at the temperatures indicated. Induction melting was employed.

Where R is the gas constant, T is the absolute temperature and ΔF° is the standard free energy change of the reaction.

At 1600 C,

$$\log K = \frac{- (-7860)}{2.3 \times 1.987 \times 1873}$$

$$K = 8.29$$

For the purpose of this investigation it was assumed that the effect of temperature changes on $\triangle H$ is small and the K value was calculated for other temperatures by using the expression following:

9.
$$\log K_{\tau_2} = \frac{\triangle H}{2.3 \text{ R}} \times \frac{(T_2 - T_1)}{T_2 \times T_1}$$

Equilibrium constants at different temperatures are tabulated below:

Temp	perature	
C	F	K value (Equation 4)
1350	2462	0.121
1400	2552	0.31
1466	2671	1.0
1500	2732	1.76
1600	2912	8.29
1700	3092	33.4
1700	3032	33.1

Values of less than 1 indicate the Reaction 4 does not progress spontaneously to the right and these values occur below 1466 C (2671 F). Substituting percentages in the equilibrium constant, Equation 8, it may be simplified to the following:

10.
$$K = \frac{\% \underline{Si}}{\% \underline{C^2}}$$
; or $\% \underline{Si} = K \times \% \underline{C^2}$.

This relationship is obtained when it is assumed that the melt is in contact with silica and a CO atmosphere. The SiO₂ and CO then both become 100 per cent or 1 and are cancelled from Expression 8. These assumptions are not often justified and other corrections then must be made before Expression 8 may be used. The use of a concentration of 1 for CO is thought to be generally applicable for indicating the direction of this reaction because when Reaction 4 progresses to the right the melt boils and is in direct contact with CO bubbles as well as a CO-rich blanket which forms immediately over the surface. Actually the escape of the CO bubbles and the presence of less than one atmosphere CO over the melt means the reaction is progressing to the right. The use of a concentration of 1 for the SiO2 may be justified in experimental work where the melt is in contact with a silica crucible. However, in melting conditions when silicious slags containing a fraction of SiO2 are present, effective concentrations of less than I should be entered into Expression 8. Additional correction might be made in 10 by substituting chemical activity instead of weight per cent of silicon and carbon. For the purpose of this paper the simplified viewpoint of Expression 10 will be used since it seems to agree with the nature of the data.

With the constants tabulated above and the expres-

sion $Si = K \times C^2$, the percentages of silicon and carbon in equilibrium at various temperatures may be calculated and presented graphically. The equilibrium curves are shown plotted in Fig. 2. Since these

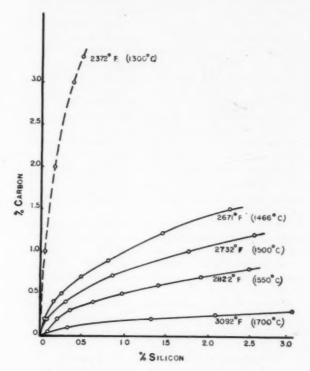


Fig. 2—Graph showing calculated equilibrium concentrations of per cent carbon and silicon for the reaction SiO₂ (s) + 2C = Si + 2CO (g) in molten ironcarbon-silicon alloys contained in a silica crucible under one atmosphere pressure of CO. Solid equilibrium curves at various temperatures indicate the reaction would progress spontaneously if an excess of carbon were present.

curves present only equilibrium relationships calculated from thermodynamic data, their interpretation must be hedged, with a view of the assumptions and limitations implied. These are briefly, that the metal is a molten alloy containing only silicon, carbon and iron, that the melt is in contact with solid silica and one atmosphere of CO. In spite of the aforementioned list of qualifying conditions, the principles involved in Equation 4 and Fig. 2 appear to have considerable application in laboratory and foundry melting of cast irons. Figure 2 may be interpreted with the view toward anticipating the direction of carbon and silicon composition changes in molten cast irons. A composition which lies to the left of the curves at temperatures above 2671 F (1466 C) would be expected to increase in percentage silicon and decrease in percentage carbon under the melting conditions previously specified.

In most cast irons more than 2 per cent carbon is present so that there is always an excess of reagents for the reduction of silica even though a simultaneous oxidation of carbon may occur. This concurrent oxidation of carbon and reduction of silica is shown by the data of Fig. 1. While silicon pickup in normal cast irons might be expected as low as 2671 F the rate of increase would be very low. It should be recognized that Fig. 2 gives no information as to the rates of the reaction. Judging from Lange's data, the rate of silica reduction at 2680 F is so slow as to cause no change in analysis within the time limits investigated. However, when the concentration of carbon is increased above that present in the cast iron, for example by the presence of coke or graphite in addition, silica may be reduced even at the lowest temperature calculated for spontaneous reaction.

When the composition of an iron lies to the right of the curves of Fig. 2, it might be expected that silicon would be oxidized, since there is an excess of silicon over that which would be protected by carbon by Reaction 4, SiO_2 (s) +2 C = Si + 2 CO (g) This effect is partially indicated in Fig. 1 by the curve

at 2375 F.

If the previous viewpoints are valid, it appears that the oxidation-reduction principles involved in Reaction 4 are extremely important in affecting composition changes in cast iron melting. The author was particularly interested in the application of such principles to commercial cast iron of composition ranges encountered in malleable iron melting. Additional experimental work was directed at these compositions.

Experimental Details

Eight-pound heats of white cast iron sprue were melted in an induction furnace lined with a silica crucible. The sprue was commercial white cast iron obtained from a single air furnace heat, having the average analysis of 2.24 C, 1.12 Si, 0.29 Mn, 0.14 P, 0.09 S, (and no chromium). Temperatures were measured with a Pt-Pt 10% Rh thermocouple. The temperature of the melt was raised to a preselected value and it was then exposed to an oxidizing condition consisting either of bubbling air through the melt or an air blast over the melt. At lower temperatures, it was necessary to bubble air through the melt because

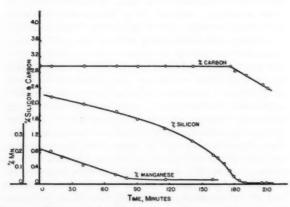


Fig. 3—Composition history of molten cast iron held at 2372 F in a silica crucible. Air at the rate of 0.15 cfm was bubbled through the metal from a silica tube held ½ in. below the surface.

of the oxide film which persists over the melt. This film appeared to prevent the oxidation of carbon and drastically inhibited the oxidation of silicon in the work of Fig. 1.

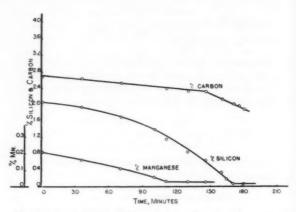


Fig. 4—Composition history of molten cast iron held at 2597 F in a silica crucible. Air at the rate of 0.15 cfm was bubbled through the metal from a silica tube held ½ in. below the surface.

An air blast of 0.15 cfm over the clear melt surface was employed at the higher temperatures to produce highly oxidizing conditions and to simulate the atmosphere movement occurring in air furnace melting. The air blast was directed tangentially downward around the refractory in order to obtain preheating prior to its hitting the melt. During the course of each heat made under the above conditions, samples for chemical analysis were withdrawn from the melt at regular intervals. The samples were obtained by drawing metal into pyrex glass tubing followed by quenching into water. The metal pins resulting from this procedure were pulverized for analysis. The composition of the melt as a function of time at temperature under the oxidizing conditions previously described is recorded in Fig. 3, 4, 5, and 6.

Oxidation of Silicon and Carbon

11

o

el

CE

vi

in

ti

ir

m

th

be

ar

Si

m

bu

A study of Fig. 3 to 6 reveals the tremendous influence of temperature and concentration on the oxidation—reduction reactions prevailing in molten cast irons. Considering first Fig. 3, some observations which do not appear on the graph must be mentioned. Initially when air is bubbling through the metal at 2372 F (1300 C), the melt behaves as though an inert gas were being employed. No sparking occurs but a dry slag gradually accumulates which must be removed periodically to prevent crusting over of the surface. This slag is mainly the result of oxidation of the silicon iron and a small amount of manganese as revealed by the analysis curves in Fig. 3. The slags will be considered separately in more detail later in the paper.

As the silicon content of the melt at 2372 F diminishes below about 0.8 per cent silicon some sparks are occasionally noted. When the silicon drops still further a percentage is reached where strong eruption of sparks suddenly occurs. This violent sparking is

d

caused by an actual carbon boil and is marked by an abrupt dropping of the percentage carbon in the melt. The marked change in slope of the carbon curve in Fig. 3 agrees with visual observation of the time of

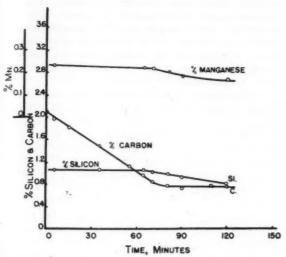


Fig. 5—Composition history of molten cast iron held at 2732 F in a silica crucible. Air at the rate of 0.15 cfm was directed over the surface of the molten metal.

beginning of the boil. Coincident with the beginning of the boil, a significant change occurs in the slag accumulation. The slag changes quickly from a dry, flaky material to a very fluid, thin, black slag. At the same time the silicon percentage in the melt drops quickly to a very low value.

The heat described above was one where the carbon and silicon percentage was raised by additions after melt down. Carbon was raised to that of a typical malleable iron air furnace charge. The silicon was raised to accentuate the effects of silicon concentration. To check the sequence of the aforementioned observations and to determine possible concentration effects, two other combinations of carbon and silicon were studied; one with 3.02 per cent C and 1.02 per cent Si and another with 3.38 per cent C and 2.53 per cent Si. The chemical composition, slag and carbon boil changes proceeded in the same manner as previously discussed. Since the three heats did behave in the same way, Fig. 3 will be considered as illustrative of silicon and carbon oxidation in molten cast iron at 2372 F under the conditions of these experiments.

The protective action of silicon in preventing carbon oxidation is well illustrated in Fig. 3. In each of the heats made at 2372 F, there was no detectable loss in carbon content until the percentage of silicon had been oxidized down to about 0.35 to 0.45 per cent and a carbon boil began. The mechanism whereby silicon may protect carbon is not clear. The persistence of the dry slag and absence of any CO bubbles formation indicates silicon reacts with oxygen more strongly than carbon. It was thought that air bubbles passing through the melt would overcome

the barrier effect previously mentioned, but even this did not cause carbon oxidation. It may be that the bubbles become immediately coated with the slag. However that may be, thermodynamic data offers some explanation. Consider the following information:

(1)
$$\underline{C} + \underline{O} = CO$$
 (g) $\triangle F^{\circ} = -8510 - 7.52 \text{ T}$
(2) $\overline{Si} + \overline{2} O = SiO_2$ (s) $\triangle F^{\circ} = -119180 + 43.7 \text{ T}$

A reaction will occur more positively as $-\Delta F$ becomes greater. It is seen from (1) and (2) above that $-\Delta F$ for carbon oxidation increases with increasing temperature but $-\Delta F$ for silicon oxidation increases with decreasing temperature. Thus according to this thermodynamic information the lower the temperature the greater tendency for silicon to be oxidized and the less the tendency of carbon to be oxidized. Data obtained at 2597 F further clarifies this effect of temperature.

The experimental procedure described for the tests at 2372 F was repeated at 2597 F. The results are plotted in Fig. 4. From Fig. 4, it is immediately evident that oxidation of carbon is no longer fully inhibited by silicon at the higher temperature. The slow drop in carbon content during the first part of the heat was visually evident as a gentle but regular sparking. As silicon was lost from the melt a concentration was reached, about 0.62 per cent, where oxidation of carbon was no longer inhibited and a rapid boil resulting in violent sparking occurred as was the case at 2372 F. The beginning of the boil coincides with the sudden change in slope of the carbon oxidation curve of Fig. 4. Comparison of Fig. 3 and 4 confirms the principle that the tendency of carbon oxidation increases with temperature while the tendency of silicon oxidation decreases with temperature. At 2372 F, silicon was effective down to 0.35 to 0.45 per cent in combining with oxygen and thus protecting carbon whereas at 2597 F a greater amount of silicon, 0.6 to 0.65 per cent, was necessary to combine with oxygen and inhibit oxidation of carbon. Except for the effect of temperature on rate of oxidation the

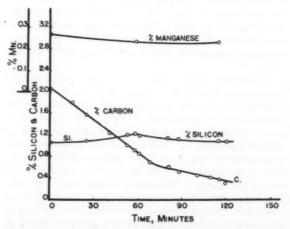


Fig. 6—Composition history of molten cast iron held at 2825 F in a silica crucible. Air at the rate of 0.15 cfm was directed over the surface of the molten metal.

sequence of events with respect to the boil and formation of a fluid slag was the same at both temperatures.

From the results of the heats at 2597 and 2372 F, a second possible mechanism whereby silicon might protect carbon from oxidation is suggested, namely by Reaction 4, SiO_2 (s) + 2 C = 2 CO (g) + Si. At temperatures below 2671 F where Reaction 4 should proceed to the left, silicon would prevent the formation of CO and thus the oxidation of carbon from the melt. This reaction is concentration dependent and thus when the silicon drops to a certain low value at a given temperature CO is no longer prevented from forming and a boil may occur. Such a mechanism would be analogous to the action of carbon preventing the formation of SiO_2 (oxidation of silicon) at 2680 F (Fig. 1) because of its ability to reduce SiO_2 .

While either of the two mechanisms of carbon protection discussed above may provide an adequate conception, it is probable that both may be involved under different circumstances as discussed later. Reaction 4 is desirable from a chemical sense since it involves three of the substances engaged in the reaction and points out their concentration and temperature dependence, i.e. the interdependence of the substances in the melt on each other is considered by Reaction 4 whereas Reactions 1 and 2 consider these substances as acting independently of each other. The interdependence of the reacting substances is further illustrated by heats made at higher temperatures.

Silica Reduction

At higher temperatures in induction melting of cast iron the surface of the melt becomes clear permitting oxidation to progress at the metal-atmosphere interface. Air was thus directed at the surface rather than as bubbles below the surface; the former being used since movement of atmosphere over the melt is more characteristic of the usual hearth-type melting process. The change in composition caused by the air movement and temperature combination at 2732 F, and 2825 F is plotted in Fig. 5 and 6 respectively. In both cases the importance of temperature and concentration is further emphasized and clarified. The composition of the melt at the beginning of the heat was that obtained by simply remelting the white iron scrap.

Except for a low carbon percentage it is seen that the composition is in a range common to malleable iron melting during the final melting period at temperatures above about 2650 F. From Fig. 5 and 6 it is seen that carbon in the melt is oxidized rapidly at 2732 F and 2825 F. This carbon loss is mainly a surface reaction but a gentle evolution of bubbles occurs from within the melt during the first 30 min of the heat. These bubbles seem to rise from the bottom and sides of the crucible. The reaction of carbon in the iron with the crucible results in pitting of the walls and bottom; the pitting being so severe at temperatures of 3000 F and above as to ruin the crucible in very few heats.

The aforementioned bubbling comes from the re-

action of carbon in the iron and the silica crucible and it positively shows that this iron and, in fact, most cast irons are nonequilibrium compositions at high temperatures. The bubbling action in the induction crucible is of course the same reaction in air furnace melting which causes the melt to assume a gentle boil during the finishing stages of a heat. This type of boil may also be noted when iron is held in silicalined ladles. In the case of these runs, when the carbon is oxidized to a low enough value, the bubbling eventually ceases as might be expected from equilibrium concentrations for Reaction 4.

Marked changes in the behavior of the melt of Fig. 5 were visually noted as in the case of the previous heats. The clear active surface of the melt prevails until the carbon drops down to a certain value. During the period when the melt surface is clear, no oxidation loss of silicon occurs as shown in Fig. 5 for a temperature of 2732 F. Shortly before silicon oxidation begins, particles of a dry scum begin to form and redissolve at the edges of the molten surface.

Abruptly at the time corresponding to the change to negative slope of the silicon curve, Fig. 5, a dry appearing scum forms over one-third to two-thirds of the surface area of the melt. This scum covers threequarters or more of the surface within 4 or 5 min of its abrupt appearance. The formation and growth of the scum results from oxidation of the silicon as revealed in the curves of Fig. 5. It is also seen that virtually no oxidation of manganese occurs until the scum develops. Simultaneous with the covering over of the melt by the scum, the rate of carbon oxidation decreases drastically. This is evidently the same protective mechanism as that which occurs at lower temperatures and discussed in relation to the heats at 2372 F and 2597 F. The heats were continued only a short time after the surface was scummed over since the composition was already well below carbon contents of interest in malleable iron melting.

A temperature of 2825 F produced the same melting phenomen as described above for 2732 F but with one added feature. Since the temperature was high enough for appreciable silica reduction, an increase of percentage silicon was noted during the early part of the heat as indicated in Fig. 6. Again in this heat, the point of inflection in the silicon curve corresponds to the formation of scum. At this higher temperature however, the scum took longer to cover the surface. As a consequence the carbon oxidation curve changes slope later and more gradually than was true at 2732 F.

The data of Fig. 5 and 6 further confirm the important influence of concentration and temperature in the oxidation-reduction reactions which cause composition changes in cast irons. The effect of temperature has been adequately demonstrated. In the proper temperature range the phenomenon of silica reduction would be expected to proceed in practically all molten cast irons in contact with highly siliceous refractories and normal atmospheres. The previous statement is made because all practical cast irons apparently contain more than enough carbon to begin and continue the reduction of silicon-dioxide once

le

h

ın

ce

le

)E

a.

g

ls

of

9

normal melting and pouring temperatures are reached. The controlling influence of carbon concentration is well substantiated in Fig. 5 and 6. Carbon acts as a protector of silicon, preventing its oxidation until the carbon concentration is lowered to one no longer effective at the particular temperature and silicon level involved. This protective power of carbon depends on its ability to reduce silica, i.e. combine with oxygen more strongly than does silicon as temperature increases according to the reaction

$$SiO_2$$
 (s) $+ 2C = 2CO$ (g) $+ Si$

The preceding equation shows one way that carbon is lost from the melt. Carbon lost by silica reduction can be accounted for by a chemically equivalent percentage of silicon pickup. Carbon lost in this way is however, only a minor part of the loss shown in Fig. 6. Direct oxidation of carbon in the iron causes the major portion of loss. Carbon losses by the two mechanisms may be readily determined by melting the iron in a CO atmosphere. Lange7 has shown that chemically equivalent amounts of silicon are gained and carbon lost by reduction of silica when the direct oxidation process is prevented by melting under a CO atmosphere. Therefore, the percentage carbon curves shown in Fig. 6 incorporate two distinctly different mechanisms of carbon loss, direct oxidation accounting for the major percentage and indirect oxidation through silica reduction accounting for an amount chemically equivalent to the per cent silicon increase.

The experimental results verify the idea that carbon may be oxidized from the melt concurrently with the reduction of silica and its accompanying increase in silicon percentage as long as a high enough carbon concentration exists. The carbon level where the reduction of silica ceases and the oxidation of silicon begins was calculated and is defined by the equilibrium composition curves of Fig. 2. These curves are limited however to the conditions assumed for the calculations. Because of the nonequilibrium nature of these particular experiments, the present data could not be expected to follow exactly the curves of Fig. 2. It is extremely significant to note from Fig. 6 that carbon, even at concentrations as low as 0.30 per cent will protect silicon from significant oxidation at much higher percentages than expected from Fig. 2.

This unexpectedly potent protective ability of carbon may be due in part to the experimental method whereby one of the products of Reaction 4, CO, was continuously removed from the reaction thus assisting in driving Reaction 4 to the right. With respect to normal cast iron carbon contents, the composition always favors silica reduction. It may be concluded that under the atmospheric conditions employed in this work, carbon is an extremely effective protective agent for inhibiting reaction between silicon in cast iron and oxygen in gas atmospheres. The mechanism of such a protective action lies in the influence of temperature and concentration on the fundamental oxidation-reduction Reaction 4,

$$SiO_2$$
 (s) $+ 2C = \underline{Si} + 2CO$ (g).

Manganese

The change in manganese percentage in the molten irons provides further clarification of the role of oxidation-reduction reactions in cast iron melting. The reduction Reaction 5, MnO (1) + \underline{C} = CO (g) + Mn is involved in this case. This reaction is temperature and concentration dependent in the same manner as silica reduction. There is, however, one significant difference in the experimental conditions. Since there is initally no MnO in the crucible or slag except that which develops by oxidation of manganese, the reaction (5) is prevented from proceeding to the right unless some MnO is added to the surface of the melt.

In these experiments, composition changes without addition of MnO from an external source were investigated. Manganese percentage changes for the previously discussed heats are plotted in Fig. 3, 4, 5, and 6. From these data it is seen that manganese is oxidized at a decreasing rate as temperature increases. This is identical with the behavior of silicon. The free energy data again indicate that this should occur even if only the simple oxidation reaction obtains:

3.
$$Mn + O = MnO$$
 (1), $\triangle F^{\circ} = -59100 + 26.97T$.

As temperature increases, the tendency of this reaction to go to the right decreases. Oxidation of manganese occurs even more readily than silicon oxidation at 2372 F since Fig. 3 and 4 show it to be the first to be almost completely removed.

From Fig. 5 for the heat at 2732 F, it may be seen that virtually no manganese is lost by oxidation until the carbon content of the molten iron drops to a low level. Evidently the same protective action of carbon applies to manganese as was discussed for silicon. These two elements are seen from Fig. 5 to both be oxidized only when the carbon has diminished to the percentage where scum formation begins and continues. As in the case of silicon the higher temperature of 2825 F more positively revealed the ability of carbon to prevent oxidation.

Figure 6 shows that only about 0.03 per cent manganese was lost over the entire heat even though the percentage carbon dropped to 0.30 per cent and silicon was oxidized 0.14 per cent below its maximum value. Thus, in this heat manganese oxidation did not occur appreciably when the scum formed; probably because carbon is more effective in reducing MnO than SiO2. If carbon reduces MnO, it would be expected that there would be a manganese pickup in the melt providing a source of MnO were available in the slag or crucible. This possibility was tested experimentally. Manganese dioxide was added to the clear, slag-free surface of a molten iron containing 3.02 per cent C, 0.025 per cent Si, and 0.03 per cent Mn at a temperature of 2597 F. Only an amount of MnO₂ chemically sufficient to raise the percentage manganese to 0.30 per cent was added. The metal was then held at temperature 10 min and a sample withdrawn. The percentage manganese rose to 0.21 per cent thus showing rapid reduction of the oxide at 2597 F.

Reduction of manganese oxides by carbon evidently occurs at lower temperatures than those required by silicon dioxide and would normally be expected to occur in cast iron melting if a source of MnO were present. Since only dilute concentrations of MnO occur in most iron melting slags and refractories, no manganese pickup normally is experienced. In fact, the very low MnO content of iron melting slags usually permits a small amount of manganese to be oxidized from the melt even at high temperatures such as used for Fig. 6, in order that an equilibrium between slag and metal be approached.

The thermodynamic reasoning employed in the case of silicon may be extended to manganese. The reduction reaction data was obtained by the same methods and is given as follows:

5. MnO (1)+C=CO (g) +Mn
$$\triangle F^{\circ}$$
=50590 -34.49T

The equilibrium temperature for this reaction was calculated as 2178 F for standard concentrations. Thus reduction of the oxide would be expected in the experiment where oxide of manganese was added to the melt at 2597 F. Protection of manganese by carbon would even be expected in the heats at 2372 F and 2597 F, Fig. 3 and 4, were it not for the lack of MnO in contact with the metal. It is also likely that the calculated equilibrium temperature may be in error on the low side since the thermodynamic data for manganese oxidation are not of a high order of accuracy.

Equilibrium constants for Equation 5 were calculated and a few are given below:

Tempe	erature	
C	F	K value (Equations)
1300	2372	1.655
1400	2552	4.37
1500	2732	10.28
1600	2912	22.18

These are seen to be significantly higher than those for silicon, Equilibrium concentrations of carbon and manganese could be plotted for the expression

$$K = \frac{CO (g) \times Mn}{MnO (1) \times C}$$
 as was done for silicon in Fig.

2. The principle point to be made from such graphs is that normal cast irons are nonequilibrium compositions sufficiently rich in carbon at normal melting temperatures so that reduction of manganese or silicon oxides can be expected to occur if an oxide-rich source is provided.

The Slag Phase

In the previous discussion reference has been made to unique characteristics of the slag developed in each of the heats. Certain relationships exist between the molten metal and the nature of the slag which it generates. Low temperature effects will be considered first. Below 2550 F in normal atmosphere, all compositions of cast iron studied by the author persist in forming a dry slag cover. This slag can readily be demonstrated as dependent on the silicon percentage present in the iron. For instance, if an iron of 3.0

per cent C from which silicon and manganese have been removed by oxidation, is held in a silica-lined induction furnace at 2372 F in air, the surface of the melt will be clear.

A boil will occur only if air is blown at the surface. If the air blast is discontinued, the boil will gradually subside. The same situation is obtained if a synthetic iron of 3.0 per cent C and no silicon is used. However, in either case, if silicon as ferrosilicon is added to the clear surface of the melt, ultimately an addition is reached where the surface scums over. Such behavior is taken to mean that a sufficiently high silicon concentration, 0.3 to 0.4 per cent at 2372 F is reached so that the dry slag is generated. Concurrent with the formation of this slag it is no longer possible to produce a carbon boil by blowing air at the surface. In other words, the protective action of silicon on carbon at low temperatures begins once silicon percentages are present in excess of that necessary to produce the scum.

As long as the higher percentages of silicon are present, air bubbled through the melt or blown at the surface does not oxidize carbon; that is, within the rate of air supply employed in these tests. However, as silicon is oxidized from the melt and the silicon percentage is reached where carbon is no longer protected, the nature of the slag changes. Within about two minutes' time its dry, flaky character disappears and it becomes extremely fluid. A carbon boil accompanies this change in the slag but only so long as the air supply is maintained. The observations pointed out above were also noted at 2597 F but occurred at a higher silicon percentage, 0.6 to 0.7 per cent.

Samples of the fluid slag were analyzed for FeO and SiO₂. At 2372 F the fluid slag contained 68.4 per cent FeO and 31.2 per cent SiO₂. At 2597 F the fluid slag analyzed 65.58 per cent FeO and 33.98 per cent SiO₂. The fluid slag was studied by X-ray diffraction methods to determine whether any identifiable compounds were present. A strong diffraction pattern was obtained, having the interplanar distance measurements for fayalite FeO 'SiO₂, reported by Taylor and Savage.² A petrographic study of the crushed material also verified fayalite as the principle constituent of the fluid slags at 2372 and 2597 F.

In addition to the strong fayalite pattern, the X-ray tests showed weak indications of Fe₃O₄, magnetite, and γ-crystobalite. On the basis of the chemical composition and X-ray and petrographic identification, it was concluded that fayalite, 2 FeO 'SiO₂, was the principle constituent of the fluid slag formed at 2372 F and 2597 F. The dry slags were more difficult to analyze because of the difficulty of obtaining representative samples. The scum layer was so thin that removal of a sample was usually accompanied by entrapped beads of iron. Analyses of slag with the beads removed did reveal a lower percentage of iron oxide in this slag, with samples running between 50 and 60 per cent.

However, removal of the entrapped iron from the sample makes analytical values on the dry slag of questionable significance. X-ray studies of the mater11

ial showed it to be of different constitution than the fluid slag. Magnetite, Fe₃O₄, and crystobalite were found to be the principle substances present with some weak fayalite indications also being noted. A magnetic test revealed some of the crushed slag particles to be strongly magnetic, a characteristic of Fe₃O₄. The dry slag was thus concluded to be composed principally of magnetite, crystobalite, entrapped iron particles and a small amount of fayalite.

The dry slag which existed at lower temperatures was found to be similar to the type of slag formed at higher temperatures. Initially at 2732 F and 2825 F, the surface of the melt is clear and remains so until carbon is oxidized down to a level which permits slag formation. The beginning of formation of the slag at 2732 F and 2825 F corresponds with the first oxidation losses of silicon from the melt as revealed in Fig. 5 and 6. Evidently carbon is the element whose concentration regulates the existence or non-existence of the dry slag at elevated temperature, whereas silicon is the controlling element at the lower temperatures.

In the usual cast irons, no slag is formed by oxidation of silicon at elevated temperature because of the reducing action of carbon. Not until carbon is oxidized to percentages below those normal to cast irons would slag or oxidation products form. However, at low temperatures the strong affinity of silicon for oxygen guarantees that a dry slag will prevail. It was found that the dry slag persists at low temperatures even when the metal is held under a CO atmosphere. Only if the silicon content is oxidized below values normal for cast irons will the nature of the slag change from dry to fluid.

Chemical analyses of dry slags formed at 2732 F and 2825 F with total iron figured as FeO revealed 42.78 per cent SiO₂ and 56.50 per cent FeO in the

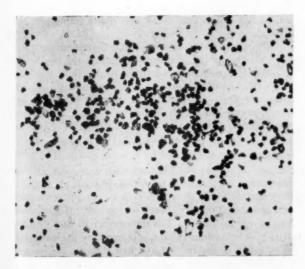


Fig. 7—Particles of crushed slag. Slag was generated by the molten iron at 2825 F after the beginning of silicon oxidation as shown in Fig. 6. Black particles are magnetic, Fe₃O₄; translucent and light-colored particles, crystobalite. Mag. 20×.

former and 50.98 per cent SiO₂ and 48.10 per cent FeO in the latter. These analyses again presented the problem of samples containing entrapped beads of iron. X-ray diffraction studies were also employed to identify constituents of these slags. Diffraction patterns on the 2732 F slag showed positively the presence of Fe₃O₄, crystobalite, and fayalite while the 2825 F slag produced a strong pattern for crystobalite and Fe₃O₄ with only faint indications of fayalite.

Coarse particles of crushed slag are shown in Fig. 7 for the 2825 heat. The translucent particles are crystobalite and appear to be almost free of iron while the dark particles are a mixture of magnetic Fe₃O₄ and silicates of iron. The dry slags are concluded to be mixtures of Fe₃O₄, small amounts of iron silicates, and crystobalites; the high temperature form of quartz increasing in percentage as temperature increases. The actual constituents of the slags are of course temperature dependent. It must be pointed out that the identifications discussed were performed on samples withdrawn from the surface of the melt and may thus incorporate some changes occurring when the material cools to room temperature.

The importance of temperature and metal composition in determining the nature of the slag generated by the molten metal cannot be over-emphasized. An excess of silicon beyond a certain value causes a dry slag to be generated under oxidizing conditions at low temperatures, whereas a lower value of silicon permits the formation of a very fluid slag consisting principally of fayalite. Under the oxidizing conditions of these experiments at temperatures above 2600 F, carbon prevents the formation of any slag until it has dropped to a level which permits the oxidation of silicon and results in the formation of a dry slag containing considerable silica as crystobalite.

The salient principles involved in the aforementioned phenomena are that the slag phases formed at any temperature are dependent on equilibrium compositions between metal and slag with changes in metal composition determining whether a dry slag, fluid slag, or no slag will be present. As a corollary it is true that for a fixed chemical composition heating and cooling normal molten cast irons will bring them through temperature ranges where a slag is generated by the oxidation of silicon, and temperature ranges where the reducing action of carbon prevents slag formation under atmospheric conditions. At elevated temperatures, possible equilibria are initially between a metal and a gas phase and these result in carbon losses under oxidizing conditions until the metal composition changes so that metal-slag equilibria become active and a slag is generated by silicon, iron and manganese losses. Equilibria thus may shift from gasmetal to slag-metal equilibria or vice-versa depending on the overall effects of composition and temperature.

Limitations

The data and principles discussed in this paper have been presented with certain limiting conditions. An attempt has been made to evaluate the normal course of reactions in molten cast irons. The change in chemical analysis of the iron in contact with a sili-

R

re

in

of

of

bo

ch

in

an

tic

TI

co

ces

co

rea

ot

W

on

be

ceous refractory and the generation of slags and atmospheres by the iron have been studied under the regulated circumstances of the experiments. The experimental results are nearly identical with those obtained in the foundry and reported by the author in another paper dealing with the air furnace melting

process. See Part II, p. 133.

However, attempts to apply these principles to other melting processes must be carefully considered because of fundamental differences involved. A major difference exists in the presence of slag over the melt at elevated temperatures. While it is not normal for cast iron to generate a slag at elevated temperatures, slags may be introduced and prevail as in the case of cupola melting. Reactions between the metal and extraneous slags, i.e. those not generated by the iron itself, were not considered in this paper and such reactions may present some differences in results. With respect to slag formation, cold melt air furnace cast iron melting is unique in that the major portion of the slag is skimmed early in the process and subsequent slag accumulation is supposed to be small.

A further experimental limitation intentionally adopted is associated with the chemistry and physics of rates of reactions. The influence of the rate of oxygen or air supply on the rates of oxidation of silicon, manganese, and carbon was not investigated in this work. The purpose of the work originally was to study the direction of reactions toward the basic equilibria which appear to be involved. Slow rates of reaction were therefore employed in the experiments. Additional work must be done to evaluate the effects produced by higher rates of reaction. Such investigations must enter the realm of chemical kinetics. a field in which little information relative to the

melting of ferrous alloys is available.

The oxidation-reduction principles discussed in this paper are of course not limited to the elements carbon, silicon, and manganese. Reduction reactions involving the oxides of aluminum and titanium, Equations 6 and 7 are known to occur. While these reactions are outside the scope of this paper, it should be noted that such reactions become increasingly important as melting temperatures increase. Residual percentages of elements introduced by reduction of oxides may considerably change the nature of cast irons and they may be considered as a natural consequence of high temperature melting.

Oxidation vs Deoxidation

The slags generated by the metal at various conditions of concentrations and temperatures suggest certain fundamentals regarding the reaction between oxygen in cast iron and the so-called deoxidizers, silicon, manganese, and carbon. The section which follows was written with the viewpoint of re-evaluating the chemical effects of "deoxidizing" elements and with the hope of clarifying the role which oxidation-reduction reactions play in the deoxidizing process in cast

A deoxidizing reaction such as for manganese in iron is commonly written as follows:

3.
$$\underline{Mn} + \underline{O} = \underline{MnO}$$
 (1)

Formation of the oxide which then is supposed to leave the iron is referred to generally as deoxidation. However, this simplified viewpoint is now justified by data for the complex compositions of cast irons. The complicating factor lies in the ability of carbon in iron at elevated temperatures to reduce the oxides which may form. Thus the oxide which is formed on reaction with oxygen is itself capable of being reduced as indicated by the following type of reaction:

5. MnO (1) +
$$C = Mn + CO$$
 (g)

The simple experiment referred to previously where the manganese content of an iron at 2597 F was increased by adding manganese dioxide to the melt is an example of the reduction effect. Thus it is difficult to see how an element such as manganese may function as a "deoxidizer" at temperatures and concentrations where Reaction 5 may proceed since Reaction 5 prevents the formation of oxides, i.e. Reaction

3, whereby deoxidation progresses.

In a like way the reduction of silica by carbon, i.e. Reaction 4, prevents silicon in molten cast iron under normal atmospheres from reacting with oxygen as long as sufficient carbon and a high enough temperature prevails. While silica reduction has been amply illustrated, the work of Bardenhauer⁶ will be referred to because of its pertinence to the viewpoint under discussion. Bardenhauer melted down an alloy of 2.2 per cent carbon, no silicon and added a siliceous slag to the melt surface holding at temperature for 38 hr. During this period the silicon percentage increased to 3.72 per cent and the carbon decreased to 0.16 per cent. This experiment by Bardenhauer illustrates the potency of much lower percentages of carbon than occur in cast irons in reducing SiO2. From the aforementioned work and that of the author it may be concluded that there are slag and atmosphere situations in which the action of carbon inhibits the reaction between silicon and oxygen, i.e. deoxidation. and in fact causes the reverse, i.e. oxide reduction, to

A more accurate picture of the deoxidation reactions can be at least partially obtained on the basis of oxidation-reduction equilibria prevailing. For example the equilibrium constant may be utilized to determine the concentrations of silicon and carbon in iron necessary to satisfy Equation 4. At 2732 F, for the Reaction 4, SiO_2 (s) + 2C = 2 CO (g) + Si, K =1.759 and for the assumptions listed early in the paper, concentrations are related by K = Si + C2. The percentage carbon in equilibrium with a certain silicon percentage is calculated from the constant at 2732 F; for example in an iron of 1.05 per cent silicon a carbon content of 0.773 per cent is required for Reaction 4 to be at equilibrium. The importance of such equilibrim concentrations is best explained by the progress of reactions when an excess of one element is present.

In the case being considered, excess carbon may cause silicon pickup or at minimum prevent reaction of silicon with oxygen, i.e. oxidation. However, were the carbon to be oxidized below the value calculated, an insufficient concentration of carbon would be present to maintain equilibrium for Equation 4 and the

reaction might proceed to the left or at a minimum permit oxidation of silicon. This concept agrees exactly with the course of these reactions studied experimentally. Figure 5 for a melt of white iron of approximately 1.05 per cent silicon, shows that oxidation of silicon does not occur until carbon drops to a value of approximately 0.95 per cent. Although the carbon value is higher than calculated, a difference is to be expected since the conditions of the experiment are not of an equilibrium nature nor do they include an atmosphere of CO over the melt, one of the assumptions of the calculations. Further differences in calculated and observed results may be due to the fact that calculations are based on synthetic alloys and not complex cast iron compositions.

From the standpoint of deoxidation, i.e. reaction of an element in iron with oxygen, oxidation-reduction equilibria are significant in that they show it is necessary to have an excess of the deoxidizing element above that percentage which is protected by carbon. The actual percentage required to lower the oxygen content, i.e. deoxidation of an iron would be in excess of the amount in equilibrium with carbon according to a reaction of the type 4 or 5. The same reasoning may be employed in the case of elements other than manganese and silicon which may be employed as deoxidizers. The products of deoxidation, nonmetallics, provide a further complicating factor. While deoxidation is commonly written as producing only a simple metal oxide phase, i.e. Reaction 2 and 3, this does not appear to be experimentally true.

In the case of silicon and manganse in steels it has been amply pointed out³ that silicates of manganese and iron are the principle products of deoxidation at lower silicon levels while pure silica as well as silicates form at higher silicon levels. The nonmetallics formed are thus a function of metal composition and in the present work would be expected to be similar to the slag generated by the metal at a particular temperature. Since no slag material forms over molten cast iron at elevated temperatures because of the reducing action of carbon, deoxidation or reaction with oxygen would not be expected to progress at those temperatures except by reaction with carbon itself. Thus the excess of carbon present in normal cast irons would appear to limit deoxidizing effects at the higher temperature to that which is obtained by the reaction between oxygen and carbon.

From these considerations it would appear that the limiting oxygen values present in cast iron at elevated temperatures are those resulting from the limit of solubility of oxygen as affected by carbon, supersaturation being accompanied by carbon oxide evolution.

While slags or nonmetallic oxides do not form in normal cast irons at elevated temperatures, they may form once the temperature drops below the range where carbon is effective as a reducing agent. In the lower temperature ranges, nonmetallic material is generated by the iron. This material may rise to the surface or form at the surface as a slag cover. If cooling is continued rapidly as in a casting, nonmetallics form and are entrapped during solidification and appear as inclusions. A severe case of such entrapped inclusions is illustrated in Fig. 8a, 8b, and 8c. The material shown in Fig. 8 was cast from cupola metal having an analysis of 3.13 per cent C, 0.9 per cent Si

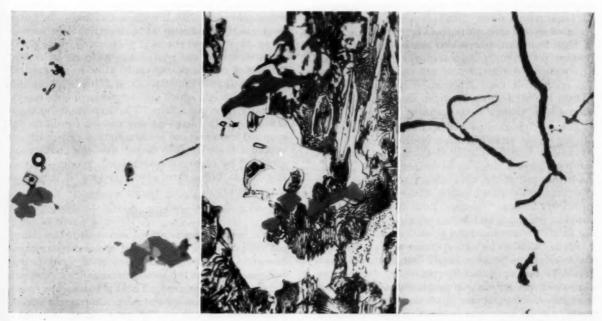


Fig. 8A (Left)—Sulphide and silicate inclusions in cupola white iron of 3.13 C, 0.98 Si. Cupola slag in contrast with the iron contained 16.6% FeO. Unetched.

Mag. 750×.

Fig. 8B (Center)-Same metal as Fig. 8A but etched in

picral to reveal distribution of nonmetallics in microstructure. Mag. 750×.

Fig. 8C (Right)—Same iron as Fig. 8A, but showing inclusion in the area of a mottled spot. Note silicates adjacent to flake graphite. Mag. 750×.

and was tapped from under a slag containing 16.6

per cent FeO.

The metal was cast in 1.2-in. diameter arbitration bars of type used for gray cast irons. The bar fracture revealed the metal to be mainly white but with light mottling. The inclusions shown in Fig. 8 are principally sulphides and silicates. The shape of the silicates indicates their formation as a liquid precipitating from the molten iron. The oxide particles vary from translucent highly siliceous pale-yellow color glasses to dark or opaque silicates evidently containing considerable FeO in solution. Coalescence of some particles is evident in Fig. 8a. The distribution of such inclusions in the microstructure of white iron is shown in Fig. 8b and in mottled areas in Fig. 8c. When flake graphite and silicate occur as in Fig. 8c, the silicates are usually a short distance off to the side of the flakes.

Similar inclusions have been reported by the author4 in normal malleable foundry white irons although the quantity present is usually far less than that shown in the severe case of Fig. 8. The microstructural and property effects caused by reactions producing these inclusions are greatly dependent on the cooling rate of the metal and for that reason are difficult to evaluate. The thermal history of the metal thus appears to be of extreme importance in influencing the degree to which the reactions may approach completion. The approach to equilibrium during cooling involves, in part, reactions and reaction rates which are a function of temperature.

At lower temperatures during cooling, the silicon and manganese in the iron may act as deoxidizers in the sense that they are enabled to react with oxygen once the inhibiting effect of carbon is lost on going to the lower temperature. Thus, it appears that the time required to cool through the lowest temperature range of 2550 F down to the lowest temperature of solidification is extremely important in permitting the aforementioned reactions to proceed. Furthermore, the progress of these reactions results mainly in a change in distribution of oxygen in the metal rather than a basic change in the quantity present.

The precipitation of nonmetallic oxide inclusions of the type shown in Fig. 8 is considered by the author to indicate that on cooling slowly through appropriate temperatures the iron generated particles within itself that are of the same nature as the slag phases described earlier in the paper. The extension of these principles to other elements such as aluminum, zirconium, titanium, and boron presents a logical field for further research in cast irons.

While inclusion formation occurs at lower temperatures, reactions between liquid cast iron and a gas phase prevail at higher temperatures. It has long been said that reducing atmospheres in contact with a molten cast iron surface promote deoxidation. Commonly carbon monoxide is considered to be a reducing gas in the sense that it may lower the oxygen content of cast iron.

However, in view of the principles presented in this paper it is difficult to understand how CO could cause any lowering of the oxygen content of a cast iron. Although a 100 per cent CO atmosphere would make oxygen less available from the gas over the melt, it does not follow that therefore the oxygen content of the metal is lower. The fact that a gentle bubbling of CO from the melt occurs at elevated temperatures either in the air furnace or when white iron is tapped into a ladle indicates that the iron is saturated with CO and that an excess of carbon is present to maintain such saturation. The ability of carbon monoxide to escape from the melt, i.e. remove oxygen, depends on the difference in the partial pressure of that gas in the atmosphere over the melt and its vapor pressure in the molten metal.

This difference of pressure is seen to decrease as CO content of the atmosphere over the melt increases. Thus with a higher CO content over the melt it would appear that the oxygen content of the melt would actually be increased rather than decreased because of any so-called reducing action of CO. Chipman and Marshall⁵ have demonstrated that raising the CO pressure considerably above atmospheric pressure causes a marked increase in the oxygen content of pure iron-carbon alloys. Their experiments emphasize the point being made. Thus, there appears to be valid doubt for considering CO as a reducing gas, or deoxidizing agent, with respect to its ability to remove oxygen from molten cast iron.

The discussion presented in the preceding paragraphs should provide a more complete picture of the principles involved in deoxidation phenomena than has heretofor been considered in cast iron melting. In the case of the effect of carbon monoxide in the atmosphere over the melt, there is actual disagreement between the viewpoint expressed and that usually presented in foundry literature. The main objective of the author in this discussion has been to indicate the relationship of deoxidation mechanisms to the physical and chemical phenomena, particularly oxidation-reduction reactions, which occur in the melting of cast irons, especially white iron employed in the malleable foundry. In so doing, explanations have been sought which are in agreement with experimentally observed results. The important influence of carbon in cast irons on slag and gas equilibria has required modification of deoxidation principles advanced for steels. It is hoped that the clarification provided by such explanations will produce further research to bring these considerations to their ultimate conclusion.

Summary

Cast irons are essentially of non-equilibrium chemical composition in the normal melting temperature ranges. Upon being heated, oxidation or reduction reactions between the molten iron and its surroundings begin and progress. These reactions are fundamentally dependent on the temperature and the composition of the iron and its surroundings. Considering temperature as a variable and molten cast iron, air, and a silica crucible as the reactants, the data presented herein substantiate the following conclusions:

1. Silicon and manganese in molten iron are read-

m

R

ilv

in

m bo

in te

CC

by H

> th ca

p tl a

Ci P vi g

0 0

a

th

NS

ke

11

of

ng

es

ed

th 11-

de

ds

in

re

0

S.

it

le

d

)-

g

11

S

g

f a

ily oxidizable below 2550 F.

2. The oxidation of carbon is extremely slow below 2550 F as long as sufficient silicon is present as an inhibitor.

3. Below 2550 F oxidation products persist as a scum or dry slag over the melt; the scum indicating an excess of silicon in the melt. Only very slow oxidation of carbon will occur as long as the dry scum

4. The dry slag appears to be a mixture consisting mainly of crystobalite and magnetite, along with a

small amount of fayalite.

5. At temperatures above 2600 F, the reducing effects of carbon prevent the formation of a slag. Reaction between carbon and a gas atmosphere is the normal course of events.

6. Oxidation of silicon and manganese from the melt occurs less readily as temperature increases. Carbon is an inhibitor of oxidation of these elements at elevated temperatures as long as a sufficient carbon

concentration is present.

7. Silica reduction, i.e. silicon pickup, by carbon in molten cast iron occurs when the equilibrium temperature 2671 F is sufficiently exceeded. Manganese oxides may also be reduced when present in sufficient concentration.

8. Silicon pickup occurs in the direction indicated by an equilibrium constant for Reaction 4. SiO₂ (s) + 2 C = 2 CO (g) + Si. A gentle CO boil is a normal condition showing that this reaction is progressing whenever molten cast irons are contained in siliceous refractories.

Considering composition as a variable with molten cast iron, air, and a silica crucible as the reactants, the following conclusions may be stated:

1. Silicon concentration regulates the oxidation of carbon and the appearance of a fluid slag at lower temperatures.

2. The fluid slag developed when silicon no longer

inhibits its formation consists mainly of fayalite.

3. A carbon "boil" occurs when the silicon concentration is below the percentage which will protect carbon at low temperatures.

4. Carbon percentage regulates the reduction of silica or the oxidation of silicon at elevated temperatures. The direction of the reactions is toward equilibrium for the equation SiO_2 (s) + 2 C = 2 CO (g)

5. The slag phase formed at elevated temperatures when carbon is no longer effective as an inhibitor is a dry scum consisting mainly of crystobalite, magnetite, and a small amount of fayalite.

6. Manganese, like silicon, is protected from oxidation by carbon as long as the carbon concentration is

sufficiently high.

7. The slag phases generated by a molten cast iron are dependent on composition of the iron with either a dry slag, a fluid slag, or no slag occuring according to the dictates of the composition.

In conclusion, it can be said that the nature of normal cast iron compositions is such that chemical reactions will begin and progress in the direction of the fundamental oxidation-reduction equilibria wheneyer they are heated or cooled through temperature ranges encountered in melting operations.

Acknowledgments

The author wishes to extend his appreciation for assistance and suggestions to the following: Prof. G. J. Barker, P. C. Rosenthal, and D. J. Mack, and Research Metallurgists L. F. Porter, and E. H. Lange, all members of the department of Mining and Metallurgy, the University of Wisconsin. Thanks are directed to Mr. W. V. Osborne, President, Lakeside Malleable Castings Co., Racine, Wis., and members of his organization for supplying cast iron melting stock for the experiments.

PART II

Air Furnace Melting

IN PART I, some principles regulating composition changes in laboratory melted cast irons have been presented. Extension of the laboratory principles to the cold-melt air furnace melting process employing an acid lining provides the purpose for this paper.

From the results of laboratory experiments, the direction of chemical composition changes of molten cast irons at various temperatures can be accurately predicted. At lower temperatures carbon is oxidized very slowly while oxidation losses of silicon and manganese are greater, the latter being most readily oxidized. At higher temperatures, the reducing action of carbon in molten cast iron inhibits the oxidation of silicon and manganese with the result that the only appreciable loss is that of carbon.

The foregoing statements are made in relation to the melting conditions of the experimental work of Part I. It is a point of concern whether principles established by laboratory melting experiments may be applied to a commercial melting process. Such direct applications are only justified when supported by data and conclusions derived from plant operations.

Melting Practice

Data for evaluating the relationships between the laboratory results (Part I) and the commercial coldmelt air furnace melting process for malleable iron castings were obtained from two foundries. Both foundries followed the general practice of melting down a charge of high carbon content, about 2.7 to 3.0 per cent and higher than desired silicon content, about 1.1 to 1.35 per cent. Manganese in the charge was about 0.10 to 0.15 per cent higher than that desired in the final analysis, melting loss taking care of the excess. A preliminary test for carbon, silicon, and manganese was made about 40 to 60 min before tapping in one foundry. The other foundry, henceforth called Foundry B, made preliminary determination of carbon about 30 to 40 min prior to tapping.

The first foundry, Foundry A, employed additions of ferro-alloys whenever the preliminary analyses for silicon and manganese were low. Foundry B used additions less frequently and only as indicated by analysis of the previous day's heats. Furnaces in both foundries were acid-lined with silica hearth bottoms. The fuel was powdered coal in each case. Foundry A operated two furnaces of the same hearth area with charges ranging from 20,000 to 51,000 lb. One of the two furnaces was used for smaller charges, 22,000 to 36,000 lb, and it therefore had a shallower hearth. Foundry B operated one furnace with a charge running from 28,000 to 32,000 lb. Tapping period varied from 31 to 71 min depending mainly on size of heat. Preliminary analysis samples in both plants were taken after the slag was skimmed and the metal temperature reached about 2625 to 2725 F.

When the metal has reached the temperature range of 2625 to 2725 F, silicon and manganese oxidation losses should have ceased according to the laboratory results. Further increase in temperature should result in silicon pickup. To follow composition changes more closely, four samples were taken during the later stages of the heat. The four samples consisted of the preliminary bar, one taken 40 to 60 min later at the beginning of tapping, one-half way through the tap, and a final bar at the tail end of the heat.

Silicon

Data from 53 heats in Foundry A will be considered to illustrate the changes in silicon percentage in the molten iron. In all cases, the analyses showed a gradually increasing percentage of silicon throughout the period from preliminary sample to the end of tapping. The average silicon pickup from preliminary to beginning of tap was 0.09 per cent, from preliminary to middle of tap for 0.11 per cent, and from preliminary to tail end of tap was 0.16 per cent. The same general trend was noted in data on 29 heats from Foundry B. In other words silicon pickup prevails in silica bottom-lined air furnaces as it does in laboratory melting in a silica crucible. Silicon pickup from silica reduction should be differentiated from increased silicon percentage due to the addition of the ferro-alloy.

In all discussion which follows, reference to percentage silicon pickup refers to net percentage of silicon increase due only to the reduction of silica. In heats where ferrosilicon was added after the preliminary, the total increase in percentage silicon prior to tapping generally exceeded the percentage silicon added, i.e. there was a net silicon pickup. Only three exceptions were noted in 81 heats. It is therefore concluded that 100 per cent recovery of silicon from ferrosilicon additions is to be expected in view of the reducing effect of carbon at these temperatures in preventing oxidation of silicon. The previous observation assumes that the ferrosilicon is added to a clear, relatively slag-free molten metal surface.

Because of wide variations in percentage silicon increases on different heats, frequency charts are used to present the data. Figure 9 shows the distribution

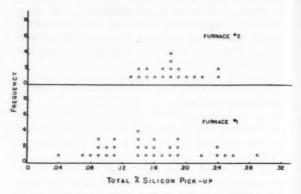


Fig. 9—Frequency of percentage silicon pickup in 53 heats made in air furnaces in Foundry A. Silicon pickup refers to net total increase from preliminary test to tail end of heat.

iı

F

ca

si

di

of silicon pickup values measured from preliminary to end of heat for two different furnaces in Foundry A. Furnace No. 2 shows a decidedly narrower range of pickup, 0.11 per cent Si, as compared with the wide variation of Furnace No. 1, 0.26 per cent Si. The two furnaces were of the same hearth area but Furnace No. 2 had a deeper bath of metal. A major difference in the melting practice of the two furnaces was the size of the charge. Charges in Furnace No. 2 ranged from 35,000 to 51,000 lb while those in Furnace No. 1 ranged from 20,000 to 38,000 lb with the majority being 22,000 to 28,000 lb. Since the weight of the charge greatly influences the time required for melting, the variation in weight would appear as one possible reason for the differences in total silicon pickup between Furnaces 1 and 2 as shown in Fig. 9.

When the data from Foundries A and B were plotted with silicon pickup as a function of weight of charge, the scatter diagram of Fig. 10 was ob-

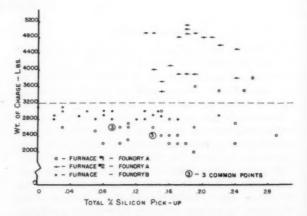


Fig. 10—Net percentage silicon pickup from preliminary test to tail end of heat plotted for different weight charges. Data from 81 heats in three furnaces are shown.

ed

tained. Heavier charges are seen in Fig. 10 to present less fluctuation or a narrower range in pickup values. Significantly, four heats in Furnace No. 1 in which the charge weight approached that used in Furnace No. 2 were found to fall within the narrower pickup range of the latter furnace. Data from Foundry B are included in Fig. 10. The latter foundry used only the smaller charges and it is seen that their results for silicon pickup present the same scatter as occurred with the lighter heats in Furnace No. 1, Foundry A. The main intent in presenting Fig. 10 was to illustrate that differences in total silicon pickup obtained in different furnaces may be dependent on more than one melting variable.

The test sample selected as representing the "final" or average silicon analysis of a heat may be visualized as a spot check on the percentage silicon in the molten metal at a particular time during the period of silica reduction. Variations in the progress of silica reduction would therefore cause fluctuation in the final silicon analyses. A frequency chart of "final" silicon percentages for the furnaces being considered in Foundries A and B is presented in Fig. 11. In Foundry A, Furnace No. 2 is seen to have a much narrower range of "final" silicon values than Furnace No. 1. The narrowest range occurred in the furnace of Foundry B.

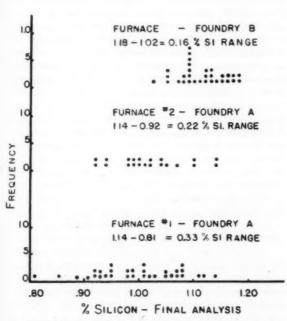


Fig. 11-Frequency of final percentage silicon analysis in 81 heats in three different furnaces.

This difference in the final analysis range may be due to the differences in total silicon pickup illustrated in Fig. 10. A widely fluctuating range of total silicon pickup values might be expected to produce a wide range of "final" silicon percentages as is the case for Furnace No. 1 of Foundry A. The "final" silicon in this case appears to be related to the reproducibility of total silicon pickup for this furnace.

However, this relationship did not exist in the results from Foundry B. Other operating variables, possibly slag, are evidently involved. Existing relationships could be established statistically with a greater supply of data and by measuring more variables, but it is not the purpose of this paper to determine the characteristics of individual furnaces. Rather it was desired to point out that the later stages of melting cast iron with the metal at elevated temperatures produce conditions under which the reducing power of carbon in molten iron is predominant.

It is concluded that silica reduction is a normal high temperature phase of cold-melt acid-lined air furnace melting. For that reason, it is characteristic that the per cent silicon in the melt continually increases during the final stages of a heat. The variation in silicon pickup values reported in Fig. 9 and 10 show that there are uncontrolled factors operating during the melting period when silica reduction is a positive and progressing reaction.

Carbon

Unlike the case of silicon, carbon loss by oxidation progresses more readily as temperature increases. Once the temperature exceeds about 2550 F and during the period from the preliminary sample to the end of the heat, decarburization of the melt continues without a halt. These carbon losses are of two kinds. A minor loss is the result of silica reduction. Where an average of 0.16 per cent total silicon pickup was reported, a chemically equivalent carbon loss, 0.137 per cent, must occur. Silica reduction by carbon causes the gentle but continuous boil evident in the bath in air furnace melting. Nothing can be done about the aforementioned carbon loss as long as the furnace is acid lined. The reaction SiO_2 (s) + C = Si + 2CO(g) is spontaneous and will progress as long as the metal is above the equilibrium temperature, 2671 F, for the reaction (Part I). This carbon loss by silica reduction may be considered a basic minimum loss depending mainly on the time at elevated tempera-

A second and major loss of carbon occurs as a result of decarburization or direct oxidation, at the molten metal-atmosphere interface. In the absence of slag covers, direct oxidation of carbon is mainly dependent on the surface area and mass of the melt, the furnace atmosphere and the rate of change of the atmosphere. For example, direct oxidation can be completely eliminated if a 100 per cent carbon monoxide atmosphere is held over the melt so that oxidation by the furnace atmosphere is not possible. The other extreme is reached when a raw air blast is directed at the metal surface so that rapid direct oxidation of carbon may occur as reported in Part I.

Thus it would appear that the relative percentages of carbon monoxide, carbon dioxide, and oxygen in the furnace atmosphere have an important bearing on the loss of carbon by direct oxidation. Occasionally atmosphere control is exercised in malleable foundry air furnace melting. However for the purpose of exercising carbon control, it is more common to make additions of graphite, petroleum coke or

other recarburizers to the surface of the metal after the preliminary carbon analysis has been determined. The size of the additions is based on experience which has been shown to give the desired carbon content at the time the final sample of the heat is taken. For example in Foundry B, an average addition of 100 lb of graphite and 130 lb of petroleum coke was made to heats of 28,000 to 30,000 lb when the preliminary carbon ranged from about 2.5 to 2.6 per cent and the final desired was 2.20 to 2.40 per cent. This would amount to an addition of about 0.8 per cent C if it is assumed that all the carbon is dissolved by the metal. This carbon, of course, does not all enter the metal, as part of it is lost directly from the surface of the metal. However even if it did, there would be a loss of not more than about 1 per cent carbon during the period from the preliminary to the final test samples. Since this loss occurs in about 11/2 hr, it follows that the rate of carbon oxidation is less than I per cent per hour. This rate of carbon loss is less than that attained in laboratory induction melting (Part I) where an air blast was used over the melt but is about the same as that occurring when induction melting is carried on in still air (Part I).

Although the air furnace atmosphere might be expected to be less oxidizing than still air because of its carbon monoxide-carbon dioxide content, the rate at which the atmosphere changes in the air furnace alters this situation. It is therefore, not until moving air is directed at the surface of the metal that oxidation loss of carbon in induction melting exceeds that of air furnace melting. Therefore, holding the metal at constant temperature in still air in an induction furnace does not appear to be a more oxidizing condition than air furnace melting with respect to oxidation of carbon or other elements at comparable temperatures. The foregoing observation is one of value to those concerned with using the induction furnace for research in malleable irons.

In either melting medium, oxidation of carbon progresses at a rapid rate at elevated temperatures. Although the oxidation of carbon occurs, its concentration remains high enough to cause the reduction of silica throughout the later stages of the heat. As reported in Part I, carbon is the active element in oxidation and reduction reactions in cast iron melting at elevated temperature. The exact rate of carbon oxidation in individual air furnaces is therefore a function of practical melting variables of furnace atmosphere, slag covering if any, recarburizers, and time at temperature.

Manganese

The experience with manganese content changes in laboratory and foundry is similar if not identical. Oxidation loss of 0.10 to 0.20 per cent Mn occurs during melt down and afterwards at low temperatures. At elevated temperatures, only limited loss occurs due to the protective action of carbon. For example, from preliminary to final analysis in the air furnace a loss of 0.01 to 0.04 per cent is experienced. This loss is of the same order as that reported by the author for induction melting (Part I) and for similar reasons.

With respect to manganese again, the oxidation-reduction reactions regulated by carbon at elevated temperatures are similar in acid-lined induction and air-furnace melting furnaces.

Discussion

The scatter of silicon pickup values reported in Fig. 9 and occurring during the silica reduction period of melting raises questions. In laboratory melting, the silicon increase for a given time and temperature is relatively reproducible from heat to heat. Why is this not so in the data of Fig. 9? Some other melting variable is evidently affecting silicon pickup. It is postulated that slag present on the melt may be an important factor. Lange and the present author7 have shown that slag containing iron oxide may inhibit or prevent silicon pickup at elevated temperatures. Higher iron oxide content slags7 may prevent silicon pickup or actually cause a silicon loss at temperatures where pickup would be expected if only furnace atmospheres and refractories were present. Lower iron oxide slags7 may not cause silicon loss but do inhibit or diminish the amount of silicon pickup. Slag thus may be considered an important factor in changing the percentage of silicon pickup during the later stages of the heat.

Referring to the laboratory work (Part I), considerable data was presented to show that a slag phase was generated by the iron itself at temperatures below 2550 F. White iron contains sufficient silicon so that at the lower temperatures a slag phase does form by oxidation of the metal. However this slag disappears on heating due to reduction by carbon. Slag only occurs on the metal surface at the higher temperatures if it is introduced from an extraneous source. Such slag may come from the refractories, coal ash, sand or some other source. Once the slag is skimmed at temperatures above 2550 F, further slag accumulation should originate only from the aforementioned sources, i.e., not from the iron itself. Thus once the low temperature oxidation products are removed by skimming, the surface of the metal will remain clear unless a slag develops from some source other than the metal. During the latter period when the surface is clear, oxidation of the carbon proceeds rapidly and silicon pickup occurs as the temperature increases.

The preliminary and final silicon analysis samples are taken during this period. Were it not for the mitigating effects of the extraneous slag, carbon would show a progressive and continuous drop from preliminary sample to the tail-end of the heat while silicon would show a gradual and continuous increase during the same period. The extent of the carbon oxidation would be regulated mainly by the furnace atmosphere and its rate of change. The silicon pickup would be effected mainly by the temperature and time at temperature for a given charge weight and furnace size. Composition changes of the type referred to above are a natural consequence of the fact that the normal composition of cast irons are not at equilibrium with silica refractories or normal furnace atmospheres at temperatures usually encountered in melting.

ì.

I

n

1

n

2

Conclusion

The basic principles of oxidation-reduction reactions established in the laboratory and reported in Part I may be applied to air furnace melting. Silicon and manganese losses occur primarily during melt down and at low temperatures during normal operation of the air furnace. The final stages of a heat in an acid-lined furnace results in increasing silicon content of the melt and protection of manganese, both effects being caused by the reducing power of carbon at elevated temperature. Carbon oxidation of two different mechanisms progresses most readily at elevated temperatures, one being the result of silica reduction and the other being the result of direct oxidation of carbon by the furnace atmosphere. Finally, a test sample for chemical analysis taken during the tapping period represents a spot check made on a molten cast iron whose composition is gradually changing in the direction governed fundamentally by chemical oxidation-reduction equilibria.

Acknowledgments

The author wishes to express his sincere thanks to those from the industry who gave assistance in obtaining data for this investigation; particularly, to the following: Mr. J. Scales, Chief Metallurgist, Mr. John Yarne, Metallurgist, Mr. W. B. Sobers, Metallurgical Chemist, and Mr. J. Kropka, Foundry Superintendent of the Chain Belt Company, Milwaukee, Wis.; Mr. W. V. Osborne, President and Mr. R. V. Osborne, Vice President, Mr. L. Osborne, Secretary, Mr. M. Harder, Foundry Superintendent, and Mr. H. McCabe, Metallurgical and Chemical Laboratory, all of the Lakeside Malleable Castings Company, Racine, Wis. The author also wishes to thank his associates, Prof. P. C. Rosenthal, Research Metallurgist, L. F. Porter, and Prof. G. J. Barker, Chairman of the Department of Mining and Metallurgy of the University of Wisconsin for their suggestions and encourage-

Bibliography

1. J. E. Rehder, "Silicon Pickup in Melting Malleable Iron," AMERICAN FOUNDRYMAN, vol. 10, no. 4, p. 50, October, 1946.
2. R. E. Savage and H. F. Taylor, "Fayalite Reaction in Sand

Molds Used for Making Steel Castings," A.F.S. TRANSACTIONS, vol. 58, pp. 564-577 (1950). 3. M. Baeyertz, Non-Metallic Inclusions in Steel, American

Society for Metals, p. 22-29.
4. R. W. Heine, "Some Effects of Deoxidizing Additions on Foundry Malleable Irons," A.F.S. TRANSACTIONS, vol. 58, pp. 277-297 (1950)

5. J. E. Chipman and S. Marshall, "The Carbon-Oxygen Equilibrium in Liquid Iron," Transactions, ASM, vol. 30, p. 695 (1942).

6. P. Bardenhauer, Die Neu Giesserie, vol. 30, pp. 161-166 (1943).

7. E. A. Lange and R. W. Heine, "Some Effects of Temperature and Melting Variables on Chemical Composition and Structure of Gray Irons," A.F.S. Transactions, vol. 59 (1951).

DISCUSSION

Part I

Chairman: C. F. LAUENSTEIN, Link-Belt Co., Indianapolis, Ind. Co-Chairman: F. Coghlin, Jr., Albion Malleable Iron Co.,

J. E. REHDER (Written Discussion): 1 In this paper Prof. Heine continues his good work on the metallurgy of oxidation-reduction reactions of cast irons. The writer finds himself in general agreement with Prof. Heine, and the comments to follow should be considered as contributions to the field covered.

In work relating thermodynamic data to practice, simplifying assumptions must always be made, and they are sometimes awkward. The use of unity for the concentration of CO over the melt as suggested by Prof. Heine would possibly be better replaced by a value of 0.3 due to the diluting effect of nitrogen in commercial operations, and as such should be useful for the conditions in electric arc furnace melting or for the conditions in a cupola well; but above the tuyeres in a cupola or in an air furnace the blast or flame movement further dilutes the CO with carbon dioxide and sweeps it away, and the concentration of CO is probably close to 0.1. This encourages reduction of SiO, by allowing the reaction

 $2C + SiO_2 = Si + 2CO$

to proceed to the right in accordance with mass law considera-

On the other hand the use of unity for the concentration of SiO₂ is justified, from the viewpoint of thermodynamics, wherever free SiO2 molecules are present; which is the case not only in a silica crucible but for all commonly used acid refractories including super-duty fire-brick. A smaller quantity of free SiO, will decrease the rate at which the reaction proceeds, which, of course, is not treated by thermodynamics.

Although the effect is probably a minor one, no mention is made by Prof. Heine of the fact that the carbon in the equations is actually dissolved in iron. Darken and others have shown that the carbon in liquid iron-carbon alloys is combined with iron, and this will affect the free energy equations by the heat of solution and compound formation. This is shown by the fact mentioned by Prof. Heine that the addition of coke or graphite to molten iron increases reduction of SiO2.

The experimental results given in the paper of oxidation of cast iron melts by air and oxygen, shown in Figs. 3 to 6 inclusive, are typical of reactions in converters as would be expected, and are corroborative of work done two or three years ago in England in connection with side-blow converter practice. It was shown clearly there that provided the cupola-melted hot metal supplied to the converter was sufficiently high in temperature, a successful blow and high final temperature could be obtained by carbon oxidation almost alone. This permits lower silicon content in the cupola metal and resulting lower cost. In older converter practice with lower cupola metal temperatures, a fairly highsilicon content was necessary to provide an ignition effect, the oxidation of silicon raising the temperature to the point where carbon oxidation can proceed rapidly.

The slag analyses and considerations presented are of interest, and in general corroborate work done by the writer in 1942 on the oxidation of cast iron by additions of oxygen or mill scale. A material balance made from analyses of slag and metal showed then that for oxidation by gaseous oxygen, 80 per cent of the added oxygen could be accounted for.

The statement by Prof. Heine that "most cast irons are of non-equilibrium composition at high temperatures" is not considered correct as it stands, since the effects of atmosphere and of crucible chemistry are not included, and these are precisely the influences producing the chemical changes and disequilibrium. Used in Prof. Heine's sense, it can be stated that a lump of coal is not in equilibrium, since it can be burned if heated and supplied with oxygen.

Prof. Heine's final remarks on oxidation-deoxidation are apparently in agreement with the writer's comments of 1950 made during discussion of a paper by Prof. Heine,4 wherein the present writer took the position that at the carbon and silicon contents and the temperature levels normally used in cast iron melting, equilibrium oxygen contents must be low due to the reducing or deoxidizing action of the carbon and silicon. The statement that CO cannot remove oxygen from cast iron melts is apparently without meaning, since although a pure CO atmosphere will inhibit Reaction 4 and decrease the bubbling of CO gas from the melt, the oxygen in this CO comes from the

SiO₂ refractory walls, not from the melt directly.

MILTON TILLEY: ² I think that the figures embodied in the paper are correct as far as they go. They demonstrate, again, the effect of the side-blow converter, as Mr. Rehder pointed out.

¹ Foundry Engineer, Department of Mines & Technical Surveys, Ottawa. Ont., Canada.

² Metallurgist, National Malleable & Steel Castings Co., Cleveland.

Those of you who have operated a side-blow converter probably recall getting melt in there so cold that it is covered with a scum. You could not get the metal hot, and the fact that the silicon is still in there when it gets hot it is swirling around and you do not have the covering of the slag. These semiequilibria conditions we find in an air furnace where the metal is quiet and the metal is fairly high in temperature might be a little misleading to the operator figuring he could not depend on oxidation of silicon.

Let us say you have melted down the cupola and it is out of control and you have oxidized your elements and your iron so the iron gets into the air furnace highly oxidized, under such conditions I think you will find you can take that oxidation out of the metal by the oxidation of silicon. We have done some work on that in relation to the effect on nodule count and we found that if you put enough ferrosilicon in your iron to raise your silicon up to where it is supposed to be, you still do not have the same iron as if you had melted it down that way. However, if you put a little more silicon in and you will get your nodule count back to normal; we theorized this as meaning that you have deoxidized your metal with the silicon. However these reactions might indicate to an inexperienced man that you could not use silicon as a deoxidizer. I think you will find you can do it, which I presume arises from the concentration of silicon.

H. C. STONE: 1 I would like to ask Mr. Tilley whether he recognized the fact that he was using aluminum for deoxidation?

MR. TILLEY: Yes, we recognize that. The amount of aluminum is not enough to do what we found is being done.

MR. STONE: How much aluminum are you using? MR. TILLEY: In a 30-ton bath we use up to 300 lb of aluminum.

MR. STONE: That is an appreciable amount.
MR. TILLEY: Yes, that is right, but it would not amount to enough aluminum.

MR. STONE: The silicon has about 11/2 per cent aluminum and 300 lb is an appreciable amount.

MR. TILLEY: Let us analyze the amount for a moment. There is 3 lb of aluminum in 30 tons of metal. That is 0.005 per cent

or 0.10 lb per ton. That is not enough.

C. L. Adovasio: 'Some time ago an English metallurgist discussed the use of boron and the beneficial effects derived with one form of boron and not with another. He found by experiment that when using equal amounts of boron from two different sources, ferroboron in one case and borax in another, beneficial effects were perceptible only when using the ferro-boron. He concluded that the beneficial effects were due to the aluminum contained in the ferroboron and not to the boron content itself. Since the aluminum content in the quantity of ferroboron generally used in treating malleable iron is very small, it would seem that the amount of aluminum in 300 lb of ferrosilicon, by comparison, would be great. Consequently, the deoxidation attributable to the aluminum, when using a large quantity of ferrosilicon, would also be great.

Part II

Chairman: W. D. McMILLAN, International Harvester Co., Chicago.

Co-Chairman: WM. ZEUNIK, National Malleable & Steel Castings Co., Indianapolis, Ind.

MR. REHDER (Written Discussion): In general the data presented on silicon pickup in air furnace melting agree with that previously given by the writer (See AMERICAN FOUNDRYMAN, vol. 10, no. 4, p. 50, October, 1950) and are more extensive in that a greater number of heats were sampled. There is considerable spread of values shown, and this is considered due to the effects of two variables not reported; one is flame condition and the other metal temperature. These are both of major importance and both vary considerably from day to day and from foundiv to foundry. The flame quality adjustments by the melter are largely subjective, and are made in accordance with the progress of the melt, being different nearly every heat; otherwise an air furnace could be run by an automatic sequence control on the pulverizer and secondary air controls. The hottest flame is an oxidizing one due to the carbon dioxide content, and CO rising from the melt is rapidly swept away and diluted by the

relatively very large volume of combustion gases. On the other hand if oxidation must be decreased, an adjustment can be made that will result in carbon being dropped into the bath from the flame.

The effects of slag are in practice also complicated. Although the bath be skimmed clean for maximum rate of heat absorption, after tapout skimming of slag soon becomes impossible and slag rapidly accumulates. Silicon pickup occurs during both coid melt and duplex melting, although the slag in the latter case contains about one-quarter as much FeO as in the former case.

CHAIRMAN McMillan: It appears that the work the author did in the small induction furnace may well be applied to melting in the air furnace in full production tonnages.

We hear that "silicon pickup" or increase in silicon content toward the end of the heat does "not count"; i.e., the silicon is in such a form that it does not have the same effect as silicon melted down from the charge with respect to the formation of primary graphite and of enhancing annealability. The author's data indicate that it is ferrosilicon and has the same effect as silicon originally present in the melt. Is that correct?

MR. HEINE: It would appear that way. I do not know. I do not see why it should have any effect. It is the same silicon we

have been talking about.

CHAIRMAN McMILLAN: Is it ferrosilicon?

Mr. Heine: I do not know whether it is ferrosilicon or not. I do not know how it exists.

CHAIRMAN McMillan: The silicon increase or pickup is the result of the reduction of SiO2 and is definitely silicon as such with the same characteristics as silicon introduced in the charge.

MR. REHDER: On that particular point, Mr. McMillan, I can corroborate you from personal experience and cross checking of commercial melts when I was working in malleable foundries. I am satisfied that the silicon picked up from the furnace bottom or from the reduction of silica acts the same way as the silicon incorporated in the charge in any other way. As we all know, silicon added late in the bath sometimes has a tendency to produce mottle, but if the bath temperature is increased or holding time in the furnace is increased, the mottling can be eliminated. I am satisfied that silicon from reduction of silica acts the same as any other form of silicon introduced.

MR. HEINE: I did not understand Mr. McMillan's question. I agree that the silicon that goes into the metal is metallic silicon. I misunderstood the point. I thought that he was asking whether that silicon was combined with the iron in some way.

I agree that it is silicon in the iron.

W. R. JAESCHKE: 5 I think most foundrymen attempt to control the melting furnace atmosphere. They try to start with a rich flame and with as little oxidizing as possible to preheat the charge. Then they melt it down quickly by increasing the heat input and the flame temperature as they go along close to a theoretical relationship of fuel and air and superheat the metal as quickly as possible. The slag is generally skimmed off early in the heat to provide a better heat transfer rate. Then after the temperature and composition are attained most operators will again try to control the composition by introducing a rich flame again, and they reduce the ratio of air to coal and probably change the coal fineness intending to drop some of that carbon on the bath to retard the carbon loss under those high temperature conditions.

Of the variables, probably the most potent is the depth of the bath. It is very difficult to have a shallow bath and control the temperature and if you do not control the temperature you do not control the carbon. Probably the range in variations is more due to the depth of the bath than to other operating con-

ditions.

I think most practical operators appreciate these technical papers as they explain things they have learned and heard from old timers many years ago. Many of these things were expressed in phrases of rule-of-thumb stories like that in an earlier paper where the silicon burns out, the manganese burns out and then carbon and in either the air furnace or converter practice; that is quite an old story to converter operators. The story was, "if the metal were hot enough the carbon would go out first." That has been known but not explained nor graphically shown as it has been in this paper and I think the practical operators get a lot of pleasure out of hearing these things explained as they have been.

⁸ Asst. Works Mgr., Belle City Malleable Iron Co., Racine, Wis.

⁴ Metallurgical Engr., Ohio Brass Co., Mansfield, Ohio.

⁵ Consulting Engineer, Whiting Corp., Harvey, Ill.

MELTING ALUMINUM AND MAGNESIUM-BASE ALLOYS

By

L. W. Eastwood*

PART I - MELTING PRACTICE FOR ALUMINUM - BASE ALLOYS

Types of Melting Equipment

THE FOUNDRYMAN must usually use the equipment at hand. If, however, new equipment is being obtained, the foundryman has a considerable choice of furnace types. Furnaces may be heated with coal, coke, oil, gas, or electricity. Gas- and oil-fired furnaces are the most common, but recently the low-frequency induction furnace is becoming quite widely used, though it still accounts for a relatively small part of total furnace capacity. Pot-type holding and reverberatory melting furnaces, equipped with electrical-resistance heating elements and high-frequency induction furnaces, have been used but are of relatively little importance for melting light alloys.

The fuel-fired furnaces are of two types: the indirect-flame, in which the products of combustion do not come directly in contact with the metal charge; and the direct-flame furnace, in which the products of combustion pass directly over the charge. The indirect-flame furnaces may be of the lift-out refractory, crucible type, or may be stationary or tilting. The direct-flame type is usually represented by the reverberatory, the barrel, and newer Sklenar reverberatory

furnaces.

be ath gh on, lag old ase se, to to

on

as

ve

11

h

n

of

11

v,

g 1.

> Gas-fired radiant tubes have had very limited use for open-hearth melting. The advantage, of course, is that the metal charge is protected from the products of combustion. The disadvantages are that such methods of heating have a high first cost, are less efficient, and are slower than the direct-flame or other indirect types of heating.

> The degree to which all fuel-fired pot-type stationary and tilting furnaces approach the indirect-flame type depends upon the design of the top of the furnace. In some instances, a very good seal is maintained between the cover and the top of the crucible so that the products of combustion cannot pass over the melt. In such instances, the products of combustion

escape through a port in the back of the furnace. If the ports are too small, some back pressure is produced which may cause some gas penetration through the wall of refractory crucibles. In other designs, it is possible for all or part of the products of combustion to pass over the top of the charge and out through a center hole in the cover, with or without an additional exhaust port in the rear of the furnace.

Cast iron pots are supported by the top flange, which forms a fairly complete muffle, thereby avoiding contact of the melt with the products of combustion. Refractory crucibles, on the other hand, are supported on a refractory stool. In some instances, the refractory underside of the furnace cover is flush with the crucible, forming a complete muffle type which excludes the products of combustion from the melt.

Choice of Melting Equipment

The selection of the melting equipment depends upon local conditions. The method of heating chosen is usually the cheapest and most reliable for the local conditions. However, the problem of furnace choice is not a simple one because, in addition to the reliability and cost of the heat source per pound of metal melted, other factors, such as initial costs of the melting equipment, installation cost, maintenance costs, relative working conditions in the melting room, temperature control, quantity of metal required per shift, the size of the melts needed, the method of pouring, and the relative melt qualities produced in the various types of furnaces must all be considered In general, the best melt quality, closest temperature control, and the best working conditions in the melting room are obtained by the use of electric furnaces. The greater recent advances in the cost of oil and gas and the unreliability of their supply in some localities, have helped to produce a trend in favor of electric furnaces. However, in some locations, electric power is costly also and in short supply.

Supplying a large quantity of metal and melting large scrap may make a reverberatory furnace mandatory. For reclaiming fine scrap such as borings and

^{*} Supervisor, Nonferrous Metals, Battelle Memorial Institute, Columbus, Ohio.

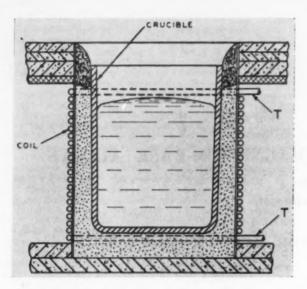


Fig. 1—A schematic sketch showing a cross-section of a high-frequency induction furnace of the tilting type. The crucible is rammed in place. The water-cooled copper coil "T" carries the high-frequency current.

turnings, the low-frequency electric furnace is finding considerable favor.

The quantity of metal required per shift, and the type and size of the castings poured, of course, largely determine the size of the melting unit. Reverberatory furnaces and electric induction furnaces are not suited for melting where frequent changes in alloy composition are necessary.

The absorption of impurities is also an important consideration in the choice of melting equipment. In addition to the absorption of iron from cast iron tools and melting pots, there is also the possibility of silicon pickup by the partial reduction of firebrick and similar refractories used as hearth linings. This is not usually a serious problem, because the great majority of the aluminum-base foundry alloys contain substantial amounts of silicon as the principal alloying element. However, silicon absorption from the refractories when melting aluminum alloys containing larger amounts of magnesium or zinc can be quite serious because the silicon is a very harmful impurity in alloys of this type. Silicon pickup is usually worse in freshly lined furnaces. High-alumina linings reduce this difficulty, as does the employment of either chromite or zirconium silicate refractories for the hearth.

Furnace Designs

Fuel-Fired Lift-Out Furnaces—In the lift-out furnaces, the crucible is placed inside of a pit-type furnace combustion chamber by means of tongs. After the melt has been prepared, the refractory crucible is removed with tongs, placed in a ring shank, and castings poured, usually directly from the crucible in which the melt was made. If a cover is not used on the crucible, this type approaches the conditions of

an open-flame furnace. It is especially useful for small operations where maximum flexibility is needed, particularly in respect to variety of alloys. Refractory crucibles, usually clay graphite or silicon carbide, are used. Grain refiners may be added, sodium modification effected, or degassing treatment applied after the melt has been made and the flame turned off, as described later in this paper.

Fuel-Fired, Stationary, Pot-Type Furnaces—The stationary furnace is usually gas or oil fired, but occasionally it is heated by means of coke. Cast iron or refractory pots may be used. The stationary furnace is usually employed for holding melts transferred to it from a tilt or reverberatory furnace. The melts are ladled from them and, therefore, such furnaces are commonly used in permanent-mold operations.

Fuel-Fired, Tilting, Pot-Type Furnaces — Tilting furnaces are either gas or oil fired and may have capacities up to 3000 lb. Cast iron pots or refractory crucibles may be employed. This is the most widely used type of furnace in aluminum alloy foundries. The melts are transferred to the pouring ladle by tilting the furnace, an operation which should be done carefully to minimize turbulence and the ill effects of the liquid cascading into the pouring ladle. If this is done improperly, gas can be absorbed and a large amount of dross formed as a result of the violent turbulence of the melt.

Reverberatory Furnaces — Reverberatory furnaces have capacities from a few hundred to 100,000 lb, and are usually coke, oil, or gas fired. The melts are contained in the shallow basin immediately below the flame. They are primarily used for the production of large quantities of metal, particularly when melting scrap for the production of remelting ingots, and for

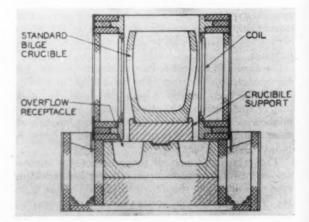


Fig. 2—This sketch shows a cross-section of a high-frequency furnace of the lift-coil type. A standard crucible is placed on the stand and the induction coil is placed around it. After the melt has been made, the coil is lifted off and the melt poured. Furnaces of this type, however, are not common in light-alloy foundries.

all

11

11

Tie

a.

he

a.

it

re

re

18

e

11

d

of g

casting into ingots for the manufacture of wrought products.

Small reverberatory furnaces with several dipping wells are frequently used for permanent-mold work. They are also sometimes used for melting metal for sand foundries, the melt being used directly from the furnace or transferred to a holding furnace for degassing and other treatments prior to casting. Reverberatory furnaces used for this purpose have a charging well at one end where charging, fluxing, degassing, and skimming are done, preparatory to pouring directly into the molds. The thermocouples placed in the dipping wells are connected with the temperature control equipment. Pre-alloyed ingots and scrap are charged into the furnace more or less continuously and metal removed from the dipping wells continuously for the casting operations.

Because the conditions in the open hearth are likely to produce large quantities of dross and high gas absorption, it is important that the flame not be directed onto the surface of the bath, where it can cause considerable turbulence and consequent formation of dross with gas absorption.

The materials used in the construction of the hearth are relatively important. Usually, common firebrick or high-alumina or chrome brick are employed. Bonding mortar of a calcined diaspore-base high-tempera-



Fig. 4—A typical lift-out or pit-type stationary furnace for melting aluminum. The flue gases pass out over the top of the crucible through the hole in the cover. The melting conditions approach the direct-flame type of operation.

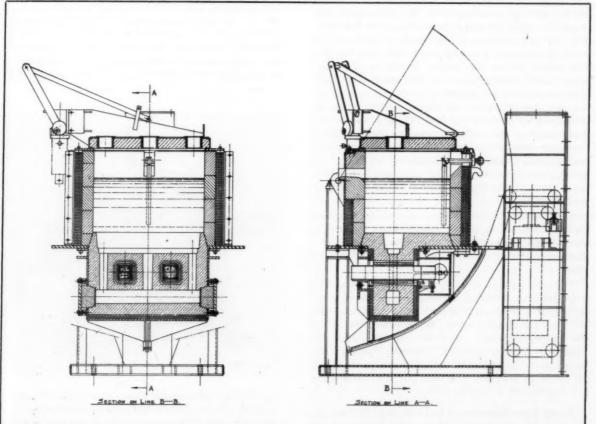


Fig. 3—This drawing shows two sections of a low-fre quency, twin-coil, 125-kw lip axis, tilting furnace for melting aluminum.

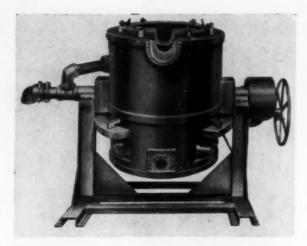


Fig. 5—A typical hand-tilting furnace for melting aluminum. The furnace is equipped with a cast iron pot. Though some flue gases escape from the opening under the lip of the pot, the major portion of the flame discharges through a port in the back and provides an indirect-flame melting operation.

ture cement or a high-alumina cement may be used. Zirconium silicate mortar is reported to provide long life by reducing penetration of the joints and, consequently, helps to avoid difficulties with buckling. The use of zirconium silicate brick in place of the firebrick produces a satisfactory hearth of long life. The initial cost of the "zircon" hearth is, of course, high. Local concentrations of high magnesium may, however, cause some difficulties with this type of refractory. Several courses of silicon carbide brick are recommended at the metal line. These bricks reduce dross build-up and make it easier to chip the dross from the refractory. Similar brick between the wells and heating chamber are advantageous because of improved heat transfer. The furnace design must allow for expansion and growth of the silicon carbide refractory.

A hard-fired chrome brick is also sometimes used for the preparation of the reverberatory hearth. A chrome-base, high-temperature cement is used as a bonding material when a high-chrome brick is employed.

The two-chamber reverberatory is made by adding a second furnace hearth forming a separate chamber with a refractory wall between and each with its own burner. The metal is melted down in one end, called the melting chamber. After melting, the major portion of the dross is removed. The metal is then transferred to the other end by means of a ladle or, better, by means of a siphon. To facilitate the transfer, the holding hearth is usually lower.

Barrel-Type Furnaces—The barrel-type furnace, gas or oil fired, commonly used in brass melting is not used extensively in the United States for melting aluminum, though they receive considerable favor in England, especially in large sizes. These furnaces rest on trunnions, making it possible to rock or rotate the furnace slowly and to tilt it for discharging the molten metal. Furnaces of this type are reported to have low thermal efficiency, to cause local overheating of the melt, and to produce a large quantity of dross and, perhaps, an unusual number of suspended dross particles, which occur as hard spots in the casting. Some of the large furnaces of this type built in England are reported to have quite satisfactory thermal efficiency.

Tilting Reverberatory — The small reverberatory furnace is mounted on trunnions for tilting, with one or more oil or gas burners at one end and a charging hopper at the other through which the products of combustion escape. The charge is preheated by the flue gases escaping through the hopper. This would be advantageous in that the adsorbed and chemically combined gases on the surface of the charge would be fairly completely eliminated by this preheating process. This feature also produces excellent thermal efficiencies and rapid melting rates. For example, a furnace holding 400 lb of aluminum will require about 20 min for the melting operation. Low maintenance costs are also claimed.

The metal is in intimate contact with the products of combustion and, as a result, excessive dross and gas absorption might be expected. As in the barrel-type furnaces, the short time required to heat the melt to the pouring temperature and the thorough preheating of the ingot and scrap before it contacts the melt surface may offset this apparent disadvantage. Some foundries report that good-quality melts are produced with relatively small metal losses. It is



Fig. 6—This photograph shows a hand-tilted crucible furnace. The crucible is mounted as illustrated by the section Fig. 11. The flame and flue gases pass over the top of the crucible and discharge through the center hole in the cover. These melting conditions approach the direct-flame type of melting.

te

he

to

al

ry

ne

of

ld

al

a

re

n.

ts

d

ts

ts

claimed that approximately 1 gal of fuel oil or 135 cu ft of natural gas (1000 Btu) are required to melt 100 lb of aluminum in these furnaces.

Electric Furnaces—Holding furnaces of both pot and hearth types equipped with electric heating elements have been used to a limited extent. When using the electric-resistance-heated pot or crucible furnaces, the melts are usually prepared in oil- or gas-fired tilt furnaces or in a reverberatory, then transferred to the electric holding furnace. The advantages of the electric furnace are the absence of products of combustion from gas or oil, which might cause gas absorption, and the good temperature control which can be obtained, as for example, in permanent-mold foundries. Good temperature control can, however, be obtained in other types of furnaces also.

Low-Frequency Induction Electric Furnace-In recent years, the low-frequency induction furnace is being used in a considerable number of melting rooms for the recovery of fine scrap, for the preparation of foundry melts, and as a holding furnace for melts prepared in other furnaces. These furnaces are produced in sizes ranging from 60 to 500 kw, and have pouring capacities of 200 to 5,000 lb of aluminum. A holding furnace of 20-kw capacity holding 600 lb of metal is available. These furnaces operate with a fairly low power factor and, therefore, require static capacitors in the circuit. They also operate with a thermal efficiency of about 70 per cent on a 24-hr basis and 50 per cent on an 8-hr basis. On a 24-hr per day basis of operation, about 0.2 kw-hr per lb of aluminum is required for melting and raising the melt to the pouring temperature. The low-frequency furnace is also built with two connected chambers for continuous operation. Melting is accomplished in one side and pouring from the other chamber. Good temperature control is reported to be obtained on the pouring chamber side.

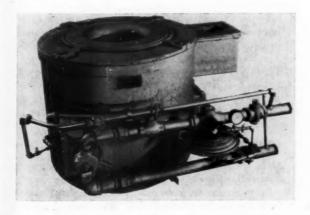


Fig. 7—This stationary furnace for melting aluminum is equipped with a refractory crucible mounted as illustrated by Fig. 12. The flame and products of combustion escape primarily through the flue at the back. This design of furnace provides a close approach to indirect-flame type of melting.

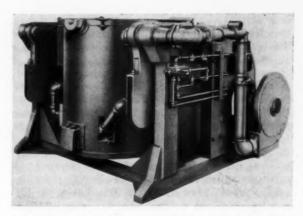


Fig. 8—This photograph shows a large motorized lip axis tilting furnace for melting aluminum. The crucible is installed as illustrated by the sketch, Fig. 12. The flue gases escape through a port in the back. The melting conditions approach that of indirect-flame type of melting.

The furnaces operate from normal-frequency power sources and are controlled with voltage-regulating transformers and contactors.

When starting a new lining, it is heated to a bright red heat in the lower channels with a gas torch or a strip resistance heater, and molten metal is then added to fill the channel. It is, of course, necessary to maintain the liquid metal in the channels during the down period. It is also necessary to empty the channels and clean them out about once a week under full operation, or less frequently when the melting operation is not up to full capacity. The frequency of cleaning depends somewhat upon the composition of the alloy being melted.

One foundry reports that in 1946 a 150-kw lowfrequency induction furnace cost \$15,000 installed, whereas a 1000-lb gas-fired tilt crucible furnace cost \$3875 installed. At 0.7 cent per kw-hr, the power cost for the low-frequency furnace was \$0.14 per 100 lb of metal melted. In the same foundry where gas cost \$0.50 per 1000 cu ft, the fuel cost in the gas-fired tilting furnace was \$0.15 per 100 lb. No comparable fuel costs for other types of furnaces are available, but fuel costs for reverberatory furnaces are approximately 50 per cent lower. No cost figures were given on maintenance. However, an average of 11/2 million pounds of aluminum can be melted with one lining in the low-frequency induction furnace. In all probability, the costs of maintenance would be in favor of the induction furnace. Likewise, reduced metal losses, improved melt quality, and better working conditions would favor the induction furnace.

Another characteristic of the induction furnace, low or high frequency, is the automatic stirring of the melt by the induced currents. This insures melts of uniform composition and greatly facilitates the preparation of alloys.

High-Frequency Induction Furnace—The high-frequency furnace in the larger sizes requires a motor-

generator set as a source of power. The first cost of the high-frequency installation is high. Primarily for this reason, they are not extensively used for melting aluminum alloys. The general melting efficiency is about 60 per cent, which is 10 per cent higher than that of the low-frequency furnace based on an 8-hr day. Unlike the low-frequency furnace, the entire melt can be poured from a high-frequency furnace, thereby making it more flexible in the foundry where it is desired to melt 1 lb of aluminum and heat it to the pouring temperature. The furnaces may be the tilt type, having capacities of from 20 to about 300 lb of aluminum, or lift-coil type, having capacities up to about 90 lb of aluminum. When using the liftcoil type, the coil is lifted after the melt has been made, and the melt in the crucible is carried to the mold and poured. This represents extreme flexibility in respect to handling a variety of alloys, avoids possible metal damage during transfer, and eliminates ladles and ladle heaters.

Because of the high thermal efficiencies of the highor low-frequency induction furnaces, the greater reliability of electric power, and the recent considerable advances in the price of fuel oil, the cost per pound of metal melted in the induction furnaces compares favorably with that produced by oil- or gasfired furnaces. Excepting for the high first costs, the high-frequency induction furnace is an ideal melting unit for almost all types of materials.

Selection of Pots and Crucibles

The most common crucible and pot materials for aluminum melting are cast iron, clay graphite, and sil:con carbide usually bonded with carbon. The refractory crucibles for the stationary and tilt furnaces must be carefully installed with proper support if they are to provide long service, and the flame should not impinge directly on the crucible. It is also especially important that an oxidizing flame does not impinge upon or even be used for heating silicon carbide crucibles. If this is done, the crucible will have a relatively short life. The silicon carbide crucible has the advantage over clay graphite in that its thermal conductivity is higher and heat transfer is, therefore, more rapid. Considerable care must be exercised when using silicon carbide crucibles that pieces broken from them do not form hard spots in the castings. The clay-graphite crucibles are the easiest to clean out after the melting operation. All crucibles and pots, while hot enough to keep the adhering metal molten, should be scraped down with a blunt, rounded steel blade after the melting cycle has been completed. It is important, however, that refractory crucibles not be damaged in this operation. If this cleaning is done when the crucibles are still red hot, the dross is more readily removed and the refractory crucibles are less likely to be damaged.

Serious contamination of the aluminum melt with iron will be caused unless the iron crucibles are cleaned and coated with refractory wash. For this reason, low-iron alloys such as those containing silicon and magnesium as the principal alloying elements

should be melted in a refractory crucible rather than in an iron pot. In any case, the iron pot must be scraped down while it is hot and the interior coated with a thin slurry of red mud, an insulating material produced from bauxite, or with a wash of whiting, talc, or finely ground mica, with an addition of sodium silicate to promote adherence. The pots are heated to a temperature slightly above 212 F before putting the wash on with a brush or spray. It is important that the refractory wash not be a hard scaly type; otherwise, if it were to flake off, it could be a serious source of hard spots in the casting. It is important that such washes be thoroughly dried before molten metal is transferred to them when the pot is being used in a holding furnace. In such cases, the cast iron pot should be heated to a low red heat to drive off any moisture in the sodium silicate, clay, or other moisture-containing compounds in the wash. If this is not done, the melt may be very seriously gassed, and there is some possibility of an explosion from the rapid evolution of steam and its reaction with the melt.

When using lift-out or pit-type crucible furnaces, refractory crucibles are employed and they are lifted out and carried to the crucible in a shank, an operation which must be done carefully with properly fitted tongs and shanks to avoid damage or possible failure of the crucible.

Tools and Their Maintenance

Furnace room tools such as stirrers, phosphorizers for adding volatile materials, skimmers, hand ladles, and fluxing tubes for degassing with inert gases are usually made of steel. It is very important that these tools be kept clean and properly coated with a refractory wash. They should be thoroughly preheated to a red heat or nearly so before inserting them into the melt. Otherwise, they may be a source of hydrogen absorption and a hazard to the workmen. Graphite tools are sometimes employed and, though they are more fragile than iron or steel, are not sources of metallic contamination. Such tools must also be carefully preheated to a red heat before they are inserted into the melt. This is especially true of carbon tubes and other devices used for introducing chlorine or halide fluxing compounds into the melt. The reason is that such devices invariably are covered with hygroscopic compounds. These compounds, present from previous use, must be thoroughly dehydrated at a high temperature.

Charging and Melting

After the new furnace linings or new refractory crucibles have been carefully and slowly preheated, the furnace may be charged. Furnaces which have been in use can be charged cold. They should not be packed tightly, however, because expansion of the charge will tend to crack the crucible. Bridging may also occur, causing overheating of the metal in the bottom. After the furnace has been charged and melting started, more charge should be added on top as the solid material in the crucible settles into the melting zone formed by the molten metal. This will in-

115

an

be

ed

ial

ng,

201-

ire

Me

m-

ilv

23

m-

re

is

he

to

or

If

lv

111

m

d

0

E

sure preheating the ingot and scrap before it is introduced below the melt surface. This preheating is an important feature because it helps to eliminate the adsorbed and chemically combined gases, particularly moisture, on the ingot surface, as described previously.

The charge should consist of clean foundry operating scrap of the alloy being melted and pre-alloyed ingot. The pre-alloyed ingot is usually prepared in reverberatory furnaces and the analyses determined, or ingots of the desired composition can be purchased from producers of remelting ingots. Some foundries prefer to melt their foundry scrap separately in large tilt furnaces, or, in larger operations, in a reverberatory furnace, analyze it, adjust the composition if necessary, and then pig it off for remelting in the foundry.

In general, it is not desirable to prepare foundry melts from charges consisting, in part, of unalloyed virgin materials, unless extreme care is exercised in obtaining the weight and correct composition of all the materials charged. There is some merit, however, in preparing aluminum alloys containing high contents of magnesium in the furnaces without pre-alloying. The reason is that considerably less dross is formed, and the use of cover fluxes is minimized or eliminated.

When it is necessary to add alloying elements to the foundry melts of aluminum, the low-melting ones, such as zinc and magnesium, can be added in the form of the pure metal. However, the higher melting point metals, such as manganese, nickel, silicon, titanium, chromium, and copper, should be added in the form of a master alloy of known composition. Silicon and copper can be added to the charge without resorting to a master alloy. However, their solution in the melt is relatively slow and considerable care must be exercised if this is done. Silicon tends to float on the surface of the melt and becomes oxidized, which greatly reduces its rate of solution. Magnesium also floats and must be submerged into the melt immediately to prevent the formation of spinel particles and reduce oxidation and metal loss.

Reverberatory furnaces may be operated as batch furnaces or semi-continuous or continuous melting units. When they are used at a batch furnace, all or most of the metal is drained from the hearth after each melting operation. In the semicontinuous method, the melt is removed in batches and additional charges are added and melted between taps. In the continuous method, metal is removed continuously from the open hearth at one end while charges are being added at the other. This is best done in a double-chamber furnace in which the charge is melted in one chamber and the metal is tapped continuously from the other.

Preparation of Remelting Ingot

The usual purpose of the reverberatory is to prepare remelting ingots of various aluminum-base alloys. When this is done the available "heavy" scrap and virgin aluminum, constituting about one-half of the charge, are placed in the furnace and melted.

After it is melted, the other half of the charge is added, melted down, and the bath skimmed to remove dirt and oxide. Continuous skimming should be avoided because it increases metal loss and promotes gas absorption. If a large amount of wet dross is formed, its removal may be facilitated by treating the melt with a solid or gaseous halide flux or chlorine, as described elsewhere in this paper. The alloy additions are made as described previously. The metal bath is then rabbled carefully to avoid undue disturbance of the melt surface, the purpose being to produce a melt of uniform composition. A chemical analysis is then made, and, if the results are satisfactory, the metal may be tapped and poured into remelting ingots. If necessary, the melt may again be fluxed with chlorine or anhydrous aluminum chloride before tapping.

The taphole may be opened by drilling through a sand or fireclay plug in the furnace wall, and permitting the metal to pour out freely into the casting ladle. Some furnace operators employ a tapered plug or bar machined to fit into the pouring sleeve in the furnace wall in order that the rate of discharge can be controlled.

The melting of wrecked aircraft parts, an important source of scrap metal during and after war, requires reverberatory furnaces with large doors to accommodate large pieces of scrap. Heavy metal inserts are recovered by melting the scrap on a sloping hearth, the furnace being equipped in some instances with charging doors on the roof. When the heavy metal inserts are free, they are raked from the sloping hearth, which minimizes contamination of the aluminum melt.

It is necessary to stir the melt from the bottom upwards to produce a uniform melt composition. This is necessary when using either reverberatory or pottype furnaces, but the surface of the melt should not be broken. Contrary to some claims, it is not likely that such stirring and rabbling will aid in the removal of dross, but it will insure uniform composition of the melt.

Some foundrymen advocate slightly oxidizing flames, while others prefer reducing flames for melting aluminum. There seems to be no reason for either preference, other than consideration of fuel economy. There is some evidence, however, that slightly reducing flames will provide less metal loss during reverberatory melting of scrap, but may promote greater gas absorption.

Some melting-room operators prefer to drain, clean, chip off the dross, and repair the reverbatory furnace after each week of operation. Others recommend that this be done after each heat to reduce difficulties with "hard spots."

Effect of Temperature and Holding Time

During the preparation of melts for the foundry, it is extremely important that the time at which the melt is molten and the maximum temperature both be kept to a minimum. In general, gating practices should be such that low pouring temperatures can be used. With the low pouring temperatures, low maximum temperatures can be utilized, and this is

of some importance in the production of melts of low gas content. Excessive melt temperatures and excessive holding periods also coarsen the grains, which tends to produce low mechanical properties in the castings. In general, melts which are prepared for pouring at high temperatures should be used only for those castings requiring a high pouring temperature. Melts subjected to high maximum temperatures should not be cast into molds that can be poured sections, even though the melt is cooled to the proper at low temperatures, especially castings having heavy pouring temperature. If melts subjected to high maximum temperature are used, the castings with heavy sections will be unnecessarily coarse grained and unsound with respect to pinhole porosity.

Because overheating is so harmful to the quality of the aluminum-base melt, it should be emphasized that the charge tends to continue to increase in temperature after the burner has been shut off. This becomes more evident when it is realized that only 17 per cent of the total heat is required to raise the charge from room temperature to the melting point; 82 per cent is required to melt it, namely, to raise it from the solidus temperature to the liquidus temperature, but only 1 per cent is required to raise the melt 50 F. It is possible to equip the melting room with automatic signals which will warn the operator when the charge has been melted.

Temperature Control and Temperature Measurement

If a melt has been overheated or held for a long time in a molten condition, it is best to pig it off and remelt it. Melt temperatures in the furnace are usually measured by means of a closed-end No. 8 gage chromel-alumel thermocouple separated by double-hole porcelain insulators in a closed-end cast-iron protection tube, 1 in. i.d., 1½ in. o.d., and of suitable length to extend deep into the metal bath. The couple is connected by means of suitable compensating leads to a millivolt meter where the temperature is read directly. To prevent iron absorption from the

protection tube and also to increase tube life, it should be kept clean and coated with a lime slurry or other refractory wash, similar to that described for cast iron pots.

The measurement of the melt temperature by a more rapid-acting thermocouple is required for pouring ladles. Open-end and closed-end thermocouples are used. No. 8 to No. 14 gage chromel-alumel wires wrapped tightly with asbestos to within 1/2 in. of the lower open end are submerged into the melt. This is a rapid acting couple, but is not so reliable as a closed-end couple. Closed-end couples for rapid reading should be of light gage, and they may be the stationary type for measuring temperature at a fixed location in the foundry or the portable type for use on the pouring floor. Thin-wall iron or silica protection tubes are sometimes used on couples of this type. As a matter of fact, the measurements of the ladle temperatures are not too satisfactorily done because the employment of a protection tube makes the action of the couple too slow, whereas both open and closed unprotected couples require a great deal of maintenance because they are rapidly attacked by the molten aluminum. The burned asbestos of openend couples is removed from the bottom, and the ends of the wires attacked by the aluminum are cut off. The damaged part of closed-end chromel-alumel couples is cut off and the ends re-welded. Some couples are made with replaceable tips, thus simplifying their maintenance.

A thermocouple can be checked readily by placing it in a closed-end, thin-wall carbon tube and immersing it in a small pot of pure aluminum which is allowed to solidify. The temperature will drop continuously until the freezing point is reached, at which it will remain constant for a considerable period of time. This is the melting point of aluminum and corresponds to about 1218 F. Careful and frequent checking of the couple and millivoltmeter are necessary to avoid serious errors in the measurement of pouring temperatures. Portable instruments especially require frequent checking.

PART II - MELTING PRACTICE FOR MAGNESIUM - BASE ALLOYS

Charging and Melting

It should not usually be necessary for the foundry to prepare alloys from virgin or unalloyed materials. Primary ingot is made by the ingot producers by large-scale open-pot melting methods and in the reverberatory furnace, described below. The flux-free metal of the desired composition is pumped from these furnaces to the ingot molds. The foundryman, then, is confronted only with the problem of remelting the primary alloy ingot, purchased scrap, and his foundry scrap. If aluminum or zinc is added to the melt, it should be done in an alloying basket or hand puddler. Very fine scrap such as machine shop borings and turnings can be recovered by special melting prac-

tice, but such fine scrap should not be included in foundry melts. Preparatory to remelting, fine scrap should be stored in covered metal drums, because it is a fire hazard. Magnesium melts are prepared by the open-pot method, crucible process, reverberatory melting, and a special melting process for die casting.

It is important that all of the materials charged into the furnace be clean and of known composition. Foundry scrap must be properly identified at the place of its origin and thoroughly cleaned of all adhering sand, if necessary, by blasting. The charges for the melting furnace are weighed, recorded, and taken to the furnace in steel tote boxes. The charge is a balance of operating scrap, primary ingot, remelting ingot, reclaimed scrap, and purchased scrap known to

be of correct analysis.

A superficial coating of dirt and Mg(OH)2 forms on common foundry alloys of magnesium, and, therefore foundry scrap should not be accumulated in the foundry but should be melted as soon as possible, the oldest scrap first. For the same reason, primary ingot should be carefully stored to minimize the formation of corrosion product on the surface. However, even with good metal-handling procedures, it is essential that all of the charge be thoroughly preheated before it is introduced into the magnesium melt. This can best be done by charging the metal on top as it descends into the bath below, and by preheating the ingots on top of the furnace. In some cases, preheating the ingots in a separate furnace is necessary. To fail to take these precautions will not only produce melts of high gas content, but personnel will be endangered because of the possibility of metal spattering. It is highly desirable for workmen to wear goggles or face shields and clothing fire-proofed with a suitable compound such as ammonium sulfamate.

Crystalline magnesium produced by the ferrosilicon process can be melted in a manner similar to ingots produced from metal made in electrolytic cells. The so-called crystalline magnesium, however, may contain some oxides and nitrides in the form of solid particles or films. Thus, these compounds are mechanically mixed with or form films on the magnesium crystals but are not dissolved in them. The magnesium crystals also present greater surface area per unit of volume of metal. The problem of melting this material, then, is somewhat similar to melting fine



Fig. 9—This photograph shows the common, stationary furnace for melting magnesium. The steel crucible is equipped with an ear at the top which rests on the cast iron furnace cover ring. Some of the flue gases escape around the rim of the crucible, but the bulk of them escape through a port in the back. Again, this design of furnace produces a close approach to the indirect-flame type of melting.

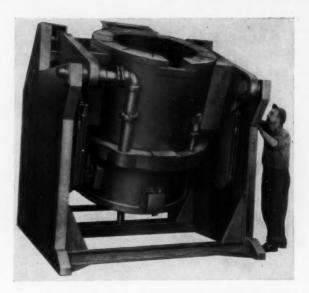


Fig. 10—This photograph shows a large motorized lip axis tilting furnace for melting magnesium. Large cast or welded steel crucibles with ears on top are supported by the cast iron flange at the top of this furnace. The flue gases escape through a port in the back and the melting conditions are essentially those of indirect-flame melting.

scrap. The magnesium crystals are submerged into a molten heel and greater amounts of melting flux of the Dow 230 type are necessary.

The Open-Pot Method

This method of melting has very wide applications in the production of primary ingot for fabrication into wrought form or for remelting in the foundry. It has relatively limited application in foundries, excepting where it is necessary to transfer the metal by means of hand ladles, as in permanent-mold work and certain small sand foundry operations. It is also used for the recovery of fine scrap. It should be pointed out that some British foundrymen prefer to prepare melts for permanent-mold casting by the crucible melting process. This is done in smoothly operating tilt furnaces and usually the melt is protected by a thick flux of the Dow 310 type or by an SO₂-filled dome placed over the melt.

When using the open-pot method of melting, a fluid flux, such as Dow 230, is placed in the melting pot which is heated to about 500 F. As soon as the flux begins to melt, preheated alloy ingots and clean foundry scrap are charged into the furnace and melted. Additional flux is sprinkled over the charge to prevent burning. When the entire charge is melted, the melt is skimmed, and fresh flux added, which is allowed to melt and then stirred into the molten metal. This refining operation is supposed to cause suspended dross particles in the melt to separate out. It should be noted that the flux, even after it has melted, contains some moisture which will react with the metal, and thereby be a source of gas. Recent

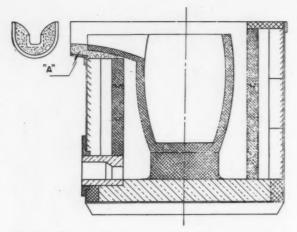


Fig. 11—This sketch shows the method of mounting refractory crucibles in tilting furnace in which, essentially, open-flame melting conditions are encountered.

work in England has shown that the inclusion of oxides and flux found in castings poured from properly prepared melts does not arise from these nonmetallic materials suspended in the body of the molten metal prior to pouring, but such defects arise at the time of casting the metal into the mold. They further con-

cluded that the principal function of the flux is to provide a cover having suitable physical properties to provide protection on the one hand, and yet permit the metal to be poured without any flux entering the mold on the other.

Crucible Melting

Crucible melting is the most commonly used method in sand foundries. The melts may be made in relatively small steel crucibles and the melts poured from them into the molds, or the melts may be made in a much larger tilting furnace and the metal then transferred to the smaller stationary furnaces for refining, after which they are carried to the pouring floor and the melts poured into the molds.

When preparing the melts by this method, the steel crucible is heated to about 500 F and a small amount of cover flux, such as Dow 310, is dusted on the crucible so that it adheres to the side wall and the bottom. Small pieces of magnesium are then placed on the bottom of the hot crucible, and larger pieces of clean foundry scrap and pre-alloyed ingots are added. Small additions of the flux are dusted on to prevent burning as the metal melts. When the crucible is filled, the melt is kept covered with a thin layer of the flux to prevent burning. This flux tends to become dry as it stands on the melt, and additional flux may be necessary from time to time to provide com-

is

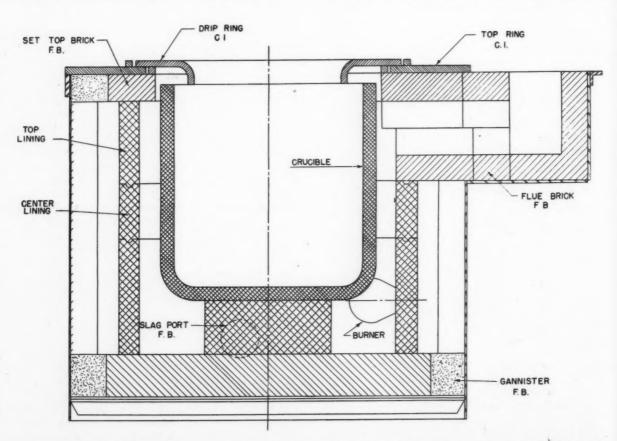


Fig. 12-This sketch shows the method of mounting crucibles and the design of the top of the furnace in

which the melting conditions are essentially those of indirect-flame operation.

plete protection.

it

d

n

If the metal is prepared in a large tilt furnace, the same procedure is used and the metal heated to 1320 to 1400 F. The Dow 230 type flux is usually used instead of 310 if a lower temperature is used, as the manganese tends to separate out, and if higher ones are used, difficulties with burning are encountered. The melt within this temperature range is stirred with an upward rolling motion, spectrographic analysis obtained, and the melt transferred to the several small, clean, preheated pots and taken to the stationary furnaces. Some foundrymen hold the melt at 1220 to 1260 F and, while waiting for analysis, flux it with oil-pumped nitrogen for 30 minutes. The melt is then raised to 1300 to 1320 F and then transferred. The walls of the large pot are scraped down, sludge removed, and the melt repeated.

A sprinkling of cover flux of the Dow 310 type may be necessary to prevent burning in the holding furnaces. With either method, the melt in the small stationary furnaces is then ready for purification and for treatments to produce grain refinement.

Metal Handling and Pouring

After grain refinement, the metal is cooled or heated to a temperature found to be satisfactory for the job to be poured, taking into consideration the drop in temperature during the transfer to the mold. Upon removal from the furnace setting, the crucible is first taken to the pyrometer station, where it is placed in the pouring shank and all of the crusted flux removed. This operation must be performed very carefully, being sure that all of the flux on the crucible walls above the metal line is scraped clean and that all loose flux adhering to the crucible supporting ears is removed. Special attention must be given to cleaning the pouring lip. The metal is protected from burning by dusting the exposed surface with a mixture of sulphur and boric acid after the flux is removed.

When calcium is to be added to the melt, an addition of 0.05 per cent is made at the pyrometer station. Calcium must be stored under kerosene until just before it is ready to be used, when it must be dried carefully and one corner filed down to clean metal. After the calcium addition, the metal must be poured without delay because the calcium burns out of the metal very quickly.

After the flux has been removed and the calcium addition made where required, and while the metal is still above the required pouring temperature, the crucible is taken to the mold. The final determination of the temperature of the metal at the time of pouring is taken at the mold with portable pyrometers.

Pouring temperature should be set for each casting at the same time the gating is established. The proper pouring temperature must be closely controlled to prevent misruns or porous castings.

Pouring of molds is done whenever possible with one crucible. However, the best gating procedure for some castings requires that two crucibles be poured into the mold simultaneously. A pouring box is usually advisable when pouring magnesium alloys. The metal must never be poured directly down the sprue but the stream directed to the side of the basin away from the sprue. The metal must be poured rapidly at first to fill the pouring box, and the rate then adjusted to keep the pouring box full until the pour is completed. The pouring rate must be maintained as smooth as possible to prevent breaking the oxide skin surrounding the metal stream. The size of the pouring box is determined by the size of the crucible being used. The smaller boxes are usually of cast iron, and the larger ones of rammed green sand held in a steel box or of baked core sand.

The lip of the crucible must be held as close as possible to the pouring box to prevent excess turbulence and oxide formation.

During the pour, one man is stationed at the mold to push back any oxide or flux particles on the metal surface in the crucible and to prevent those nonmetallics from entering the mold with the metal stream. This man also dusts the mixture of sulphur and boric acid on the exposed metal to prevent burning.

All of the metal in a crucible should not be poured into molds, but a heel of 10 to 15 per cent left in the bottom of the crucible to prevent flux in the bottom of the crucible entering the mold.

After the pour, the heel is pigged and later charged to a remelt furnace. An alternate procedure is to empty these heels from several crucibles into one and pour this metal directly into a remelt furnace without pigging.

The walls and bottom of the pouring crucible must be scraped clean before it is used for another melt.

Melting in a Reverberatory Furnace

Melting in a reverberatory furnace is carried out in a manner quite similar to the crucible method of melting. No fine scrap should be employed because of the difficulties with dross formation. The clean foundry scrap and primary alloy ingot are charged into one end of the furnace on a preheating shelf from which they are pushed when adding the next charge. Alloy additions, aluminum, zinc, and manganese chloride are charged in a well at the end opposite that from which the alloy is pumped. Melting flux of the Dow 230 type is used as a cover during the melting operation to prevent burning. The melt is ladled or pumped from the furnace to the stationary furnaces where the metal is cleaned, degassed, and subjected to grain-refining procedures. It is also possible to degas the metal by chlorine fluxing in a manner similar to. that used in reverberatory melts of aluminum-base alloys. To insure freedom from the 230 flux, the melt is pumped to a settling pot before transferring it to the molds. The reverberatory furnace is particularly useful for the production of alloys cast into remelting ingot, or for the production of large quantities of foundry melts. Melting rates of 5000 lb per hour are reported.

Temperature Control and Measurement

The problem of temperature measurement in the

magnesium foundry is much simpler than that in the aluminum foundry because closed-end chromel-alumel thermocouples with a plain-carbon or high-chromium heat-resisting steel protection tube can be used. Another couple which is quite successful consists of a single insulated wire of constantan within and attached to the bottom of a closed-end iron pipe. Because this is a rapid-acting couple, it is very useful for the measurement of pouring ladle temperatures with either stationary or portable equipment.

DISCUSSION

Chairman: J. J. WARGA, Piasecki Helicopter Corp., Morton,

Co-Chairman: P. J. Scherbner, Sperry Gyroscope Corp., Great Neck, Long Island, N. Y.

MARSHALL STEVENS: 1 What effect does fluxing with dry nitrogen have on 220 alloy in regard to formation of dross, gas holes, or porosity?

DR. EASTWOOD: If we use nitrogen for fluxing 220 alloy containing about 10 per cent Mg, the results are less consistently good than if we use chlorine. Nitrogen does not have the fluxing action, and therefore is not so likely to remove dross from the metal. If the nitrogen is not quite pure it might actually be a

means of introducing dross into the metal. My experience with nitrogen for fluxing 220 alloy has shown that it generally is not beneficial. On the other hand, chlorine usually is beneficial if it is carefully done and care is exercised that the fluxing system does not become contaminated.

MR. STEVENS: What is the next best material other than chlorine?

Dr. Eastwood: As far as I know there is no second best material for fluxing the 220-type alloy.

D. L. LAVELLE: ² When discussing the use of chlorine, you mentioned that it is not suitable for aluminum-silicon alloys. Will you elaborate on that?

DR. EASTWOOD: The difficulty with using chlorine with aluminum-silicon alloys is that the chlorine tends to remove the sodium. Aluminum-silicon alloys require small amounts of sodium to produce silicon particles having a desirable shape. If there is some sodium in the melt, the silicon particles will approach a spherical shape. If sodium is absent in the melt, silicon will form a plate-like structure which reduces the ductility of the casting. Chlorine removes the sodium to a greater extent than nitrogen. For that reason, many foundrymen prefer to use nitrogen rather than chlorine for fluxing aluminum-silicon alloys. If chlorine is used, usually some method must be employed to restore the sodium content.

P

F re ti d fe

p

le

51

¹ Metallurgist, Peerless Aluminum Foundry Co., Inc., Bridgeport, Conn.

² Research Metallurgist, American Smelting and Refining Co., Barber, N. J.

RADIOGRAPHY AS AN ASSISTANT TO FOUNDRY PRACTICE

By

Charles B. Johnson* and Stanley A. Brosky**

EARLY IN 1944, the production of non-ferrous pressure-tight castings, from a quality standpoint, apparently reached an all time low. Raw materials, especially virgin metal, were difficult to secure. Supplying foundries were enjoying a peak production of castings which did not have to withstand pressure and could be produced from the materials available. For our castings the attempted usage of this material resulted in an abnormal percentage of rejections, at times this percentage reached 100 per cent for individual shipments. Apparently sound castings were found faulty when subjected to final tests. The nature of the product was such that it was almost impossible to run preliminary tests because leakage was encountered only after machining through surfaces of which one side had been machined and the so-called protective skin removed.

Figure 1 shows an assortment of valve bodies and plugs as they are received from the foundry and sent to our laboratory for x-ray examination.

A rejection at the final test resulted in the loss of machine-hours. This was costly as was the loss of material, assembly and dis-assembly labor. These losses amounted to a considerable total, especially when the product was assembled on a production basis.

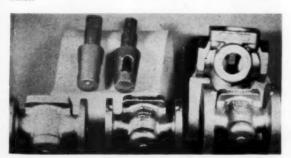
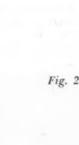


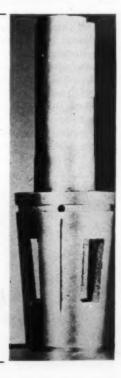
Fig. 1

Figure 2 shows a completely machined valve plug. You will note the numerous areas that are machined.

Castings that appeared sound after machining had to be rejected at final test because of leakage. The rejection of machined castings which appeared to be sound from a visual inspection caused continual controversy with the supplying foundries. This controversy reached its height particularly when rejections were made in the face of favorable hydraulic tests conducted by the supplying foundries. This seeming inconsistency was cleared up satisfactorily when it was pointed out that the finished product would be used sometimes on gases or violently active chemicals that would penetrate the metal through pores so fine that they could not be discovered by ordinary testing and visual inspection.

In order to alleviate these conditions and supply





*Chief Engineer, Rockwell Manufacturing Company, Pittsburgh.

**Department Head of Metals, Pittsburgh Testing Laboratory, Pittsburgh.

111

St

th ve of

ill

11. 11.

11

n

castings to the Manufacturing Section that would not leak after machining, the co-operation of the various supplying foundries was solicited. Arrangements were made to work directly with a captive foundry. The raw materials and the charging of the furnace was carried out under close supervision. Temperatures. both melting and pouring, were closely controlled. Various heats were purged, using different purging agents. Melts were made of both virgin metal and secondary ingot. A detailed log was made of every heat. This included chemical analysis of both raw materials and finished products, test bars, to determine physical properties, temperatures, both furnace and pouring, and the type of raw material used. Under these conditions we were able to increase the yield of usable castings, but not to the point where all casttings could be machined with the assurance that they would be pressure-tight when finally assembled.

We further developed the fact that by using secondary ingot satisfactory castings could be produced. Our experience was that the use of virgin materials increased our yield to some extent. It thus became evident that some method of non-destructive, pre-

machining inspection was necessary.

Several methods were investigated and the x-ray method was finally accepted as offering the most promise. With the use of this method for accepting or rejecting these castings, supplying foundries immediately raised violent objections. In an effort to provide a method or technique which would be infallible, we examined many castings by x-ray and then cut them up into wafers or slices, varying in thicknesses from $\frac{1}{16}$ in. to $\frac{1}{2}$ in. X-rays were made of the individual slices and also of the slices after they had been re-assembled in order to determine just what could be expected from the x-ray when used on castings of varying thicknesses. Many castings were machined, assembled and tested to substantiate the x-ray findings.

It was discovered early in the investigation that x-raying the entire casting would prove too costly. The critical areas for each type of casting were located by

machining, testing and x-raying again.

Figure 3 shows a cross-section of a typical valve body indicating the critical areas and where the x-ray radiograph is taken.



Fig. 3

At first the results obtained were not entirely satisfactory because the obvious defects were not the ones that were causing the so-called leakage or seepage.

As mentioned before, these castings that leaked after final machining showed no flaws or defects upon visual examination of the machined surface. When x-rays were taken of relatively small spots and the exposure was correct it was readily discerned that we had been overlooking a type of porosity which was causing our trouble. This porosity we have called microscopic. Particular care was given to various metals to see that the best technique was being used to bring out this peculiar structure.

In the case of the leaded bronze* a technique was set up that would most clearly show the porosity that

caused the leakage.

A typical exposure would be as follows:

Material: 3/4-in. Leaded Bronze*
Voltage: 215 kv
Current: 15 ma
Exposure Time: 1 min

Exposure Time: 1 min
Target to Film Distance: 36 in.

Filter Used: 0.015 in. of Copper Penetrameter: 2% of Thickness Type of Film: Eastman Kodak K

in

to

gr

fe

ag

ac

of

qu

ac

in

an

gr

in

The authors would like to point out that all their work using this technique is done with Eastman Kodak K film since it is very fast and the grain structure or size makes interpretation of radiographs of rough castings easier. We have found that the coarseness of the film grain helps eliminate the surface roughness of the casting on the radiograph.

We refer to our system of radiographing castings as the "single exposure technique" since in most instances a single radiograph of as little as 4 sq. in. of area will enable us to obtain sufficient information to accept or reject the castings in sizes up to and including 4 in. Such a casting will weigh approximately 60 lb.



Fig. 4

^{*}Typical analysis:—Cu, 88.0; Sn, 9.75 Pb, 2.00; Ni, nil; Impurities, 0.25.



Fig. 5

Above this size we x-ray one critical spot in the body of the casting and one such spot in the connecting flange area. This technique has proved satisfactory for use on castings weighing up to 4,000 lb.

Figure 4 is a photo showing the way a typical radiograph is taken and how the machine is focused.

Interpretation of the radiographs is somewhat different than that usually practiced with routine radiographs. Obvious defects, such as blow holes, shrinkage cracks, gas porosity, etc., are secondary in the acceptance or rejection of the casting. The first item of major importance is whether or not the metal quality is satisfactory. If the quality of the metal is acceptable, then the obvious defects play their part in determining the acceptance or rejection of the casting. Naturally, the size of the defects, their location and their dispersal also have a great bearing on this acceptance or rejection.

Figure 5 shows a radiographer interpreting radiographs.

Our interpretation of the negative must of necessity include the visual examination of the condition of the surface of the casting at the same time the radiograph is viewed on the view box.

Figure 6 shows how rough a casting might be when it is received at the laboratory for x-ray examination. We have found, for example, that castings that are



Fig. 6

smooth as far as outside appearance is concerned can be porous when internal structure is considered. Furthermore, castings that show a rough surface condition have been found in many instances to be sound and of good metal quality.

In general, the interpretation of the radiographs is based on the ability of the radiographer to see the fine mottled structure, at times interspersed with pinpointed dark areas which we have come to call "dross-like" porosity. We have used this term "dross-like" porosity to describe a condition that exists in the non-acceptable castings since it appears to be caused by a material similar to dross that has moved through or has been entrapped in the casting. In the past, this condition has been referred to as micro-shrinkage or micro-porosity. It has been found that castings showing this condition to a minor extent have been leakers under pressure testing, and, of course, are rejects after machining. Consequently, they are now





Fig. 7

Fig. 8

in su ex th

a fo

ce ce th min 500 pr chi sin jec th in th qu

ca

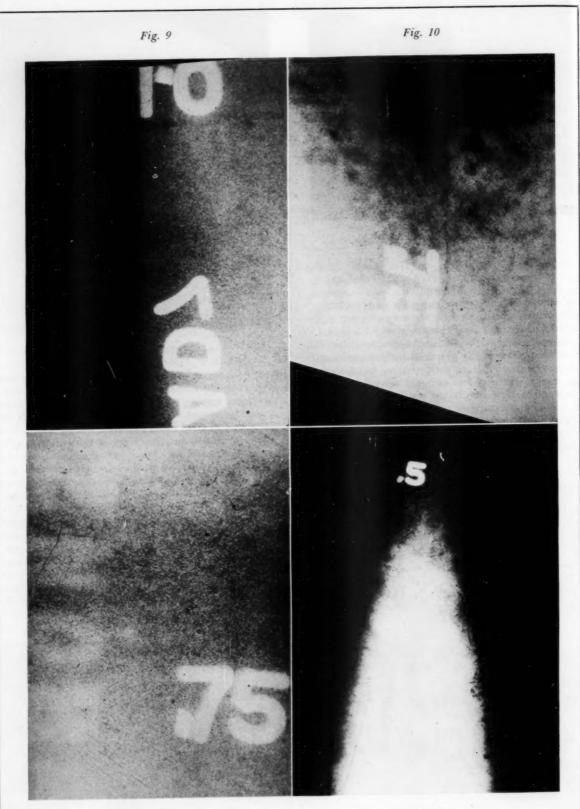


Fig. 11

Fig. 12

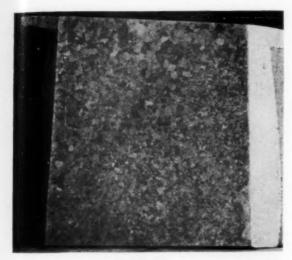


Fig. 13

rejected by the radiographer.

A system of numbering the castings was inaugurated in order that the castings could be checked and pressure tested to determine the accuracy of the x-ray examination. All castings were checked at double

the rated working pressure.

The results of this checking enabled us to set up a 3-way classification of castings as received from the foundries. These three classifications were (1) acceptable, (2) borderline acceptable, and (3) not acceptable. Acceptable castings (1) were those of which there was no doubt as to their being satisfactory after machining and assembly. Borderline acceptable castings (2) were those in which there was better than a 50 per cent chance that the castings would prove pressure-tight after machining. These did not include such defects as cracks, large blow holes or other similar defects for which castings are normally rejected. The not acceptable classification (3) included those in which the porosity was great enough to eliminate any possibility of the castings being usable. In this classification checking has proven that the metal quality of castings was poor.

Figure 7 shows a typical scene showing the classification of these castings. Figure 8 shows a group of castings that have been classified. You will note the paint color on the castings and the areas outlined

which had been radiographed.

Figures 9, 10, 11 and 12 are positive photographs of actual negatives of castings and of each of the three groups or classifications. Figure 12 shows a radiograph of a casting that contained many fine cracks and the mottled structure. This was the same casting that was used in making a macrograph, showing its internal structure.

After the radiographs have been interpreted, the castings are divided into the previously mentioned three classifications and stenciled with the same number as that identifying the film. These castings and films are used for further case history studies. To eliminate any further inspection, the acceptable castings are painted green; the borderline group green-

red, and the not acceptable castings red. This marking enables us to determine the acceptance or rejection of the entire heat.

The reason for the borderline classification is that it has been found from actual tests that these castings in over 50 per cent of the cases will have a metal quality which is acceptable so that abnormal losses are not experienced after machining and assembly.

A standard metallurgical technique was developed in an effort to double check the x-ray findings. Representative castings were sectioned on the bandsaw and subsequently machined to a smooth finish by shaping which brings out the grain structure to a certain extent and finally macro-etched by immersion in an aqueous solution of 15 per cent nitric acid. The period of time of immersion will depend to some extent on the defective condition. It has been found that a uniform grain structure and equiaxed individual grains are generally speaking indicative of a sound casting. Defects such as porosity, blow holes, etc., will be readily visible to the naked eye after this preparation. In all but a few exceptional cases 100 per cent agreement was found to exist between the findings of the x-ray examination and metallurgical investigation.

Figure 13 shows a macrograph of a typical good metal structure.

Figure 14 shows a macrograph of a typical very poor structure. This was the same castings which was radiographed and shown in Fig. 12.

Figure 15 shows a photomicrograph of a good metal structure.

Figure 16 shows a photomicrograph of a poor metal structure.



Fig. 14

Figure 17 shows a photomicrograph of a good metal structure, the light spots are the gamma phase and

the dark spots are lead.

With the adoption of this practice and subsequent additions to it, we have been able to eliminate lost machine time and reduce machining and material cost as well as improve quality. These savings have been reflected in supplying foundries' practice. In view of the thorough and rigid inspection to which the castings are subjected more care is taken in the foundry, resulting in increased yields of acceptable castings and the reduction of foundry losses. This is true to such an extent that those suppliers who felt that under these conditions it would be necessary to receive an increased price per pound have actually found that they can, if they so desire, reduce the price per pound.

An additional advantage is that the life of the valves has been definitely extended. To what extent, we are unable to say, because sufficient time has not elapsed. We do know that an increase has taken place, particularly when they are used under varying corrosive conditions in handling many different products. We feel this is true because the elimination of the porosity on the finished surfaces has made it more difficult for the various media to attack the metal. The feeling is that when the porosity reaches the machined surfaces and is exposed to these products, that the porosity provides easier access to the metal and that something similar to intergranular

corrosion takes place.

In the beginning, as previously stated, the foundry reaction was one of violent opposition. Insistence on the use of this method of inspection met with a more or less passive resistance. We are pleased to report that the better foundries have now reached the point where on any new production they refuse to make any production runs until the samples have been submitted for x-ray examination and the complete re-



Fig. 15



Fig. 16

sults reported to them. Usually a single set of samples is sufficient for them to make necessary corrections.

We have made it standard practice to permit the supplier to examine the x-ray negatives with the object of improving the quality of his castings.

If it is found necessary, changes are made in the foundries' techniques, such as method of gating or risering, pouring temperature, etc. Once a satisfactory casting has been secured, they apparently experience little difficulty in maintaining this quality. At present, when rough castings are received from the foundry, they are impounded immediately upon receipt and 5 per cent are inspected or examined by the x-ray method. Never less than two castings are used to check a given heat of metal. If more than 10 per cent of the samples are classified as "not acceptable," an additional 10 per cent of the original shipment is inspected. If more than 10 per cent of the second lot fails, the entire lot of castings from that particular heat is rejected. It is worthy of note that rejections have been reduced to the apparent minimum, probably less than 2 per cent of the castings received being rejected. This has resulted in the increasing of production runs with the assurance that the assembled product will satisfactorily pass the required tests.

Figure 18 shows a medium size valve after it had been x-rayed, machined, tested and ready for use.

This paper has been based on experience with approximately 50,000 bronze* castings for valves. However, with some modification, it is being used on other non-ferrous metals, such as mercoloy,** monel*** and nickel.† It is being extended to other products where metal quality is a prime requisite.

ta

re

^{*} Typical analysis:--Cu, 88.0; Sn, 9.75; Pb, 2.00; Ni, nil; Impurities, 0.25.

^{**}Typical analysis:—Cu, 60; Zn, 10; Ni, 25; Sn, 1; Pb, 2; Fe,2.
***Typical analysis:—Ni (includes a small amount of cobalt), 63.00; Cu, 32.00; Fe, 1.50; Mn, 0.75; Si, 1.60,

⁺Typical analysis:—Fe, 0.25; Mn, 0.50; Si, 1.25; C, 0.50; Ni (including cobalt), 97.00.

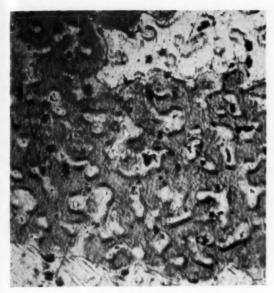


Fig. 17

DISCUSSION

Chairman: W. B. Scott, American Brake Shoe Co., National Bearing Div., Meadville, Pa.

Co-Chairman: H. G. SCHWAB, Bunting Brass & Bronze Co., Toledo. Ohio.

WM. ROMANOFF: The authors made the statement that they had slightly better results with virgin metal than they did with secondary metals. I would like to know the reason why.

MR. JOHNSON: I should say in the beginning that we are not foundrymen. We buy the bulk of our castings from outside sources. We started out with the premise that virgin metal was a necessity and*were able to increase our yield by different methods of purging. When virgin metal was difficult to obtain, we resorted to secondary metal, and our results, to all intents and purposes, were the same as those for virgin metal. Why, we cannot tell you.

W. N. RUTEN: Did you make fracture tests in connection with your x-ray examination for comparison purposes?

Mr. Johnson: Yes, with the 88-10-2 alloy.

MR. RUTEN: What did you find out from the fracture?

Mr. Johnson: The fractures will definitely support the radio-

H. J. Roast: One would expect the foundries producing the valves would take the same standard as the people to whom they sold the castings. Why do they not install an x-ray inspection unit, or at least employ someone to do it for them, rather than wait until the castings have been used?

Over a great many years I have not found the superiority of virgin metal over properly made composition ingot. The whole question seems to resolve itself in metal quality which can be controlled by a fracture test or the x-ray itself. It would seem to me that it is for the foundry to adopt this standard and that prior to the user of the castings.

CHAIRMAN SCOTT: X-ray technique is one thing and interpretation of the x-ray results is another. Most foundries are not equipped financially or personnel-wise to take on a complete radiographic study. Most of us are still small foundries by nature. It is helpful to find a laboratory that has enough know-how of foundry practice and willingness to go along with the step by step procedure to correct the condition.

Some of us had experience with aircraft-type castings which required 100 per cent x-ray inspection. On x-ray there was no quantitative method, while here they do have a grainy or mottled color. I would like to ask the authors whether they arrived at a quantitative method of x-ray interpretation?

MR. JOHNSON: We adhere to the three classifications mentioned in the paper. We have found that 50 per cent of those valves classified as borderline will be acceptable after machining. J. O'KEEFFE, JR.: The original fracture test that Howard Tay-

lor and I developed and published in 1948 indicated that the principal problem in obtaining high quality bronze valves, once the gating and risering set-up for production has been established, was to obtain uniformly high quality melts.

In testing this metal, we poured a fracture test from each pot of metal, and six bushings were poured from each pot of metal. We found that of 60 bushings poured from metal of fair or low quality, 30 of these bushings leaked when tested under 250 lb steam pressure. On the other hand, of 60 bushings poured from high quality heats, all of these castings were satisfactory when tested.

We found that as soon as the quality of the melt had been established by means of the fracture test, we could then predict whether the bushings cast from this melt would be uniformly sound or whether we would get a high percentage of leakers; that is, after the castings had been machined on the inside surface and on the outside.

I wanted to emphasize here the melt quality aspect. No matter what the previous history, whether the metal is virgin or ingot, so long as you know that the metal you have put into the casting is good quality, the results will be high quality

M. J. Davison: The authors stated that they rejected leaky castings on the basis of x-ray examination. Did they increase the yield of satisfactory castings from use of secondary metal by purging? I believe that is what they said. What did they purge with and was there any attempt to determine what was removed by purging? Was the material slag or sulphides?

MR. BROSKY: I will answer the question on the analysis of the material first and Mr. Johnson will answer the question on purging. Mr. Johnson submitted three samples to me of the material he purged out. The materials were so complex that we did not conclusively determine what it is, but it was a slaggy or silicate-type material.

MR. JOHNSON: We tried a number of different reagents forpurging and finally settled on chemically pure nitrogen. We purged the furnace for 6 to 8 min at a flow of about 10 to 12 cu ft per min. We purged the ladle for about 3 min with a flow of about the same magnitude. Anyone would have probably achieved the same results with other methods of purging.

Mr. Davison: The purpose of purging was to remove this slag from the secondary material was it not?

Mr. JOHNSON: Yes.

B. N. AMES: I would like to compliment the authors for introducing radiography as a control tool to improve castings. The authors stated that equiaxial crystallization is perhaps the best

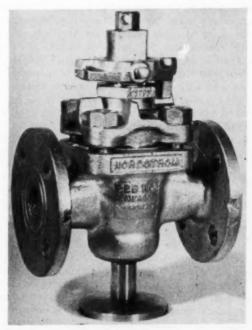


Fig. 18

structure. We have felt for a long time the mode of solidification of the tin bronzes can contribute considerably to pressure tightness or leakage, and it may or may not be tied up necessarily with metal quality. Equiaxial crystallization is more apt to give a greater amount of this microshrinkage due to the mode of solidification.

Utilization of the fracture test for melt quality control is excellent but I still feel we need an inspection field for cooling rates of castings which differ from that of the actual fracture

Mr. Johnson: In our own captive foundry we probaby were feeling our way. We changed the method of gating and experimented consideraby with the pouring temperatures. We

ti h C st

tı b ir d tı to

in ne to ti aı ol

m di re er fie of

of

51-

^{&#}x27;IVice President and Tech. Supt., H. Kramer & Co., Chicago.

a'Chairman, Industrial Process Dept., Polytechnic Institute of Brooklyn,

Brooklyn.

a'Bron're Foundry Consultant, London, Ontario, Canada.

a'The Falcon Bronze Co., Youngstown, Ohio.

Senior Metallurgist, New Lork Naval Shipyard, Brooklyn.

a'Met. Engr., Ohio Brass Co., Mansfield, Ohio.

Metallurgist, The Canada Metal Co. Ltd., Toronto, Ont., Canada.

simply pointed out to the foundry that this casting was bad and showed them why. We are now getting satisfactory commercial castings to such an extent that we have discontinued our own foundry, because we can buy them cheaper.

H. C. Ahl., Jr.: It is my understanding that you take sample pourings only for x-ray. Are the remainder of the valves hydraulically tested? Secondly, have you any comparative data where, rather than hydraulic testing, the valvces have been tested radiographically and does this show any improvement as a production-wise method?

Mr. Johnson: The finished valve is definitely tested at three different pressures: at twice the rated working pressure, at 11/2 times its working pressure for mechanical leaks, and finally at operating pressure.

MR. AHL: Any leaks you find during the subsequent hydraulic testing are not due to metal quality but to some outside source.

Mr. Johnson: I would say that in 98 per cent of the cases that is true, due to something other than metal quality.

RECENT ADVANCES IN DIELECTRIC CORE BAKING

By

J. Wesley Cable*

APPROXIMATELY 35 INSTALLATIONS of equipment for baking foundry cores dielectrically are in operation today. The accelerated rate at which these units have been installed is far more indicative of the acceptance of the method than the total number of installations. In 1945, two units were in operation. At the end of 1946, five units were functioning. In 1950, 17 of the total 35 units were placed in operation. It is often the case, when a radically new process is introduced to industry (and certainly dielectric core baking falls in this category), that a large number of installations are made immediately upon its introduction as a result of claims made by the manufacturers and developers of the process. It is also common to see further adoption of the process drop off rapidly because the equipment fails to perform in accordance with these claims. Such a trend in installations indicates either one of two things: the process or equipment is basically unsound, or it was offered to industry prematurely and further perfection was needed before it could be considered a production

The rising curve of dielectric core baking installations shows no such trend. Yearly increased acceptance of the process seems a certainty, based upon data obtained from its users. Advantages and savings are being obtained which were unheard of in the foundry industry.

Equipment

There have been no basic changes in the fundamental equipment used to heat foundry sand cores dielectrically. Consequently, there is no reason to repeat a discussion of the manner in which heat energy is introduced into the core by the dielectric field. Instead of altering the original design concepts of the equipment (and this shows that sound engineering went into the early production equipment), expansion of productivity has been the goal. Units of greater capacity have been made available to the

foundry industry. Not only have the pounds of cores baked per hour in a single unit been increased-up to four tons per hour-but the width of the conveyor belt and the spacing between the electrodes have been increased accordingly. Standard large-capacity units now accommodate core plates 72 in. in width, with an over-all core height of 18 in. Special units are readily constructed to any specific requirements. As a result of the development of larger units with greater capacity, the cost of equipment on a tonnage basis has been reduced considerably over the smaller capacity units. The larger size dielectric units also represent considerably less investment than completely mechanized conventional-type ovens of equivalent production capacity. And, they occupy only a small fraction of the floor space.

Cores are carried between the high-frequency electrodes progressively on a conveyor. Experience has shown that the steel chain-belt conveyor is far superior, under operating conditions encountered in the foundry, to rubber, canvas duck, or other materials.

Improvements have been made in the electrical circuits of dielectric core baking units to improve the operating characteristics of the equipment. Higher and higher electrode voltages are being utilized, and this improves the heating ability of the unit with no appreciable adverse variation in the mass distribution within the core. For example, cores with a high central body and wide base flange, such as the bell core for cast pipe, can be heated more uniformly and at a more rapid rate with increased electrode voltage. This higher potential also has done much to overcome the reduction in output which resulted from cores of different height being run through the unit simultaneously. Special control circuits have been developed which eliminate shut-down of the equipment as a result of occasional uncontrollable flash-overs between the high-voltage electrode and the cores passing beneath it. An "electrical brain," so to speak, is incorporated within the unit which differentiates between inconsequential flash-overs that cause the power circuit to be momentarily overloaded, and short-circuits of a serious nature that might cause damage to the equipment if it remained energized. Installations

n

ta

^{*} Consulting Engineer, Thermex Div., The Girdler Corp., Louisville, Ky.

J.

re

01

fo

ac

all

we

th

era

Cy

ers

fac

ba

ba

to

bii

gra

are

ing

the

the

esp

du

ura

ten

equipped with such a control circuit are automatically re-energized after flash-overs which occur from localized water condensation, ionization of the air between the electrodes due to instaneous high voltage gradients, and other phenomena which do not represent a faulty operating condition. This innovation has practically eliminated shut-downs of the equipment under normal operating procedure.

Resin Core Binders

While the core binders used in conjunction with dielectric core baking equipment do not represent an actual part of the installation, the over-all effectiveness of the process depends completely upon their performance. Consequently, it is in order to review the progress made in thermosetting resin type core binders so essential to the dielectric core baking method. The two basic materials from which such binders are made consist of urea-formaldehyde and phenolformaldehyde. This does not mean that these two compounds are the only resins from which core binders can be formulated. Others, especially melamine, have been used to advantage, particularly in Europe. It is the author's opinion that new materials will be developed which will prove superior in the future to those presently available . . . a viewpoint supported by the extensive research programs being carried on by practically every binder manufacturer. Nevertheless, the urea and phenol-base core binders available at present are vastly superior to those used at the introduction of dielectric core baking to the foundry industry.

In the early use of resin type core binders, dermatitis was occasionally encountered among workers who were exposed to these materials. Investigation showed that this was largely attributable to the free formal-dehyde present in the core mix as a result of the use of formaldehyde-base binders. Improvement in formulation, and better quality control in processing, have permitted the binder manufacturer to offer products which give off practically no free formaldehyde, before, during, or after mulling. As a result of these improved products, the problem has been reduced to the status of general industrial hygiene.

The reduction of free formaldehyde to a negligible amount has also done much to eliminate the problem of irritating fumes given off from the mix prior to baking, and also from the baked cores as they are removed from the dielectric equipment. Smarting of the eyes, accompanied by tearing, is caused by the presence of free formaldehyde. Any reduction in amount will correspondingly reduce the annoyance to operating personnel. This problem is closely allied with the odor problem which must be recognized when urea-base binders are used. While the reduction of free formaldehyde has done much to minimize this problem, the decomposition products of urea-formaldehyde have an unpleasant odor.

Like all objectionable odors, new ones are more noticeable than ones that have existed and been tolerated over an extended period. This accounts, in the opinion of many people, for the occasional objection to the use of urea-formaldehyde base core binders in the foundry. Good housekeeping, becoming more and more recognized as a necessary adjunct to successful foundry operation, will do much to eliminate the odor problem. If, however, this is not practiced, and proper shakeout procedure and ventilation cannot be offered to relieve the situation, the substitution of phenol-formaldehyde base binders will usually solve the problem, since these materials do not have the objectionable odors inherent to urea-formaldehyde compounds. Only in exceptional cases, has such a replacement failed to duplicate the physical requirements of the cores, though the mix may be more expensive when phenol-base materials are used.

Resin-type core binders can no longer be referred to as "dry binders," since they are now furnished in both the liquid and powder form. Liquid formulations seem to be favored by both the users and suppliers because the resin is more easily and uniformly dispersed throughout the sand with standard mulling equipment when the binder is added to the liquid state. Some shelf life is sacrificed, however, since the resins tend to "kick over" when stored for any appreciable length of time at room temperature. Improvements have been obtained through the use of liquid resins in that the total concentration of base resin has been substantially increased, thereby enhancing the performance of a given quantity of binder through greater percentage of active ingredients. Whether the resin binder chosen will be in the dry or liquid state will depend, to a great extent, upon the practice best suited to the particular foundry. Excellent results have been obtained on all types of cores with both classes.

Accelerators for Resin Core Binders

Accelerators have been developed which improve the characteristics of the cores made with resin-type binders. These manifest themselves in a number of ways: lowering the curing temperature, speeding up the cure at a given temperature, effecting a more complete cure, and producing a bond which is unaffected by water or moisture. Lower curing temperatures are important in the use of binders of this type in conjunction with dielectric baking. The heat energy required to bake a given weight of cores is drastically reduced, and a corresponding increase in output for a unit of given power rating is obtained. The uniformity of heating throughout the core, and the complete absence of "hot spots," makes the dielectric process especially suitable for use with the lower temperature resins. The advantages obtained through the use of accelerators, however, make such resin binders more critical under conventional baking procedures, since the inherent relationship between temperature distribution and mass distribution when a core is baked by thermal transmission from an elevated temperature atmosphere is bound to overbake the sections of less mass. Cores are also more susceptible to surface overbaking with lower temperature resins when conventional ovens are used.

This brings up the point of the evaluation of resintype binders, as applied to dielectric core baking, from the data obtained from test runs made in conventional ovens. Many prospective users of dielectric

NG

ful

the

nd

be

of

lve

he

de

a

re-

eX-

ed

in

la-

ly

ng

id

he

·e-

6-

id

as

1e

gh

te

St

ts

h

e

e

of

p

d

e

e-

ı

C

r

ì

d

e

e

core baking first question the use of resin-type binders to replace their presently-used core oil. In an attempt to determine whether or not the substitution will be satisfactory, representative cores made with resin-type binders are baked in their conventional ovens. In certain types of cores, of more or less uniform mass distribution, or where critical surface characteristics are not important, such tests are suitable to prove the point in question. More often, the geometry of the core is such that certain sections will overbake, or the surface characteristics will be impaired by overheating when conventional baking is employed, and the binder is considered unsatisfactory.

This same core, on the other hand, baked dielectrically and using the same binder and formulation, would, in all probability, have met specifications. If the test had been made with the conventional oven temperature held at the optimum specified curing temperature of the particular resin binder being tested, there is no doubt that good cores would have been produced. Baking with the lowered oven temperature would be so slow, however, that the baking cycle would equal or exceed that used with oil binders, and the claims made by the resin binder manufacturer for rapid baking would not be realized.

The suppliers of resin binders have formulated products which are suitable for conventional oven baking, and at the same time, capable of being baked dielectrically. This approach naturally leads to a compromise in baking characteristics, since the binder must be able to withstand the temperature gradients always present within the cores when they are baked in conventional ovens, and yet have a curing temperature sufficiently low to be adaptable to the dielectric method. There is every indication that the binder manufacturers will soon offer a product especially formulated for dielectric baking, with a low temperature polymerization point. The introduction of such a product will certainly be welcomed by the manufacturers of dielectric core baking equipment. With such a binder available, even greater advantages will result to the users of the process. Naturally, this binder will undoubtedly be sufficiently temperature-sensitive to preclude baking on conven-



A high-frequency dielectric core baking installation.

tional ovens, and the prospective user will not be able to try out the binder by baking test cores in his presently-used ovens. This will eliminate, however, the misconceptions of the characteristics of resin binder performance which, as already has been pointed out, result from such procedure.

As more and more foundries install dielectric core baking equipment and additional operating data are collected, the need for proving the acceptability of the process has rapidly diminished. When and if a binder especially compounded for dielectric use is offered, sufficient production data will undoubtedly be on hand to eliminate the necessity for demonstration. It is gratifying to observe the close cooperation between the binder manufacturers and suppliers of dielectric core baking equipment, since the realization by one of the problems that confront the other is bound to bring about results beneficial to the foundry industry. The high degree of perfection which has been attained today in synthetic resin core binders is a result of this interchange of knowledge. Continued and equally impressive improvement will undoubtedly be seen as time goes by.

Green Handling

Many claims were made for savings and advantages resulting from the handling of the core material and cores up to the time they entered the dielectric baking unit. These included less mulling time, loading directly from the core bench or core blower onto the conveyor of the dielectric equipment, elimination of intermediate loading racks or trays, and general labor saving in core making procedure. Analysis of operating data from foundries using dielectric core baking offers the following substantiating evidence for these claims.

Foundry "A" reports that it previously required five men to load the cores onto racks, place the racks in the ovens, remove them from the ovens to the cooling area, and finally transport the baked cores to the inspection room. The production line features of dielectric core baking achieved the same production with only two men, thereby saving three man-days of labor per day. Additional savings were obtained through increased productivity of the core makers. Under conventional oven baking procedures with rack-type ovens, approximately 15 per cent of the core makers' time was spent in loading and transporting the cores he made.

With dielectric core baking, only 5 per cent of his time was utilized in transferring the core plate from the core bench to the conveyor, thereby saving 10 per cent of the core makers' time for more productive endeavor. Mulling time was also decreased by 40 per cent, which practically doubled the capacity of the existing mulling equipment. Improved productivity of the core makers was also observed after the dielectric equipment was installed, and supervisory personnel believe this was due to the improved working conditions brought about by the introduction of the process. Previous oven installations had made the core room hot and dirty. These conditions, unhealthy

from a labor standpoint, were eliminated. Fewer rejected cores from sagging and slumping while in the

green state were also reported.

Foundry "B" was able to reduce the number of men needed for loading and unloading the ovens from four to two, thereby saving 2 man-days per day. The number of core makers for equivalent production was reduced from 18 to 15 through the increased productivity brought about by the streamlined features of the dielectric installation. Savings were also realized through less green state failures from slumping and sagging. And, improvement in general core room morale was noted as a result of the more attractive working conditions.

Foundry "C", on the basis of converting one-half of their core baking requirements to the dielectric process, and retaining conventional ovens for large cores (which represented approximately 50 per cent of their work in the core room), were able to get along with eight oven tenders as compared to 14 when

the core room used only oil-fired ovens.

Foundry "D", by converting completely to dielectric core baking, was able to eliminate entirely the labor of eight men required for oven loading and unloading, with the core room handling 16 tons of cores per 8-hr shift. The expected reduction of unsatisfactory cores as a result of change in dimensions while the cores were in the green state was also realized, and improved working conditions in the core room brought favorable comments from the workers.

Foundry "G", with a production of 22 tons of cores per 8-hr day, all cores being made on roll-over machines, was able to eliminate one of the three core makers and two of the four helpers . . . an appreciable

reduction in the total labor involved.

Foundry "H", in which the core room turned out 41/4 tons of cores on two shifts, was able to reduce the number of core makers to 15 from an original crew of 23, and eliminate three oven tenders and core handlers.

Obviously, not all these savings come from green handling improvement. But the majority of them are, in some way, attributable to this phase of dielectric core baking. The examples given have been chosen at random from the data accumulated, and each illustrate the point in question.

Baked Handling

Since the conveyor carries the baked cores from the dielectric baking equipment directly to the inspection department, or to tote boxes in which they are transported to the molding area, savings in baked handling also come about, just as in the case of green handling. The cores are relatively cool when they leave the high frequency electrodes, compared to those leaving a conventional oven, and can therefore be handled manually without intermediate cooling. This entirely eliminates the labor involved in the double handling of conventionally baked cores: once from the ovens to the cooling area, and again from the cooling area to inspection and the molding floor. The handling of core plates and driers in conjunction

with dielectric core baking is also facilitated along similar lines. Data obtained from the various foundries support these projected savings.

Foundry "A" now uses one man to handle and route the baked cores, compared to five men formerly employed when the baking was done in conventional oil-fired ovens. The absence of fumes and heat given off from the baked cores greatly improves the working conditions for this particular work and effects greater worker morale.

Foundry "D" reduced the number of handlers required to unload the ovens and transport the cores to the molding floor from 24 to 8, thus reducing labor expense for this particular operation to one-third its former cost. This was especially important because this foundry was located in a high cost labor area.

Foundry "F" was able to entirely eliminate the need for handlers after the cores had left the dielectric unit, since the exit conveyor was extended to carry the baked cores directly to the inspection tables. Four man-days of labor per day were thus saved.

Increased Core Room Yield

The data given above for green handling and baked handling largely involve the savings in labor brought about by the use of dielectric core baking. In every case, however, appreciable gains in over-all core room yield were realized. Causes for rejected cores were minimized in every way and the yield was increased to greater than 90 per cent in all the foundries studied. The improvement was brought about both in the green and baked states. In many instances, cores of vastly superior quality to that ever obtained from conventional baking methods were produced. One foundry reports that it previously had a core room yield of only 50 per cent, largely due to cores sagging in the green state, and that no methods of support or reinforcing could provide a remedy for the condition. Dielectric core baking completely solved the problem, because of the extremely short time the cores remained in the green state. After dielectric equipment was installed, the core room yield for this particular operation was increased to 93 per cent.

Over-baking is a constant source of trouble in attempting to increase the percentage of good cores produced in the core room. The over-baking may occur at a certain section of the individual core, caused by the nonuniform temperature distribution throughout the core as a result of variation in geometry, or a group of cores may come from the oven completely over-baked. The latter condition may be attributable to the manner in which the cores are placed with relation to each other as they are loaded into the oven. This is proved by the tendency to over-bake of the outer rows of a group of pin-cores placed on a coreplate, or by temperature variation within the oven over-baking the cores placed in the "hot spot" area. The absolute insurance that it is impossible to overbake cores which the foundry obtains when dielectric core baking is utilized entirely removes this cause of reduction of core room yield. Stresses set up within the cores as a result of nonuniform temperatures

to co tic an tro

COL

po

J.

th

to

reg fin wi ver the ver

are fact the mir from open

rea lab hal gas tio

70 and bot imp

has

ate

\G

ng

id-

nd

lv

al

en

k-

ts

e-

CS

01

its

se

ic

11

d

ıt

n

d

n

n

n

g

1

ı

throughout their mass have been shown in practice to be virtually non-existent when dielectric core baking is used. Consequently, core failures resulting from deformation and changes in dimensions beyond the specified tolerances caused by these stresses do not plague the core room foreman by contributing to low core room yield.

Improvements in Castings

The data accumulated as a result of the experience gained by the foundries using dielectric core baking show that the benefits of the process are not confined to the core room, in labor savings, improved working conditions, better cores and increased yield. Production improvements are noted on the molding floor and in the cleaning department, and the quality control group and sales department also share the advantages in the improved product that the foundry offers to its customers. The greater collapsibility of cores baked dielectrically using resin binders is reported by the foundries using the process as one of its outstanding features. For example, Foundry "L" reports that certain cored castings, involving thinfinned sections, could never be successfully produced with cores made from oil binders and baked in conventional ovens. The use of dielectric core baking, therefore, made it possible to take on work that was very profitable . . . work that formerly had to be turned down, much to the chagrin of the sales group.

Foundry "M" reports that harder and stronger cores are obtained by dielectric baking, and that the surfaces are smoother, thereby improving the finish of the casting on the surfaces formed by the core, and minimizing "burn-in." These surfaces are also free from sand inclusions, and cleaning and machining operations are thus greatly simplified and less costly. This particular foundry, along with numerous others, realizes improved shakeout to the extent that the labor involved in this effort has been reduced by one-half. Blow holes are avoided as a result of the low gas content of the resin binders and the rate of evolution of the gas minimizes casting rejects from this cause.

Resulting Economics

Foundry "I", making small pipe fittings, reports a 70 per cent reduction in labor required for cleaning, and a complete elimination of rejects from gassing—both the results of using dielectric core baking. The improved collapsibility of cores made by the process has also been reported to have done much to eliminate hot-tearing in thin-walled sections.



Samples of cores baked dielectrically.

Resin Mixes and Costs

Contrary to the opinion of many foundrymen, core mixes using resin binders are often less expensive than the oil mixes which are replaced. In substantiation of this original claim, it is well to look at some corresponding oil and resin mixes used in various founddries which have replaced their conventional ovens with dielectric core baking equipment.

FOUNDRY "A"

Ingredient	Oil Mix	Resin Mix
Sand "A"	1100 lb	1200 lb
Sand "B"	200 lb	100 lb
Core Oil	31/2 gal - \$4.41	
Cereal	6 lb - \$0.36	10 lb - \$0.60
Resin-urea-formaldehyde	****	15 lb - \$3.45
Water		6 gal
Binder Cost	\$4.77	\$4.05

FOUNDRY "B"

Ingredient	Oil Mix	Resin Mix
Sand	650 lb	650 lb
Core Oil	2.5 gal - \$2.375	
Cereal	2 lb \$0.12	5 lb - \$0.30
Resin-urea-formaldehyde		7.5 lb - \$1.725
Rinder Cost	\$9 405	\$2 025

FOUNDRY "D"

Ingredient	Oil Mix	Resin Mix
Sand	1440 lb	1440 lb
Cereal "A"	6 lb - \$0.34	10 lb - \$0.56
Cereal "B"	2 lb - \$0.13	
Core oil	6 qt - \$2.23	
Resin-phenol-formaldehyde		6 lb - \$1.35
Kerosene	****	3 qt - \$0.12
Parting agent		1 pt - \$0.24
Binder Cost	\$2.70	\$2.27

FOUNDRY "M"

Ingredient	Oil Mix	Resin Mix
Sand	800 lb	800 lb
Core oil	2.9 gal - \$2.90	
Resin-urea-formaldehyde		8 lb - \$1.76
Cereal	4 lb - \$0.24	4 lb - \$0.24
Binder Cost	\$3.14	\$2.00

For certain rigorous requirements, the cost of resin binders may be equal to or greater than the equivalent core oil, especially where the use of phenol-formaldehyde base binders is called for. In such cases, however, the additional binder cost represents a small part of the total operating expense and is offset many times by other savings resulting from the use of dielectric heating.

Fuel and Energy Costs

The high overall thermal efficiency of the dielectric core baking process makes it, without doubt, the most economical method for baking cores from a fuel cost standpoint. Even in the low cost, natural gas areas, it has shown itself able to compete favorably with gas-fired ovens of the most modern design. Since all of the heat energy from the unit is concentrated wholly in the cores themselves, with the equipment remaining at room temperature, this high thermal efficiency may be expected, and the basic concepts of

the operation of the process contrast sharply with conventional heating methods. Comparative operating costs of fuel-fired and dielectric core baking equipment verify these savings.

FOUNDRY "A"

5 tons of cores processed per 8-hr shift: Fuel cost—oil-fired ovens, 270 gal per shift at \$0.09	\$24.30
dielectric core baking, 442 kwh per shift at \$0.01	\$ 4.22
Savings - \$20.08 per 8-hr shift	

FOUNDRY "B"

7.4 tons of cores processed per 8-hr shift:	
Fuel Cost-oil-fired ovens	\$30.60
Incidental electrical energy	0.48
	\$31.08
Energy cost-dielectric core baking	\$ 9.95
Savings - \$21.13 per 8-hr shift	

FOUNDRY "E"

8 tons of cores processed per 8-hr shift:	
Fuel cost—coal-fired ovens	\$17.20
Energy cost—dieletcric core baking, 442 kwh at \$0.015	6.6
Savings - \$10.55 per 8-hr shift	

FOUNDRY "F"

8.2 tons of cores processed per 8-hr shift:	
Fuel cost-gas-fired ovens, gas at \$0.43 per 1,000 cu ft	\$12.96
Energy cost, dielectric core baking	
Electrical energy at \$0.007 per kwh	3.08
Savings - \$9.88 per 8-hr shift	

Maintenance

Maintenance costs are always difficult to establish, since they depend to such a great extent upon the operating procedures followed in individual foundries. Many plants are satisfied to let equipment run, with little or no maintenance, until an operating failure occurs, while others have scheduled maintenance work. However, since dielectric core baking equipment has reduced moving parts to just the conveyor, small blowers and their motors, and high temperatures, burners, and the dirt and soot inherently present with combustion equipment have been eliminated, physical maintenance should be reduced to a minimum. Production experience has borne this out. However, the electronic tubes, which are the heart of dielectric core baking equipment, require replacement after failure, and, based on service life data, the cost of tube replacement amounts to approximately \$0.25 per ton of baked cores. This value should be charged against the operating and maintenance cost of dielectric core baking equipment when cost studies are made.

As far as over-all maintenance is concerned, users of dielectric core baking equipment report that an occasional blow-down of the units with an air-hose will remove dirt and sand that has collected through usage. This constitutes the only maintenance operation required for satisfactory continuous production.

Space Requirements

The evaluation of space within a foundry and the calculation of savings resulting from the reduction of area for a given operation also present a difficult assignment. However, the basic design of dielectric core baking equipment plus the extremely small area that is occupied by the cores during the baking cycle, insure that the floor space and volume needed for a given output would be far less than equivalent conventional baking equipment. The comparison of space requirements for various installations of dielectric core baking equipment, along with evaluation of the space made by the users, show substantial savings. This space saving also reflects the reduction in floor area brought about by the streamlining of production, elimination of racks and cooling areas, and other similar advantages gained by the use of dielectric core baking.

FOUNDRY "A"

6 Tons of cores processed per 8-hr shift:	i	
Floor space-conventional oil-fired ovens	800 sq	fı
Floor space-dielectric core baking	115 sq	fı
Savings	685 sq	fi
Evaluated at \$1.25 per sq ft annually	\$856.00	

FOUNDRY "B"

7.4 tons of cores processed per 8-hr shift:		
Floor space-conventional oil-fired ovens	5,400 sq	ft
Floor space-dielectric core baking	3,600 sq	ft
Savings	1,800 sq	ft
Evaluated at \$1.50 per sq ft annually \$		

FOUNDRY "G"

22 tons of cores per 24 hr, three 8-hr shifts:		
Floor space-conventional oil-fired ovens	1,516 sq	f
Floor space—dielectric core baking		
Savings	1,016 sq	f
Evaluated at \$1.50 per sq ft annually	\$1,524.00	

Driers and Core Plates

When dielectric core baking was originally developed, it was immediately recognized that metal core plates and driers could not be used. First of all, the thermal capacity of the mass of metal is usually sufficient to pull heat away from the core and cause under-baking of the surfaces of the core in contact with the drier or core plate. Secondly, the fact that the metal is an electrical conductor affects the dielectric phenomen to the extent of often making the baking unsatisfactory. This latter effect is especially noticeable in driers where the unbaked cores rest in a relatively deep cavity. The metal of the drier, extending up the sides of the cores, produces an effective, electrical shield to the action of the high-frequency field.

The use of nonmetallic core plates presented no problem in the introduction of dielectric core baking to the foundry industry, since many foundries had already shifted over to asbestos-board core plates. Plates of this latter type were often less expensive than metal plates, were lighter in weight and, therefore, better from a handling standpoint, and absorbed less heat from the core baking equipment. Plates of NG

he

of as-

re

in-

n-

ce

ric he

or

n,

er

re

ft

ft

ft

ft

ft

ft

ft

ft

ft

e

f-

e

t

a

.



Sand core driers and core box bases, made from a strong plastic, for use with high-frequency dielectric core baking equipment.

asbestos-board composition are suitable for use with dielectric core baking equipment, but because of the total absence of high temperatures encountered in conventional oven baking equipment, materials of superior physical properties for the dielectric process are available. Extremely light weight composition board, when used for core plates, greatly improves the operating characteristics of the dielectric process, and plywood core plates have been used to advantage accompanied by an appreciable savings in investment.

The importance of using core plates and driers of low thermal capacity should not be overlooked in the application of dielectric heating for core baking. Any nonmetallic material placed between the high frequency electrodes will absorb a certain amount of heat energy. Therefore, it is advisable to choose a material having electrical characteristics which will result in the minimum amount of energy being transferred to the plate or drier, and yet not provide a chilling action to the cores. The formula

$$W = \frac{3413 \times P}{30(1 + K) + 11M}$$
 (1)

can be used to approximate the pounds of cores per hour which can be obtained for a dielectric unit of given output capacity. In this formula,

W = pounds of cores per hour

M = Moisture content of sand, per cent

K = Compensating factor for driers, core plates, reinforcing rods or chills

P = Power rating of dielectric unit, kilowatts

The factor K may be computed from the respective weights and specific heats of the cores, core plates, driers, and metal parts such as chills and rods. For example, for a core weighing 10 lb and a drier weighing 5 lb, with 40 per cent of the area of the drier covered by the core,

$$K_1 = \frac{\frac{\text{per cent area of coverage}}{100}}{\text{Wt. of core} \times \text{Sp. Ht. of drier}} \times \text{Wt. of drier} \times \text{Sp. Ht. of drier}$$

$$= \frac{0.40 \times 5.0 \times 0.20}{10.0 \times 0.20} = 0.20$$
(2)

If the core contains a reinforcing rod weighing 1.0 lb, additional allowances would have to be made in computing K.

$$K_2 = \frac{\text{Wt. of rod} \times \text{Spt. Ht. of rod}}{\text{Wt. of core} \times \text{Sp. Ht. of core}} = \frac{1.0 \times 0.10}{10.0 \times 0.20}$$

$$= 0.05$$

$$K = K_1 + K_2 = 0.20 + 0.05 = 0.25$$
(3)

Therefore, with a dielectric heating unit rated at 15 kw output, the weight of cores of the above type which could be produced, with a moisture content of 2 per cent, is, from Eq. 1

$$W = \frac{3413 \times 15.0}{30 (1 + 0.25) + 11 \times 2.0}$$

= 860 lb per hr

If no driers or reinforcing rods were used in the cores, the value of K becomes zero, and the pounds per hour of cores which can be produced by a unit of given capacity becomes a function of the moisture content of the sand mix. Under these conditions, Eq. 1 can be expressed

$$W = \frac{3413 \times P}{30 + 11M} \tag{4}$$

Values of W as a function of M, with P equal to 15.0 kw, are shown in Fig. 1, plotted from Eq. 4. The importance of retaining the moisture content as low as possible for maximum production is graphically demonstrated.

Figure 2 shows a family of curves of different moisture contents which give production as a function of K, plotted from Eq. 1.

The economies obtained from dielectric core baking which are reflected in the investment in core plates and driers result from the short baking cycle inherent to the process. Reports from the foundries using the method show that the number of driers or core plates for a given operation can be reduced to one-twelfth to one-fifteenth the number needed with conventional oven practice. Replacement and maintenance are also reduced appreciably by the use of dielectric heating, since the lower temperatures and short heating cycles greatly increase the service life of the plates and driers. The accumulation of dirt and oil on the surfaces of the parts is minimized, and the cost of cleaning the plates and driers periodically is reduced accordingly.

The fabrication of nonmetallic plastic driers for dielectric core baking is an accepted procedure. Tech-

h

b

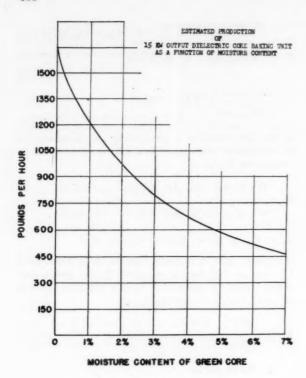


Fig. 1

niques have been developed for producing driers which are satisfactory for use in dielectric heating equipment, and which have suitable physical characteristics to give the required dimensional tolerances, along with resistance to wear and abrasion over continued service life. The particular material which has proven most satisfactory for the fabrication of these driers is a polyester casting resin, reinforced with glass fiber.

The material is supplied in the liquid state, and upon subsequent addition of accelerators and catalysts, is ready for use. It is poured into a metal mold, made in accordance with the same procedures and tolerances for metal driers, and backed up with a matching member. The assembly, clamped together, is then baked in an oven. After the resin is cured, the mold is parted and the drier removed. The matching surface is then finished on a surface grinder to fit the core box, sufficient grinding stock having been allowed in the design of the master mold. The cost of making plastic driers is comparable to that for metal driers, and the personnel of the pattern shop are, in general, well qualified to carry out this work. If, on the other hand, the foundry desires to purchase driers from an outside source, at least one manufacturer of dielectric core baking equipment will furnish them. Typical cost figures for a shell-type drier having an area of 100 sq in. are as follows:

Material: 1 to $1\frac{1}{2}$ lb @ \$0.50 per lb \$0.50 to \$0.75 Labor: 2 man-hours @ \$1.50 per man-hour \$3.00 Total cost of drier (not including overhead) . . . \$3.50 to \$3.75

On the basis of new work, the savings obtained from reduction of the number of driers through the use of dielectric core baking will often exceed the cost of capital equipment involved. This may seem incongruous, but an example will serve to illustrate the point.

Foundry "J" reports that where 600 driers were formerly required for a given operation, 50 plastic driers were ample for a comparable job using dielectric core baking. At an initial cost of \$3.00 per drier, which they consider a conservative figure, the savings in drier cost for this single core amounts to \$1,650.00. If ten such jobs were processed on the high-frequency equipment, the saving would be greater than the capital investment for a unit rated at one-half ton of cores per hour. Obviously, this line of reasoning cannot be applied to work for which metal driers are already fabricated and available for use in conventional ovens. These driers are carried on the books of the foundry as a capital expenditure, and they would have to be scrapped and retired if dielectric heating replaced the conventional ovens. In spite of this, however, many foundries have seen fit to approach the problem in this manner.

Blowing Cores Directly into Driers

One leading manufacturer of dielectric core baking equipment was quick to realize the potential advantages combining the process with "blow-in" driers would offer. Therefore, not satisfied simply to develop a drier that was suitable for use in the dielectric equipment, they continued their investigations until a drier was developed and made ready for production that could form half of the core box and have cores blown directly into it. Driers can now be fabricated to meet these requirements by the method

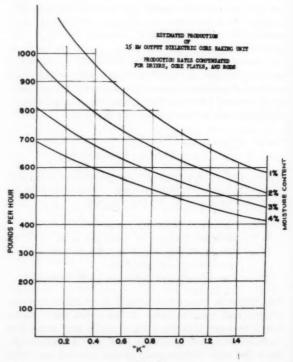


Fig. 2

1.\G

on-

the

ere

tic

ec-

er,

igs

00.

CY

ap-

of

m-

re

n-

of

ld

ng

he

d.

rs

e-

18

0

already outlined. This advancement placed an entirely different light on the economic phases of dielectric core baking, since it soon became obvious that a prospective user of "blow-in" driers could not afford to use any type of baking equipment other than the dielectric process. In the first place, metal driers in general that are used with conventional ovens cannot be adapted to the "blow-in" procedure, due to their construction and dimensional tolerances. Therefore, if the foundry is to blow directly into the drier in their core blowing operations, new driers would be required to replace the existing ones, and these latter components scrapped.

This places the operation on the same status as a new job coming into the shop, and the reduction in the number of driers brought about through the utilization of dielectric core baking can result in tremendous savings. For example, a pipe fitting foundry which employs dielectric core baking, tooled up 51 jobs for blow-in driers. With conventional baking methods, approximately 1,200 driers per job would have been needed, whereas only 100 driers per job were required in conjunction with the dielectric process; the larger numbers being a result of the extremely high production obtained on each core blower when blowing directly into the drier. The cost of metal blow-in driers for these particular operations averaged \$3.70 each, and the savings involved in the reduction of the total number of driers by 61,200 as a result of using dielectric core baking, indicated an ultimate savings of \$226,440.00 in driers alone. Similar jobs show comparable savings, and where the practice of blowing directly into driers is applicable, dielectric core baking is a natural adjunct.

The labor savings which result from blowing into driers are appreciable. And, although it is not a direct part of a discussion of advances in dielectric core baking, the combination of the two processes is sufficiently interesting to warrant inclusion in this paper. On a typical job, using a standard core blower, the best production which could be obtained in one particular foundry by blowing into a split core box and turning out the cores by hand into a transfer drier was 31/2 blows per minute. With the same blower and core maker, the production was speeded up to 6 blows per minute using blow-in driers as half of the core box, but parting the box manually. By changing the blower so that the box was parted automatically, the production was increased to 8 blows per minute. Many of these core boxes contained as many as six cavities, and using the procedure outlined above, a single core maker could produce close to 3,000 cores per hour. Over a number of jobs in the shop, this foundry reported that an increase of 4.2 to 1 was obtained in the average productivity of each core maker through the introduction of blow-in drier procedures with automatic parting.

New Operating Techniques

Certain operating techniques have been developed that have enhanced the applicability of dielectric core baking to general foundry practice. Scheduling of core making operations so that the conveyor belt of the dielectric unit is kept fully loaded has done much to insure maximum production from the equipment. Proper placement of cores on core plates has also contributed to the productivity of the units. Since the same belt area is occupied by the core plates regardless of the number of cores on the plate. Consequently, any increase in the total mass of cores placed upon the core plate will improve the loading of the dielectric core baking equipment.

Spraying and fogging of the cores, either with water or a resin solution, before they enter the electrodes has improved the physical characteristics of the baked cores. Data on the properties of cores made with resin binders show that the hardness is increased as the moisture content of the mix is raised. Therefore, fogging of the surfaces of the cores with a water spray prior to baking improves their surface hardness, yet does not add sufficient moisture to the total percentage to reduce the production rate of the equipment appreciably.

The use of a resin solution as a spray further improves the surface characteristics of the core, since it increases the resin content of the outer layers of core mix. Adaptation of this procedure has permitted some users to eliminate the necessity of a wash or blackening after baking, thereby bringing about a saving as a result of the elimination of this entire operation. There is some evidence that a water spray will cause some migration of the resin binder to the surface of the core and so concentrate the resin in the outer masses of the core. This phenomenon tends to further the beneficial results obtained by the use of water sprays, but where higher physicals are required, the resin solution spray is often better suited to produce the desired results.

For such techniques, the spray nozzles are mounted directly above the conveyor at the entrance to the dielectric unit. They should be directed in such a manner as to completely cover the core surface with water or resin solution, but at the same time, prevent the fog from entering the dielectric equipment. The introduction of appreciable quantities of water vapor between the high voltage electrodes makes the equipment susceptible to flash-over, and should be avoided. Where resin solution sprays are used, additional maintenance may be necessary to clean the conveyor belt and remove the accumulated resin. Occasional lubrication of the belt will prevent the resin from adhering to it and will improve its mechanical operation.

Summary

Dielectric core baking has proved its worth during the relatively short period of its use in the foundry industry. The savings predicted have, in general, been realized; often being far greater than anticipated. Resin binders have been improved, and their use with dielectric heating shows that practically any core can be produced to the required properties. Further improvement in binder formulation and performance can undoubtedly be expected in the future.

Driers are available for use with dielectric core baking equipment. These driers, made of plastic,

J.

sh

re

21

in

bl

an

he

ef

are capable of being blown into directly from the core blower. Unheard of savings in drier cost, usually exceeding many times the cost of capital equipment involved, can be realized with their use in conjunction with the short baking cycle inherent to the dielectric process. As a foundry production tool, dielectric core baking is receiving wide adoption, with installations being made in all branches of the industry. It appears safe to predict that in the not-too-distant future, a large percentage of all cores made will be baked by the dielectric process.

DISCUSSION

Chairman: E. C. ZIRZOW, Deere & Co., Moline, Ill.

Co-Chairman: H. K. SALZBERG, The Borden Co., Bainbridge,

N. Y.

W. B. BISHOP (Written Discussion): ¹ The subject of dielectric baking is arousing interest on the part of all foundrymen striving for production of better castings at lower cost. To be able to form accurate judgments as to the economics of dielectric baking as compared with other forms of baking, it is necessary that foundrymen be presented with as many accurate facts as possible. In the next year many foundrymen attempting to judge the economics of dielectric baking will refer to Mr. Cable's paper. The accuracy of their judgments depends, to a large extent, upon the accuracy of the figures presented.

Insofar as the economics of fuel, energy, operation, and maintenance costs of dielectric ovens as compared to traditional ovens are concerned, I am not sufficiently conversant with these factors

to venture a judgment.

A good portion of Mr. Cable's paper is devoted to a comparison of oil-sand mixtures as used in traditional ovens to synthetic resin mixtures as used in dielectric ovens. In substantiation of his claim that "core mixes using resin binders are often less expensive than oil mixes," Mr. Cable gives four examples in his paper. These are listed as Foundry "A," "B," "D," and "M." As a representative of the core binding industry who concentrates his daily efforts primarily upon core sand and binding problems, I believe I can comment with some accuracy upon these mixture cost comparisons.

The choice of examples used is unfortunate and may tend to be misleading to the uninitiated if not thoroughly analyzed. In each of these examples impractical and inefficient core oil mixes are used as being typical. Comparing them to the resin mixes presented is like comparing a 1924 Ford with a 1951

Chevrolet.

For example, Foundry "A" is represented as using 2.13 per cent oil, Foundry "B" 3 per cent oil, Foundry "M" 3 per cent oil, all by weight, in their core sand mixtures. In each case this amount is exceedingly excessive and because of this excessiveness would not be considered a good commercial mix by any practical foundryman. The most intricate cores in the United States requiring the highest tensile values (automobile jackets, radiator loops, etc.) are made daily with 1 per cent oil binder. Amounts in excess of this are unnecessary in good sands and only serve to consume fuel and retard the bake.

Using the cost for oil binders as submitted and reducing the per cent of oil used by weight to slightly over 1 per cent in each case the proper cost comparison for binders in Foundry "A," using 15 lb of oil, would have been \$2.71 for oil mix versus \$4.05 for a resin mix. In Foundry "B," using 71/2 lb of oil, the cost comparison would be \$1.00 for oil mix versus \$2.025 for resin mix. In Foundry "M," using 8 lb of oil, the cost comparisons would be \$1.24 per oil mix versus \$2.00 for resin mix.

Aside from this, it is general knowledge that resins are unworkable without the addition of a release agent to alleviate their inherent stickiness. These release agents cost a minimum of \$.07 per lb, must be incorporated in all resin mixes, and their cost is an integral part of the total binding cost of the mixture. Resin mixes for Foundries "A," "B," and "M," make no provision for this release or parting agent. The example of Foundry "D" does indicate that 3 qt of kerosene and 1 pt of parting agent, totaling \$.36 in cost, must be added to this mixture. Undoubtedly some similar cost value should be applied to

each resin mixture shown if cost comparisons are to be accurate.

Mr. Cable states that in some cases "the cost of resin binders

may be equal to or greater than the equivalent core oil—however, the addition of binder cost represents a small part of the total operating expense and is offset many times by other savings resulting from the use of dielectric heating." This gives me the impression that binder costs are a small factor in core production. The fact is that in a ton of core sand the binder content usually costs at least as much, if not more, than the total sand involved. \$4.00 per ton of core sand binder cost is a conservative figure. Mr. Cable discussed fuel cost per ton of core sand as being anywhere from \$.66 to \$1.25. Looked at in this manner, binder costs are approximately four to seven times as important to consider in core production as are fuel costs.

Since a core is an expendable item, the cheapest core that will do the prescribed job should be used. It has been my experience, and I think it is accurate to say, that "at the spout" oil-sand core costs are, day in and day out, 30 to 50 per cent less expensive than resin-sand core mixtures. Lower fuel costs may compensate for a portion of this difference in cost but seldom, if ever, for

all of it

Therefore, I think it is accurate to say that economies through the use of dielectric baking must be looked for elsewhere than in the cost of the core sand binder mixes used. I believe that Mr. Cable erred in claiming possible savings in this phase of core production. That the speed of dielectric baking may permit economies which will compensate for additional mixture costs may be entirely possible. However, I believe that a technical paper should be as accurate as possible in all phases and that the mixture cost comparisons used in Mr. Cable's paper are misleading and tend to confuse a foundryman seeking accurate information upon which to base his decisions.

Mr. CABLE: Mr. Bishop is quite correct in his statements. I did not bring out those points to attempt to show that any appreciable savings are brought about through the use of resin binders. It was only an attempt to show typical core mixes. In regard to those oil-sand mixes, we simply took the mixtures

the foundry gave us.

When you say it makes no difference in economy in the speed of bake I must take issue. That \$266,000 worth of driers which

was saved resulted solely from rapid baking.

MR. BISHOP: I agree. I cannot contend with that. The reason for faster binders today is speed of bake necessitated by shortage of oven capacity. I think as a corollary to speed you must sacrifice cost, and if you have adequate oven capacity you can usually make the most economical core by using the cheapest binder which, fortunately or unfortunately, today is still core oil.

V. E. HILLMAN (Written Discussion): 2 J. Wesley Cable's paper pertaining to recent advances in Dielectric Core Baking has been read with interest and in our opinion the author has covered the subject in a most commendable manner. The results of his experience are in accordance with those of our own practice. However, we would comment on the following points:

Equipment—It was noted that the author tells of the improvements that the manufacturers are making in this equipment, such as higher and higher electrode voltages. This to our minds will increase the all-around efficiency of this type core oven. As we do not believe at the present time a complete cure is

affected when phenol formaldehyde is used.

Resin Core Binders—We believe as Mr. Cable does that the resin manufacturers through research will develop a new resin that will be superior to the urea and phenol formaldehydes. In fact, we have tested a new resin that the manufacturer claims is not a urea or a phenol. This resin cures completely in the dielectric equipment. However, the cost of this resin at the present time makes its use prohibitive. Another disadvantage is that the cores made using this resin were very hygroscopic.

Release Agents—It is our opinion that Mr. Cable missed a point when he did not mention the use of release agents in his paper. Resin binders are noted for causing stickiness in the core box. This condition has plagued foundries for some time. We believe that by the use of a good release agent, production

can be increased.

Accelerators for Resin Core Binders—We would be rather hesitant at the present time to substitute anything for boric acid as an accelerator. It has been our experience that other types are more expensive and in most cases very critical in use.

¹ Mgr., Foundry Sales, Archer-Daniels-Midland Co., Cleveland.

³ Director of Research, Crompton & Knowles Loom Works, Worcester, Mass.

ate.

lers

the

1123

the

uc-

nt

nd

ive

as

er.

101

ill

ce,

re

or

gh

an

at

of

iit

its

al

at

S

te

S

es

d

h

ı

re

Y

g

15

19

Resin Mixes-It is our opinion that the author should have brought forth the importance of control at the muller. By control we mean that sand, resin, cereal and water should be weighed very carefully, as this will affect the baking time. It should be pointed out that the man operating the muller is responsible for the good or poor work of all the core makers. We have cut mulling time from 12 to 5 min with the use of resins and this gives the operator ample time to be conscientious in weighing the material used.

D. C. WILLIAMS: 3 One of the most interesting statements made by Mr. Cable is that they were working toward lowertemperature curing binders. Does he mean the cores will be cured below 212 F? If it is possible to go below 212 F why could

we not expect to have room temperature curing?

MR. CABLE: I did not intimate that we were going below 212 F. Of course, we have to eliminate the moisture in the core. I think we all know the effect of moisture content. Therefore, in order to remove that moisture, we will always have to go up

to the evaporation point of the water present.

I do not think it can be open to criticism if I say this about the core binder suppliers. They have not seen fit, and I do not blame them, to develop a specific binder solely for core baking. They have had to look for markets for binders and in so doing they had to compromise as far as baking was concerned. So I am not thinking in terms of temperatures under 212 F and I certainly do not want to go down to room temperature curing because then we would have no use for dielectric heating.

H. W. MEYERS: 4 What is the maximum weight core that can

be baked in a dielectric oven?

Mr. CABLE: We baked one that weighed 680 lb. However that was not in a standard oven. We have chosen design standards in building dielectric baking equipment which we felt would be adaptable to the average foundry, but there is no limit to the size and mass of the core that can be baked dielectrically.

Mr. MEYERS: Would that size equipment be practical from

the standpoint of cost?

MR. CABLE: Yes, if you had sufficient cores to bake. We simply set our electrodes on the standard unit to 18 in. in height and 36 in. in width. That covers most of the core applications we encounter, but there would be no reason why these dimensions could not be increased.

Mr. MEYERS: Do you always have to select or group cores of approximately equal size? Can you bake a large core then immediately bake a core of about 1/3 the size of the first one?

Mr. Cable: Suppose we start with a 3 to 1 height ratio. You would find there would be very little reduction in overall production with such height variation. On the other hand, if the height ratio was 12 to 1 or 15 to 1, then an attempt should be made to schedule those cores.

MR. MEYERS: Approximately how much did you have to increase the drying time or slow down your cycle in baking the

680-lb core?

MR. CABLE: The problem resolves itself into a Btu-evaporationmass consideration. I think that large core required something like 28 or 30 minutes. It is purely a Btu transfer for driving off

MR. MEYERS: Would the moisture be removed throughout the core?

MR. CABLE: Absolutely.

Mr. MEYERS: Please comment on rod reinforcing of cores.

Mr. CABLE: Reinforcing rods in a horizontal plane offer no problem except that they must be considered from a heatabsorption standpoint. Because of their higher densities, this effect will be more pronounced than the equivalent volume of core sand. From the electrical standpoint, if the rod is horizontal or within 45 degrees, it will not affect the baking process. However, if we were to have a rod going from the top to the bottom, you can see that would prevent building up a dielectric field across the core. Under those conditions, we might try to change our methods and use a wood or plastic reinforcing rod.

H. ATKINSON: 5 I would like to have the author discuss the subject of core curing further. Suppose you use a synthetic resin core mix and only half bake the core. Can a good casting be

made from such core-will it not cause a blow?

and driving off the moisture. They are accomplished practically simultaneously, because as we reach the cure point of the resin, the water is being driven off and the resin will polymerize and cure. It is the degree of cure needed to achieve the desired core characteristics which determines the ultimate core temperature. It is often possible to increase the output of the oven by operating at lower temperatures. In some cases, the cores are not completely cured when they immediately leave the ovens, but subsequently cure when the water fully leaves the core. CHAIRMAN ZIRZOW: I want to correct one statement. The assertion was made that you cannot make a core for a casting

MR. CABLE: We have two distinct functions to perform in

running a core through a dielectric unit; curing the resin binder

free of defects or blows from the conventional core oils. That is not true. We made tons and tons of cores that produced good

castings out of the conventional core oils.

C. H. BARNETT: 8 I do not wish to take issue with Mr. Cable's remarks in regard to plastic dryers. I think they are a very interesting development; and, if they work out successfully, they will be a help to the dielectric core baking process.

There are other points in Mr. Cable's paper, however, to which I would like to take exception, either in whole or part. For example, in going through the paper, we come across repeated references to "conventional ovens." I think we ought to get straight on just what we mean by a "conventional oven." The right definition of this explains Mr. Cable's high fuel cost figures and it explains the rich mixtures with which Mr. Bishop took issue. If, by a "conventional oven," Mr. Cable means oldfashioned massive brick construction, with very high temperature gradients, because of which you have to use very heavy plates and racks and very rich mixes, or where you have temperatures in various parts of the oven plus or minus as much as 150 or 200 degrees from the control point, these are emphatically not "conventional ovens" by the standards we have today. We know that resin mixes and lean oil mixes of 1 per cent by weight or less are being successfully baked in modern "conventional" fuel-fired ovens and are producing very good cores that are free from gassing and other surface defects and which shake out easily. We know we are getting these cores in what we mean today by "conventional ovens," that is, accurate automatic temperature control and recirculating heating systems, using a large volume of air as the heat transfer medium, thereby holding down the temperature gradient and making temperature uniform throughout the oven.

In today's fuel-fired ovens, we can bake a lean oil or a lean resin mix at the maximum temperature which the binder will stand, because we know and have complete control over the temperatures throughout the ovens. I dare say there is a far greater tonnage of resins being baked in modern, conventional ovens than there is in dielectric ovens. A certain foundry is making 32 tons of cores daily, and baking about 80 per cent of this tonnage in resin sands. I know of another foundry where they are making 20 tons a day of one particular core. This core is about 31/2 in. thick and about the size of your head, and they are making these with a urea-formaldehyde resin and doing a

wonderful job in modern fuel-fired ovens.

I would like to cite a few examples myself in connection with these fuel cost figures. When you analyze the examples given in Mr. Cable's paper, it becomes evident that the utilization of the fuel is very inefficient. The figure of 30 odd dollars per shift, for instance, in the example of Foundry A, actually should not be more than 4 or 4 and a half dollars for the tonnage baked, which makes a saving at least equal to that using electrical energy in dielectric baking. To substantiate that, I have analyzed these other examples. They come out, generally, with fuel dielectric energy costs at worst a stand-off, at best with fuel showing a better figure. If we calculate fuel cost for the lower temperatures used with resins, we get even better results. I would like to mention again the same foundry previously mentioned, baking 32 tons daily, 80 per cent of which is resin bonded cores. They are using fuel oil and doing it at a cost of less than \$1.00 per ton.

In another foundry, using natural gas at 60 cents per 1000 cu ft, they have a fuel cost of \$1.35 per ton.

Another case, with 5 tons of oil sand cores, shows a fuel cost of 80 cents per ton; and, in still another, using fuel oil at \$.095 a gallon, a fuel cost of \$1.13 per ton.

⁸ Ohio State University, Columbus, Ohio.

Foreman, Sand Testers, General Steel Castings Corp., Granite City, Ill.

⁵ Methods Engr., International Harvester Co. of Canada, Ltd., Hamilton, Ontario, Canada.

⁸ Foundry Equipment Co., Cleveland.

fo

eff

tic th co th eff

in

by no the so na en of ca sec po tio ca co an wl

m

Now these are all in modern, conventional, forced convection, recirculating heat type ovens making efficient use of the fuel. In most instances, in mechanized ovens, a portion of the heat is recovered.

I would like to touch briefly also on maintenance and repair costs, which I am afraid Mr. Cable glossed lightly over in his discussion. Mr. Cable gives us a figure for this item of \$.25 per ton of cores, which he says should be charged to the operating cost of dielectric baking equipment. Going back to the examples of Foundries A, B, E, and F, you will find that this \$.25 per ton, when added to the tabulated operating costs, alters the picture considerably.

We have found in our experience that a figure of about \$.05 per ton for maintenance and repairs is normal with modern conveyor-type fuel-fired ovens; and, in fact, one foundry, which has a top notch preventive maintenance program, reports a maintenance and repair cost on its conveyor core ovens of \$.045 cents per ton of cores baked.

Incidentally, quite a few operators of dielectric core baking equipment have told us that their maintenance and repair bills run around \$1.00 per ton of cores baked.

In connection with increased coreroom yield, I am afraid that Mr. Cable is making an advantage for the dielectric process out of the limitations of antiquated fuel-fired ovens which cannot safely be "pushed." After all, any piece of equipment has a maximum capacity; but, because modern ovens are highly flexible as to operating temperature and provide temperature uniformity even at very high temperatures, satisfactory cores can be obtained more quickly at the high temperatures, although there is a limit to the practicality of doing this. You must consider the class of work, the binders, etc. Still, we know of numerous instances where foundries are getting twice and even three times as much production as originally designed for, out

of modern conveyor type ovens. They do this by increasing temperature and shortening the cycle. On the other hand, if you try to push a dielectric oven, you are going to get underbaked cores.

Most of the advantages Mr. Cable mentions are not advantages, as I see it, of the process itself, but of mechanized handling. You get the same thing in modern, fuel-fired conveyor ovens where you place the work on a moving conveyor and the cores go directly into the oven, so that you do not have the coremakers looking for a place to put their work and generally spending their time doing things that are not coremaking. The same can be said for baked core handling.

Briefly, on wires and rods, I would like to mention that we had a foundryman tell us that, after he put in modern fuel-fired ovens to replace his old-fashioned ovens, he was able to eliminate wires completely from many cores and cut way down on wiring and rodding of the others.

Therefore, it seems to me that you cannot generalize about the dielectric process—that not only each foundry, but each core job in each foundry has to be studied on its merits for its dielectric adaptability.

MR. CABLE: My use of the word "conventional" is solely one which denotes a type of baking other than dielectric heating. If you have a better word I am willing to use it. I simply use the word "conventional" in that terminology.

Mr. BARNETT: The word we would like to use is "old-fashioned" or "relic." Unfortunately, Mr. Cable, there are a lot of those old ovens around, and I join with you in condemning them. It is remarkable what some foundries are able to accomplish with these, but we and Mr. Bishop could show them a few tricks in keeping up to date.

MR. CABLE: That is right. My knowledge is statistical and there is nothing implied to the contrary.

THE CONTRIBUTION OF RISER AND CASTING END EFFECTS TO SOUNDNESS OF CAST STEEL BARS

By

H. F. Bishop,* E. T. Myskowski* and W. S. Pellini**

ABSTRACT

The length of cast bars which can be made sound has been determined radiographically and related to thermal gradient conditions. It was found that for steels containing 0.25 to 0.35 per cent carbon the maximum distance in inches which can be made sound follows an empirical rule given by the formula $D_{max} = 6/T$ (T = section thickness). This relationship holds for bars in a thickness range of 2 to 6 in., cast in either the vertical or horizontal position. Vertical 8-in. bars can be made sound for somewhat greater distance due to convection current effects.

Thermal analyses showed that a minimum gradient of 6 to 1? F/in. is required to feed bar sections to complete soundness. This observation is discussed in relation to the gradient conditions required for feeding of plates reported previously to be in the order of 1 to 2 F/in. It is deduced that differences in the course of solidification of bar and plate sections attributable to the geometries of the respective sections are the basis for the effect.

It is concluded that directional solidification does not necessarily assure soundness. Depending on the geometry of the casting a minimum gradient condition in excess of that required for directional solidification may be required for feeding to soundness.

EFFECTIVE FEEDING DISTANCE OF RISERS is dictated by thermal gradient conditions which are influenced not only by the riser itself but also by the geometry of the casting. Proper positioning of risers to assure solidity with utmost yield therefore requires that the nature of the association of riser gradients with gradients arising from casting edges be known as a function of casting geometry.

A previous report¹ described the nature of risercasting edge associations for plate castings of uniform section. A system of triangulation was indicated for positioning risers on plate sections such that all portions of the plate were influenced by riser and/or casting edge gradients of proper values to determine complete soundness. By calculating the surface area and volume of the segment which each riser feeds when located according to these relationships and then applying the formulae developed by Caine,² the minimum riser sizes which prevent both underriser shrinkage and centerline-shrinkage can be determined.

The present report presents the results of similar studies for bar castings of uniform section. As in the previous report the limits of riser and casting edge effects were determined radiographically. Correlation was then made with the gradient conditions and course of solidification determined by thermal analysis

The radiographic studies were made on 1-in. thick sections removed from the center of the test castings; the x-ray sensitivity was 1½ per cent. Accordingly, complete solidity must be interpreted as the absence of shrinkage to these limits. Since the entire shrinkage of the bar sections is concentrated at the center of the section the method represents a severe practical test of solidity.

The investigation included horizontal and vertical-type castings fed by risers having a diameter equal to 1½ times the casting thickness and a height equal to two times the casting thickness. The metal was gated into the bottom of the riser for both the vertical and horizontal castings. All castings were poured at 2950 F with induction melted Navy Class 3 steel. Analyses of the various heats showed variations in the range of 0.25 to 0.35 C, 0.50 to 0.80 Mn, 0.30 to 0.50 Si. Final deoxidation was accomplished by the addition of 0.10 per cent aluminum to the ladle.

Radiographic Studies

Bar castings having cross-sections of 2×2 , 4×4 , 6×6 , and 8×8 in. were cast in vertical and horizontal positions. Each bar size was cast in a series with variations in lengths equal to half the casting thickness in order to determine the longest bar which could be made sound. The results obtained were verified by casting a second series in which the bars were varied in shorter steps on either side of the critical length determined in the first tests. A summary of all of the tests is presented in Table 1. The casting lengths were measured from the periphery of the riser in the horizontal casting and from the bottom of the riser in the vertical casting.

In all cases, bars which are longer than the critical length were noted to have sound zones adjacent to the riser and to the ends opposite the riser, with shrinkage located between these zones. Increasing the

if eres, ig, ens

lly

we elto vn

ut

its

ot

ng

n-

d

^{*} Metallurgist, ** Head, Metal Processing Branch, Metallurgy Division, Naval Research Laboratory, Washington, D. C.

TABLE 1

	Bar Size, In.	Riser Diam., In.	Dist. to be fee In.	o Plane d, of Casting		Sound Riser Zone In.	, Remarks
-	2x2	3	6	Н			Sound
	23.2	3	7	Н			Sound
	**	27	8	Н			Sound
	**	22	9	Н	4	21/2	Shrinkage
	**	22	10	Н	5	31/2	Shrinkage
	**	**	11	Н	51/2	31/2	Shrinkage
	**	**	12	н	41/2	1	Shrinkage
	,,	**	12	н	31/2	21/2	Shrinkage
	**	**	14	H	3	1	Shrinkage
	**	**	8	v			Sound
	**	12	10	v.	31/2	2	Shrinkage
	9.9	**	12	V.	4	ī	Shrinkage
	**	27	14	1.	4	1	Shrinkage
			1.1	•			Sillinkage
		Average			4.1	2	
	4x4	6	10	H		_	Sound
	**	9.9	12	H			Sound
	9.9	**	12	H		-	Sound
	2.5	22	12	H			Sound
	**	**	14	H	9	2	Shrinkage
	**	2.7	14	H	9	11/2	Shrinkage
	**	4.5	16	H	8	21/2	Shrinkage
	**	15	18	H	8	2	Shrinkage
	22	22	12	V			Sound
	**	9.9	14	V	6	21/2	Shrinkage
	99	22	14	V	61/2	11/2	Shrinkage
	9.9	9.9	16	1.	51/2	21/2	Shrinkage
	10.	2.2	18	V	6	21/2	Shrinkage
	2.0	9.9	20	V	6	2	Shrinkage
		Average			7.1	2.1	
	6x6	9	15	н			Sound
	F2	27	16	Н	10	41/2	Shrinkage
	22	9.2	17	Н	10	4	Shrinkage
	9.0	9.9	18	Н	9	31/2	Shrinkage
	22	**	19	H	10	3	Shrinkage
	**	**	21	Н	9	41/2	Shrinkage
	**	27	15	V	3	1/2	Sound
	**	**	161/2	v	9	4	Shrinkage
	,,	**	18	V	10	5	Shrinkage
	**	99	21	v	101/2	31/2	Shrinkage
	**		24	V	9	31/2	Shrinkage
				•			Sillinkage
		Average			9.5	4	
	8x8	12	16	H			Sound
	12	319	17	H	111/2	41/2	Shrinkage
	**	9.5	20	H	101/2	41/2	Shrinkage
	**	2.0	21	H	11	41/2	Shrinkage
	11	**	28	H	121/2	4	Shrinkage
	A	verage			11.25	41/2	
	8x8	12	16	V			Sound
	219	**	19	V	-		Sound
	22	22	19	V			Sound
	22	**	21	V		1	Sound
	22	99	24	V	14		Shrinkage

Fig. 1—Transverse radiograph of 2x2x16-in, bar section showing shrinkage area and sound end and riser zones.

length of the bar section above that which first produced shrinkage resulted in increasing the length of the unsound zone while the lengths of the sound end and riser zones remained unchanged.

A typical radiograph showing the nature and location of this shrinkage is shown in Fig. 1. The lengths of the sound end and riser zones were measured from the radiographs and are listed in Table 1 to indicate the contribution of end and riser effects to casting soundness.

In section sizes up to 6 x 6 in. the maximum length of bar which can be made sound is the same in both vertical and horizontal planes. An 8 x 8-in. section, however, can be fed to complete soundness for a greater distance in the vertical plane than in the horizontal plane. It will be noted from Table 1 that the sound end zone in the incompletely fed 8 x 8-in. vertical casting also is greater than in similar horizontal castings. The cause of the more effective feeding obtained in this section and casting position is discussed later in relation to differences in thermal gradient conditions.

The maximum bar length which can be made completely sound is approximately equal to 6 times the square root of the casting thickness ($D_{max}=6\,/\,T$) except for the vertical 8 x 8-in. casting. Table 2 lists the various maximum lengths for comparison with the values calculated by the empirical formula.

TABLE 2

Bar Cross Section, in.	Max Measured Length of Sound Bar, in,		Calculated Length 6/T, in.
	Horizontal	Vertical	0
2 x 2	8	8	8.5
4 x 4	12	12	12
6 x 6	15	15	14.7
8 x 8	16	21	17

The maximum lengths of sound bars as well as the average lengths of the sound end and riser zones in bars which contain shrinkage are plotted in Fig. 2 and 3 in terms of casting length and number of thicknesses respectively. The maximum lengths of bars which can be made completely sound as calculated from the empirical formula $D_{\text{max}}=6\,/\,T$ are also indicated.

Figure 3 shows that maximum distance which can be made completely sound, expressed in terms of casting thickness, decreases progressively with increasing section size. The same trends may be noted for the lengths of the sound riser and end zones in bars which develop shrinkage. It may be concluded that riser and casting end effects become proportionately less pronounced with increased section size thus limiting the relative expanse which can be fed to complete soundness.

fo

th

co

en

en

an

tio

The sum length of the sound riser and end zones

of

d

21-

m

te

g

h

h

n,

e

ıl

d

t

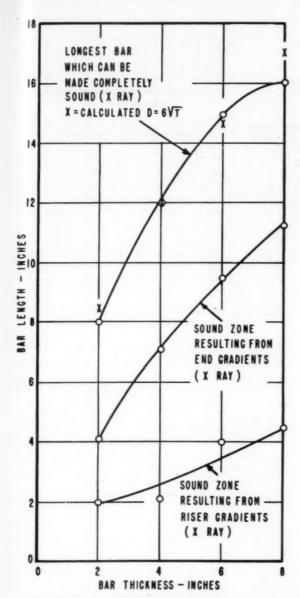


Fig. 2-Soundness relationships for various bar sizes.

for bars which contained shrinkage was generally less than the length of the section which could be made completely sound. The average of the subject difference expressed in terms of thickness decreased progressively with section size from approximately 1T for the 2-in. section to essentially no difference for the 8-in. section. Expressed in terms of distance the difference was approximately constant at 2 in. for the 2, 4 and 6-in. sections.

Thermal Studies

The nature of the thermal gradients developed by the casting and riser determine the course of solidification; hence the feeding conditions during the critical last stages of solidification. The relation of thermal gradients developed in the various bar sections to soundness was determined by thermal analysis methods similar to the procedures used for plate castings in the previous report.¹

Temperatures were determined with 26 gage Pt-Pt Rh (13%) thermocouples, insulated with alundum tubing and protected by fused quartz tubes having an O.D. of $\frac{3}{16}$ in. and an I.D. of $\frac{1}{8}$ in. The thermocouples were projected through the mold into the mold cavity at the desired distances from the end of the casting so that their junctions were at the casting centerline. Temperatures at each position were recorded at 15-second intervals on a multi-point automatic potentiometer recorder of 1500 to 3000 F scale with a sensitivity of 2 F. Liquidus and solidus temperatures were determined by inverse rate cooling curves, plotted on a basis of time required for a temperature drop of 10 F.

The term solidus as used herein represents a "technical" solidus, considered as the temperature indicated by thermal analysis techniques (which are inherently not sufficiently sensitive to detect the last vestige of

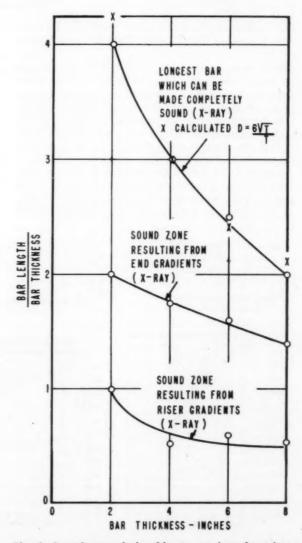


Fig. 3—Soundness relationship for various bar sizes expressed as number of bar thicknesses.

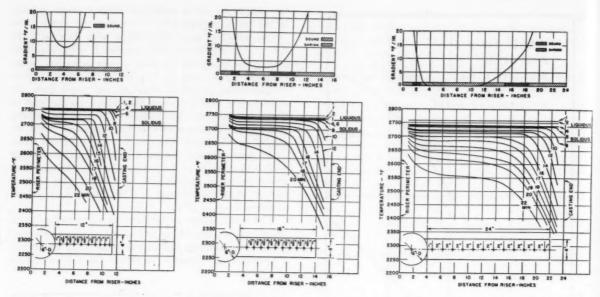


Fig. 4—(Bottom) Gradients along centerline of 4x4-in. bars of various lengths, and (Top) Relationships of

gradients at the solidus temperature to radiographic soundness.

solidification) as the point at which delayed cooling is no longer observed. The "technical" solidus point therefore represents a condition of essential solidity except for thin, isolated films of interdendritic liquid. The determination of an "absolute" solidus in preference to a "technical" solidus was not within the scope of this investigation and was moreover not necessary for the thermal analyses which are to be described.

Figures 4 and 5 show the gradients at various times after pouring along the centerlines of bars of 4 x 4 and 6 x 6-in, cross-section of maximum sound lengths as determined by radiography and of sections which

exceeded this length and contained shrinkage. Figure 6 shows gradient curves for 8 x 8-in. sections which are in excess of the critical lengths for the horizontal and vertical positions. It will be noted that the center-lines of the castings pass through the primary liquidus temperature shortly after pouring without measurable thermal gradients. Gradients, however, are soon established at each extremity of the castings. The risers, because of their mass and because of the re-entrant angle formed with the casting, reduce the rate of heat extraction, while the casting ends, because of the added surface area which they present, increase

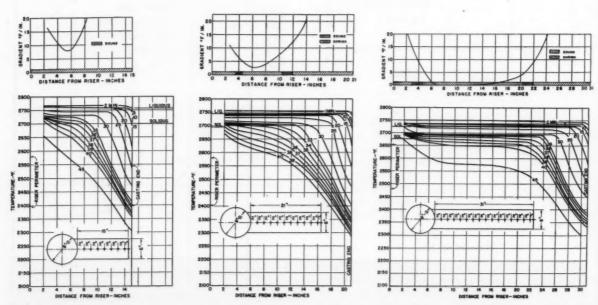
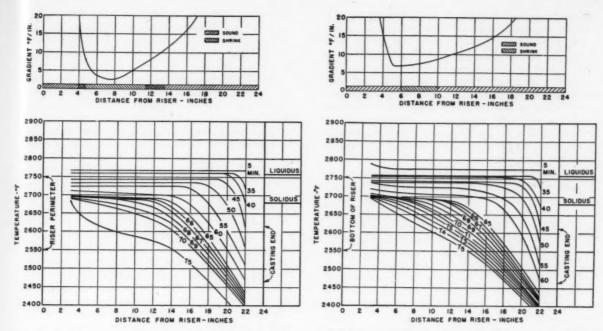


Fig. 5-(Bottom) Gradients along centerline of 6x6-in. bars of various length, and (Top) Relationships of

gradients at the solidus temperature to radiographic soundness.



6-Comparison of gradient conditions during solidification of horizontal (Left) and vertical (Right) bars of 8x8-in. cross-section.

the rate of heat extraction. Short gradients are thereby established near these locations soon after entering the solidification range. The gradients tend to progress toward each other in an attempt to effect a junction; this occurs at the expense of a central uniform temperature plateau.

It will be noted that instead of the usual smooth thermal gradients generally associated with heat flow the gradients from the end of the casting change abruptly as junction with the plateau is approached. The "knee" formed by this abrupt change is due to the necessity of extracting heat of solidification from this zone before the thermal gradients can move inward. The "knee" thus represents a narrow zone of rapid,

progressive solidification.

al

e

e

The inward progression of the end gradients provides the basis for directional solidification conditions. As the liquid metal is cooled below the liquidus, the first solid forms and solidification proceeds at a rate dependent on the rate of heat extraction. Since heat is extracted faster at the end of the casting, solid particles will develop at a faster rate at this position. As the overall temperature level drops the ratio of liquid to solid decreases; effectively there is less liquid remaining to solidify. Since heat of fusion is the primary source of heat evolved during solidification, and since the rate of liberation of such heat is strictly proportional to the amount solidifying at any given instant, it follows that toward the end of solidification less and less heat will be liberated as the amount of liquid remaining decreases. The rate of heat extraction at the casting end, however, does not decrease in such a pronounced manner. It is therefore evident that as the liquid at any fixed point within the "knee" zone is gradually exhausted the thermal gradient at that point must necessarily become steeper. The

"knee" will therefore progress inward past this point to maintain the dynamic heat balance characteristic of the progressive solidification process.

The riser gradients are due to a somewhat different mechanism-flow of heat from the riser to the casting. As such these gradients depend primarily on the temperature differential between riser and casting. Since the riser cools more slowly than the casting this differential increases with time. This effect, together with the increased conductivity due to increased amount of solid in the casting, results in increased rate of heat conduction as solidification progresses; hence with the moving of the gradient inward.

The thermal effect of increasing the bar lengths is to increase the length of the essentially isothermal zones present after solidification starts and to maintain these zones to lower temperatures within the solidification range thus decreasing the extent of overlapping of the riser and end gradients at the time the solidus temperature is reached. The steepness of the gradients in the central regions during the final stages

of solidification is thereby reduced.

The actual values of thermal gradients* at the time the various positions of the central portion of the casting reached the solidus temperature are plotted above the thermal curves. The resulting curves of gradient values are related to the sound and unsound zones determined radiographically (Table 1) by the bar graphs at the bottom of the figures. While the transition point from soundness to unsoundness in bars in excess of the critical length is not clearly defined, as indicated by variations in lengths of the sound zones, the midpoints of the scatter zones

^{*} Determined by tangent slope method.

1

ti

d

lie

0

fa

in

th

re

ch

in

th

gr

tic

ed

ris

tic

cre

co

of

an

(shown by the overlapping of the cross-hatching) correspond to gradients of from 6 to 12 F per in. It can be noted that shrinkage developed whenever the value of the thermal gradients at the critical last stage of solidification fell below the 6 to 12 F per in. range.

Castings of maximum length which can be made completely sound develop a minimum gradient condition of 7 F. It may be concluded that interdendritic feeding at the critical last stages of solidification of bar sections requires a minimum gradient condition of 6 to 12 F per in., at least for the steels of 0.25 to 0.35

C content investigated.

Comparison of the gradient curves for the 8 x 8-in. vertical and horizontal castings poured simultaneously (Fig. 6), indicates that the greater expanse of soundness in the vertical castings is due to its generally steeper gradients. As the result, the vertical bar casting may be made longer before the gradients fall below the critical range for the development of shrinkage. The effect arises from the comparatively slower cooling rate of the near-riser zone. While temperature measurements in the riser itself were not made it is observed that at 3 in. from the riser the vertical casting remains at a higher temperature for a longer period of time during solidification. Steeper gradients are thereby established during the course of solidification of vertical bar castings. Since the 2 x 2, 4 x 4 and 6 x 6-in. sections have the same critical maximum length for complete soundness in both vertical and horizontal positions, it is concluded that the effect is basically related to convection currents which provide for the gradual rising of hot liquid metal to the riser during the slower solidification of the more massive 8 x 8-in. section. This action reinforces the general gradient condition produced by the chilling effect of the casting end and the heating effect of the riser and thus develop a generally steeper gradient condition in heavy vertical bar sections.

General Discussion

Comparison of the results of the present investigation of bar sections with the previously reported investigation of plate sections provides important information as to the effects of geometry factors on the thermal gradient conditions required for feeding to complete soundness. It will be noted that while progressive solidification is always to be expected in the presence of gradient conditions, geometry effects may dictate whether complete solidity will result. It is thus possible to have progressive solidification without complete solidity as is noted in the case of bar sections.

It was reported that plate sections up to a thickness of 4 in. could be fed to complete soundness for a distance equal to $4\frac{1}{2}$ times the casting thickness. If bar sections behaved in the manner of plate sections, a 4-in. bar could be fed for a distance of 18 in. which is much greater than the 12 in. actually found to be possible. On this basis of comparison, the riser feeding radius for bar sections is considerably less than that of plate sections. The relationship of section thickness to feeding distance in bars furthermore is not constant as it is in plate sections but decreases as

the section thickness is increased. The differences in the behavior of plate and bar sections may be due to differences in the thermal gradient conditions required for the critical last-stage interdendritic feeding of the two shapes.

In the case of plate sections interdendritic feeding to complete soundness occurred as long as any measurable gradients were present. It was shown that a short, apparently gradientless* zone of a length equal to the casting thickness could be tolerated without producing shrinkage determined at 11/2 per cent radiographic sensitivity. The extent of the thermal effects of the riser and casting end, as indicated by the distances to which the gradients extended into the casting at the time of final solidification of the short gradientless region, was noted to be 1T and 21/2T respectively. Increasing the plate distance above the maximum size for complete soundness resulted in increasing the length of the gradientless region without effect on the lengths of the riser and casting edge gradient zones. Shrinkage was then observed in the gradientless section. It was concluded that shrinkage developed in regions which behaved as sections of infinite plates; i.e., plates in which solidification proceeded only from the flat surfaces without longitudinal thermal effects from the riser or casting edges.

The inability of bar sections to feed to complete soundness in the presence of gradients below 6 F per in. which are adequate for plate sections indicates an effect of casting geometry which is not related to gradient conditions. A possible explanation is provided by a consideration of differences in the course of solidification of plate and bar sections which affect

last-stage feeding conditions.

Investigations by this laboratory of the lateral wall solidification characteristics of plates and bar sections3,4 have shown that the solidification of steels in sand molds proceeds in such a manner as to develop a mushy condition of intermixed solid and liquid throughout the casting section. Thus in the case of both the bar and plate castings interlaced solid dendritic structures are present at the center of the casting during most of the solidification cycle. Solidification must therefore be considered progressive only in the sense that the center is in a less advanced stage of solidity than near-surface regions. Gradually the surface becomes completely solid and a wave-like travel of a plane of complete solidification moves toward the center of the section; however, during this time the center portions continue to develop increasing amounts of solid by growth of the already present

While the foregoing analysis considers solidification only in lateral planes, it must be remembered that the lateral solidification occurs at varying rates along the casting,—the rate is fastest at the casting end and decreases at locations nearer the riser. Hence, feed liquid from the riser must move lengthwise along the casting section and then laterally through tortuous interdendritic channels in the direction of the walls to satisfy the liquid to solid shrinkage occurring as the

Slight gradients probably existed which were below the 2 F sensitivity of the measuring devices.

result of wall growth. In the last stages of solidification linear rates of freezing become exceedingly rapid due to the very small amount of liquid which remains to be solidified and the increased thermal conductivity occasioned by presence of solid steel. In these respects the solidification modes of bar and plate castings are exactly similar. The channel conditions available for feed liquid transport and the relative amount of feed liquid which must be transported during the last stage of freezing accordingly become primary determinant factors for the solidity of the sections. It is in this important respect that bar and plate sections differ. In the case of plate sections longitudinal liquid flow occurs along a plate-like channel with lateral feeding required only in two directions. In the case of the bar sections longitudinal flow must take place along a single rod-like channel with lateral feeding required in a full 360-degree sector. Figure 7 illustrates these differences schematically.

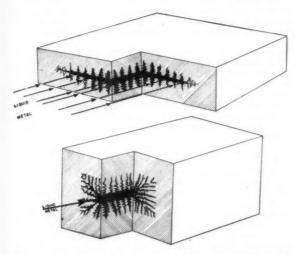


Fig. 7—Schematic of channel conditions for feed metal flow in plate and bar sections.

The increased difficulty of liquid transport along the rod-like channel of the bar section and the greater relative lateral area which must be fed by this limited channel apparently requires steeper longitudinal thermal gradients, i.e., a more "open" condition for feeding as indicated by the thermal gradient measurements. The steeper gradients required for bar sections thus necessitate an "overlap" of the riser and end gradients rather than a simple adjoining which is all that is required for plate sections.

The fixed relationship of section thickness to maximum feeding distance (41/2T) noted for the plates resulted from the invariant relationship between section thickness and the effects of the riser and casting edge. In the case of the bars the relative effects of riser and casting edge decrease progressively with section size (Fig. 3), resulting thereby in a similar decrease for the relative distances which can be fed to complete soundness with increasing section size. It is noted that 2 x 2-in. sections can be fed for a distance of 4T, 4 x 4-in. sections 3T; 6 x 6-in. sections, 21/2T and 8 x 8-in. sections only 2T.

It is observed that the sum length of the sound riser and casting end zones for plates and bars which contain shrinkage is somewhat less than the length of the respective sections which can be made completely sound. The apparent lack of correspondence was tentatively ascribed in the previous report1 on plate sections to hot tearing which originated at locations of shrinkage and extended toward the riser. In fact in many of the plate castings this condition actually connected with the riser proper. In the case of bar sections, however, the tear-like condition was not noted; normal shrinkage whenever present did not extend to the riser thus permitting measurement of the effect. It was noted that in all cases except the 8-in. section for which no measurable difference could be found the discrepancy consistently amounted to approximately 2 in. As in the case of the plate sections no clearly defined difference could be observed from the thermal gradient conditions to explain this discrepancy. A possible, but highly tentative explanation, may be deduced from the noted tendencies for shrinkage to be located in regions having gradients in the range of 6 to 12 F per in. for bars which contained shrinkage while completely sound bars tolerated gradients as low as 7 F per in. It is hypothesized that shrinkage once initiated tends to extend over a somewhat greater area than would be dictated by thermal gradient conditions possibly by partial drainage of interdendritic liquid from the adjoining areas.

References

H. F. Bishop and W. S. Pellini, "The Contribution of Riser and Chill-Edge Effects to the Soundness of Cast Steel Plates," Transactions, A.F.S., vol. 58, pp. 185–197 (1950).
 J. B. Caine, "A Theoretical Approach to the Problem of

2. J. B. Caine, "A Theoretical Approach to the Problem of Dimensioning Risers," Transactions, A.F.S., vol. 56, pp. 492-501 (1948).

3. W. S. Pellini and H. F. Bishop, Discussion of paper by V. Paschkis, Transactions, A.F.S., vol. 58, pp. 151-152 (1950).

4. H. F. Bishop, F. A. Brandt, and W. S. Pellini, "Solidification of Steel Against Sand and Chill Walls," Transactions, A.F.S., vol. 59 (1951).

DISCUSSION

Chairman: J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.

Co-Chairman: H. W. DIETERT, H. W. Dietert Co., Detroit.

V. PASCHKIS (Written Discussion): ¹ The authors' continued efforts to correlate temperature gradients during solidification with feeding and casting soundness are interesting. In studying this paper carefully some questions came to mind, which are discussed below, mainly in way of speculation, in the hope that they may contribute to an understanding to the difficult and abstract happenings during solidification.

1. The authors seem successful in correlating temperature gradients in the casting with the ability to feed properly, and thus avoid shrinkage. The idea behind this concept seems to be as follows: if the temperature is near the surface a certain temperature prevails, and in a closely adjacent section the temperature is slightly higher (according to the prevailing gradient), then this steel of higher temperature is also less viscous (has higher fluidity), and can therefore fill the voids under the given ferrostatic pressure from the riser. The minimum gradient, which is required, would then have to be different for different steels, according to the viscosity-temperature relationship of these steels. Thus—if measurements of viscosity were possible, it would be more direct to speak of viscosity gradients. The inaccessibility of viscosity observations makes the indirect descrip-

^{*}Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York.

F

PNtle

h

ct ba pl

di

b

in

Ci

sa

fu

of

Pe

th

W

tie

ał

th

SE

sp

sta

T

en

be

tion by means of temperature gradients a necessary detour.

2. The gradients shown in the upper parts if Fig. 4 to 6 are taken at the moment when each point in the casting has reached the solidus temperature. In other words the gradients shown in any one graph for different distances do not prevail at the same time. Theoretically the time, for example in form of the rate of change of the gradients, enters the consideration. Are the data published here sufficient, to warrant the statement that for practical purposes the time factor may be disregarded?

3. The solidification process is necessarily three-dimensional in nature. The authors recognize that in the general discussion, but, as far as the discusser can see, disregard it in the main part of the paper. But the three-dimensional nature of the heat flow and solidification pattern may well account for the relationships

shown in Fig. 3.

In addition to these major comments a number of smaller

questions arise.

a. The difference between the length of section which could be made completely sound and the sum of the sound lengths at riser and end is in itself somewhat disturbing; but even more surprising is the variation in this difference: 1.9 in. for the 2 x 2 casting; 2.9 in. for the 4 x 4 casting and 1.5 in. for the 6 x 6 casting. One would expect a steady decrease, particularly in view of the very small value (0.25 in.) for the 8 x 8 horizontal casting.

b. In Fig. 6, right side (vertical casting) the minimum gradient is about 7 F per in. Based on the other discussion in the paper one would then expect, that the entire vertical casting of 24-in. length should be sound, instead of only 20 in. of the

length

c. In discussing Fig. 6, the authors mention the relatively slower cooling in the near riser zone of the vertical casting, compared with the horizontal casting. Here again the three dimensional nature and the geometry enter the picture. The angle between riser and casting in case of the horizontal casting provides for more sand to be in contact with a limited amount of steel, in case of the horizontal as compared with the vertical casting, which probably would explain the reported behavior.

AUTHOR'S REPLY

MR. PELLINI (Written Reply to Dr. Paschkis' Discussion): Dr. Paschkis' discussion of the paper is appreciated. We cannot, however, agree with Dr. Paschkis that metal fluidity has any

significant effect on feeding.

Fluidity is primarily a function of superheat. During the solidification process liquid and solid co-exist at the same temperature and the liquid metal flows through interdendritic channels which are at the same temperature as the liquid. In other words the liquid metal is without superheat regardless of the temperature within the solidification range which is considered.

It is true however that at high temperatures within the solidification range, feeding can occur more easily than at low temperatures. but this is due to the fact that solidification has not progressed as far and the interdendritic channels through which feed metal must flow are larger and offer less flow resistance.

Dr. Paschkis asks if the time factor, as it affects the rate of change of gradients, can be disregarded. We believe that only the instantaneous value of the gradient at any point is important. The 4-in. and 8-in. bars for example each require the same gradient for complete feeding, but this critical gradient is reached in the 4-in. bars in about 25 per cent of the time required for the gradient to be reached at comparable locations in the 8-in. bar.

Heat flow in solidifying castings is, as Dr. Paschkis points out, three dimensional. We have not specifically considered this in the report because we believe that the gradients which we have measured indicate the resultant effect of all the complex heat flow patterns (lateral and longitudinal) existing with-

n the casting.

Dr. Paschkis is puzzled by the fact that the sound zones at the riser and the casting ends in bars which contain shrinkage as shown in Fig. 3 add up to a quantity which is somewhat less than the total length of bar which can be made sound. We have also noticed this occurrence in plate sections. It seems that whenever shrinkage can start there is a tendency for its spreading into areas which would normally be sound. We can only hypothesize that the first shrinkage formed creates a notch con-

dition within the casting which is propagated by cooling stresses as hot tears into the areas adjacent the riser which would otherwise be sound.

Apparently the propagation of shrinkage toward the riser is sensitive to slight variations in casting conditions since the sound riser zones, particularly in the smaller castings, vary appreciably. The length of the sound end zones are more constant.

The vertical casting to which Dr. Paschkis referred actually had a temperature gradient slightly less than 7 F. This is a borderline case where the casting is only a few inches oversize.

We interpret Dr. Paschkis' remarks regarding the effect of the plane of the castings on feeding to be that the angle between the 8-in. by 8-in. horizontal casting and the riser provides far more sand to remove heat from this area than in the vertical casting which would account for the faster cooling rate of the horizontal casting in the vicinity of the riser. Actually the reverse is true because the sand corner at the riser-casting junction must remove heat from both the riser and the casting which would cause the horizontal casting to solidify slower than the vertical casting. The fact that the vertical casting actually solidifies slower is additional evidence that convection currents are active in the 8-in. vertical sections.

J. B. CAINE: The authors of this paper and the staff at the Naval Research Laboratory are to be congratulated for a series of very worthwhile and much needed contributions to the art of founding in respect to risering. They have filled probably the most important gap in our knowledge of risering, feeding dis-

tance

What is the location of the gate on these bar castings?

The authors in a previous paper discussed the fact that they could feed the ½-in. thick plates for a greater distance than the thicker plates because the voids were so small in the ½-in. plates they did not register on radiography. I understand in this series they have radiographed a standard 1-in. section. If my memory serves me right, there is no definite statement in the previous paper in regard to the exact section thickness radiographed.

In this bar series the authors radiographed constant 1-in. sections that were cut from different cast sections. For a definite overall density the individual void might be larger in the 8-in. section than in the 2-in. section. When a slice of constant thickness is cut from the cast section, it is possible that the larger voids in the 8-in. section would be visible on radiography, whereas the smaller voids in the 2-in. section would not. This theory would explain the peculiar square root relation for bar sections, that differs from the constant, section thickness-feeding distance ratio for plate-like sections.

Were the cast sections as a whole radiographed?

Mr. BISHOP: The gates were all located at the riser. In the study of plate sections which we reported last year we did vary gate and riser locations. In some cases we put the gate at the end opposite the riser and it did not seem to make any difference in the feeding distance.

As for the question of variations in void sizes, I do not know that that would necessarily be a factor to be considered. At the center-line of some of the large castings the voids are no thicker than they are in the thinner bars. I think the condition arises towards the end of feeding, that the casting does not know whether it is a large casting or a small casting.

The question of radiographic sensitivity was brought up. We are trying to get castings to the ultimate in soundness and we have not x-rayed the entire casting but only transverse strips.

MR. CAINE: Were all of the slices cut from the plates 1 in.

Mr. Bishor: In the plate sections they were all about ½ in. thick. We did not believe strip thickness would make any difference in the plate because the shrinkage in the plate occurs along a plane, and whether you cut a ½-in. strip or a 1-in. strip you have uniform distribution of shrinkage over the strip. That would not be true in a bar casting.

CHAIRMAN RASSENFOSS: You used a temperature of 2950 F.

At what point was that temperature measured?

MR. BISHOP: In the ladle.

CHAIRMAN RASSENFOSS: In the mold it might have been somewhat less than that?

MR. BISHOP: Yes, it undoubtedly was.

Consultant, Wyoming, Ohio.

is

he

P

1).

ly

of

e.

0-

1e

ie

ly

g

g

y

ts

e

of

CHAIRMAN RASSENFOSS: I am reminded of a discussion which followed presentation of a paper by Duma and Brison a few years ago on centerline shrinkage. The question then was the problem of feeding a plate horizontally and vertically. The Naval Research Laboratory had done work which indicated that a riser attached at the midpoint of the distance on the long dimension of a vertically-cast plate fed the top of the plate and not the bottom. Apparently you have work here which indicates that that is so. As I got it, your reason was that convection currents were the factor which gave a difference between the horizontal and vertical feeding of a 8-in. by 8-in. section. By that do you mean the convection current set up by agitation as the metal enters the mold or something that occurs after the metal is in the mold?

MR. BISHOP: We are speaking of convection in a quiescent liquid. Convection currents do not start to move around in the casting until sometime after pouring of the metal when it has come to rest. We have noticed no indications of convection currents in sections less than 8 in. by 8 in. The 6-in. by 6-in. bar fed to the same degree in both the horizontal and vertical planes and it was not until we went to the 8-in. by 8-in. section that we got evidence of convection currents. By that it is apparent that convection currents become active sometime after the casting is poured.

COMMENTS BY DR. SIMS

C. E. Sims: I would like to know what the evidence was in regard to convection currents. In the paper "Solidification of Steel Against Sand and Chill Walls" by Bishop, Brandt and Pellini contained in this volume, it was brought out that after a casting is poured, the whole center comes to an isothermal condition in a very short time. How could there be a convection current in an isothermal liquid?

MR. PELLINI: First of all I wish to point out that the deduction relative to convection currents represented to us the most reasonable explanation for the conditions noted. We found steeper temperature gradients (from the riser into the casting) for the vertical 8-in. by 8-in. bars than for the horizontal 8-in. by 8-in. bars. There was no difference between the 6-in. by 6-in. and smaller bars. These are experimental data and as such cannot be controverted. The reason for the differences represent an hypothesis and as such may be incorrect. However, we considered this hypothesis likely and reasonable.

We reason that convection currents existing during early stages of solidification may be expected to set up differences in the temperature of the liquid metal depending on its position within the casting. Thus, the temperature of liquid metal and sand in the riser become generally higher initially. We know further from our research on solidification (See "Solidification of Steel Against Sand and Chill Walls" by Bishop, Brandt and Pellini in this volume) that higher temperatures result in delayed solidification not only at early stages but during the entire course of solidification.

Accordingly, we deduce that for sections large enough to allow extensive thermal connection (the 8-in. by 8-in. size) the riser metal and sand are heated appreciably more than would be expected normally. The resultant slower solidification of the riser due to this effect results in a lag over and above that due to the size of the riser. This is exactly the condition which is shown by the increased thermal gradients noted for the vertical 8-in. by 8-in. castings.

MR. SIMS: It is easy to see how there could be convection currents while superheat is being dissipated but not after isothermal liquid prevails. It is possible that the apparently steeper gradient along the center line in the vertically-cast specimens may be only evidence of the eccentricity that would occur in the horizontally-cast bar. Although it is not specifically stated, it is assumed that in both cases the riser was vertical. This would produce an L-shaped casting in the horizontal specimen with a re-entrant angle at the top and an external angle at the bottom. There would be retarded cooling at the reentrant angle and, near the riser, the last metal to freeze would be above the center line of the horizontal specimen rather than at the centerline. This would make the two sets of curves in Fig. 6 different.

Another factor that may have a bearing is that, in this steel, free ferrite crystals could form at the peritectic temperature.

If these fell to the bottom, they could cause an effect akin to a convection current. Again this would cause a symmetrical condition in the vertical and an asymmetrical freezing in the horizontal specimen.

MR. PELLINI: Such is the case—convection occurs only during the stage of superheated liquid. However, it is due to this event that the riser is kept hotter longer and the sand surrounding the riser is raised to a higher temperature. While this event occurs early its effects are felt all through solidification, i.e., to completion of solidification.

COMMENTS BY MR. WENNINGER

C. E. Wenninger: Did you pour the 8-in. by 8-in. sections as fast as you poured the 6-in. by 6-in. sections?

MR. BISHOP: The gate sizes were the same. In case where we were trying to determine the convection currents in the 8-in. by 8-in. sections we had a gating arrangement whereby the horizontal and vertical bars, although in separate molds were poured simultaneously, and at very nearly the same rate and same time.

CHAIRMAN RASSENFOSS: Was anything done in the case of a plate poured horizontally to keep the metal which was poured into the horizontal mold from indiscriminately filling the mold, or was there a method of controlling the pattern in which it spread? Might not the placing of the plate in the vertical direction make it mandatory that the metal would go to the bottom and as it rose create a gradient type heating of the sand. If you poured the mold horizontally the metal would have less of a chance to fill the mold in an orderly manner.

Mr. Bishop: That is possible. In the vertical casting, of course the metal goes down the length of the casting and rises, but from some of these movies on gating it appears that when the metal falls down into a pool of liquid it creates such a tremendous amount of turbulence that I cannot see that the method of gating would have any effect. In fact, that is true on horizontal castings also. Even though the gate is at one end of a casting, the movies show that the metal might run down to the opposite end of the casting and come back to where it started. So I do not think, in these large bars, that the location of the gate and the way the metal runs into a large casting could create much of a gradient condition.

Co-CHAIRMAN DIETERT: I would like to have more information on the quantity of sand against the 2-in. by 2-in, section as compared with the 8-in. by 8-in. casting inasmuch as the quantity of sand wil influence the heat absorption.

MR. BISHOP: These castings all had at least 8 in. of sand around them. The 2-in. by 2-in. bars had more sand in some cases than the larger bars, but as far as the sand was concerned, we felt that the mold was for all intents and purposes semi-infinite. Our studies on the 7-in. by 7-in. bars indicated that if the sand thickness was roughly $\frac{7}{10}$ ths of the thickness of the casting it was, to all intents and purposes, sufficient.

COMMENTS BY MR. FAIST

C. A. FAIST: Getting back to the discussion of convection currents, we are seeing here the later stages of solidification where we have a small pool of apparently isothermal metal remaining. Can the so-called convection currents in this apparently isothermal material be explained in another manner?

I was thinking mainly of the fact that, being a pool of late solidifying liquid material, we have a very finite amount of it left, and could this convection current be explained more in the light of a gravimetric explanation during the final nucleation of the very finite amount of this material left. In other words, we have a process of nucleation and solidification going on and might not this convection be more a process of a very finite settling of particles rather than because of currents generated principally through various temperature gradient levels? I think we are calling it an isothermal pool and yet we are trying to explain convection in the light of temperature gradients, which does not agree.

Might not this generation of currents in this pool be more a process, since we are speaking of atomic amounts now, the generation of these very minute amounts of nucleation and entrance into the picture of the mass effect of the very slight atomic amounts of energy being combined in your process of nucleation and dendrite growth? It might be more mechanical, actually, than thermal.

Assistant Director, Battelle Memorial Institute, Columbus, Ohio,

Dir. of Foundry Res. & Dev., University of Kentucky, Lexington. Research Metallurgist, Burnside Steel Foundry Co., Chicago.

th

tu pr

fe no

sa co

ef

in re tr. de

fo hy

Br

MR. PELLINI: I do not feel it is as complicated as all that. There are two things we have to watch in talking about the development of an isothermal condition. We may have a near-isothermal condition across the section of the bar but not along the length of the bar. This is the very crux of the matter.

the length of the bar. This is the very crux of the matter.

We are talking here in go, no-go conditions as far as feeding is concerned, i.e., in terms of very small differences in the longitudinal gradients required for feeding. The thermal gradient conditions imply in schematic terms that there is a more open condition near the riser. In other words a "funnel"-like channel with the large end at the riser. If I may express the differ-

ence in terms of a "funnel" for feed liquid we have a fine funnel for the feed liquid in the case of the horizontal bar and a broader funnel in the case of the vertical. The thermal gradient curves specifically point that out as being the case.

The process of nucleation mentioned does not fit the conditions. We have a process of continuous growth. The dendrite skeletons are attracting more and more solid to them. The amount of solid is continually increasing. Therefore the effect is one of building up gradually on the originally existing dendrite skeletons not of continuing nucleation.

KINETICS OF GRAPHITIZATION IN CAST IRON

By

B. F. Brown* and M. F. Hawkes**

ABSTRACT

The morphology and kinetics of the various graphitization reactions in cast irons are discussed. In first stage malleablizing, the rate of growth of flake aggregate nodules was found to decrease parabolically with time and to increase with the temperature of reaction. The rate of nucleation was found to increase progressively faster with time and also to increase with temperature. The two rates are interdependent, and it appears to be the rule that an alloying element which increases the overall graphitization rate increases the rate of nucleation in such proportions that the rate of growth of individual nodules is slower than that for unalloyed iron.

The eutectoid decomposition of austenite to graphite and ferrite and also to pearlite was studied in a magnesium-treated nodular iron. The resulting "T.T.T" curve is presented. The rate of the former reaction is shown to be diffusion controlled. The subcritical graphitization of pearlite was studied using the same iron. Rates of growth of ferrite were measured, and a comparison of the effects of reaction temperature on the times required to obtain a ferritic matrix by eutectoid decomposition and by subcritical graphitization is presented.

A possible mechanism (involving carbide stability) for the effects of alloying elements on rates of graphitization is suggested.

I. Introduction

THE VARIOUS GRAPHITE FORMATIONS which occur in iron and steel may be classified according to the reactions by which they are produced; these are illustrated in Fig. 1 and are listed below in the order of decreasing temperature range of the reaction:

This is the proeutectic precipitation which accounts for the "kish" (ASTM type C) graphite flakes of hypereutectic irons, and possibly for part of the spherulites in some "nodular" cast irons.

This, the eutectic reaction, is responsible for most

of the flake graphite (ASTM types A, B, and E) of commercial gray cast irons and for at least some of the spherulites in "nodular" cast irons.

This reaction is the essential part of the first stage annealing of malleable iron, during which the massive carbides are decomposed. It is also responsible for the undercooled (ASTM type D) graphite of some gray irons, for the graphite in most mottled irons, and probably for the "quasi flakes" formed in some partially "nodular" irons. Either reaction II or reaction III is responsible for the spherulites in "nodular" iron.

Since the reaction is thus of considerably wider importance than simply that of first stage malleabilization, the overall term "gamma-range graphitization" will be used here.

$$(Carbon)_{austenite} \rightarrow Graphite$$
 (IV)

This refers to precipitation along the Agraphite line on cooling. The graphite so rejected from solution in austenite almost always deposits on existing graphite,

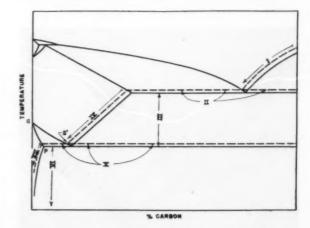


Fig. 1—Iron-iron carbide and iron-graphite diagrams.

Roman numerals refer to the various types of reactions which produce graphite.

al

te

^{*}Research Associate, Department of Engineering Research, North Carolina State College, Raleigh, North Carolina.

^{**} Associate Professor, Department of Metallurgical Engineering; Member of Staff, Metals Research Laboratory, Carnegie Institute of Technology, Pittsburgh, Pennsylvania.

This paper represents part of a thesis submitted by B. F. Brown to the Graduate Committee of the Carnegie Institute of Technology in partial fulfillment of the requirements for the degree of Doctor of Science.

but under rare conditions may form in a Widmanstätten figure.

This, the stable iron-carbon eutectoid reaction, probably occurs in most partially or wholly ferritic matrix gray iron and in almost all ferritic malleable irons as the latter are cooled through the "second stage" of malleablization. Most of the graphite so formed deposits on existing graphite particles. This reaction is the basis for one of the heat treatments used to produce ferritic nodular iron.

This reaction is the one by which subcritical graphitization in steel occurs. It is also one commonly used to fully ferritize a pearlitic matrix in nodular irons, although it will be shown later to be a somewhat longer process than that involving reaction V. It possibly occurs to some extent in ferritic matrix gray and malleable irons.

Presumably the solubility of carbon in ferrite that is in equilibrium with graphite decreases with decreasing temperature. Graphite precipitating from ferrite on cooling would probably deposit on existing particles of graphite formed by one or more of the aforementioned reactions.

It is the kinetics of reaction III (gamma-range graphitization), reaction V (the eutectic reaction), and reaction VI (subcritical graphitization) which is the subject of this study.

II. The Crystallography and Metallography of Graphite

The crystallography and metallography of graphite have been thoroughly described by Morrogh and Williams¹ and will only be summarized here to establish a nomenclature for subsequent discussion. The crystal structure of graphite is layered hexagonal with successive layers displaced with respect to one another.



Fig. 2-Normal flake graphite illustrating reflex pleochroism. Mag. 1000x; Nital etch; Sensitive tint illumination.

The graphite formed by reactions I and II above usually occurs as flakes. These are essentially single crystals which are thick basal sections of the hexagonal lattice. Figure 2 shows such flakes in gray cast iron photographed using sensitive tint illumination, under which their color varies with position from red to green; thus the dark (red) flakes in this photomicrograph appear light (green) if they are rotated 90 degrees. The "warping" of the flakes seen in most gray iron (Fig. 3) apparently results from fine-scale fragmentation of the single crystals.



Fig. 3-Warped flakes in gray cast iron. Mag. 600x; Nital-picral etch.

Spherulites, on the other hand, are built up of a great number of pyramidal crystallites, having their apices at the center, in such a way that the basal planes are essentially perpendicular to radii of the sphere. Such spherulites may form in the solid state (reaction III) under conditions described later, or perhaps in the liquid state (reaction I or II) by the addition of Ce, Mg, Ca, or other agents. Figure 4

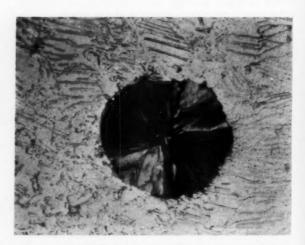


Fig. 4—Spherulitic graphite in whiteheart malleable iron. Mag.—1000x; Nital-picral etch; Sensitive tint illumination.

ON

one gle nal on ler to ro-

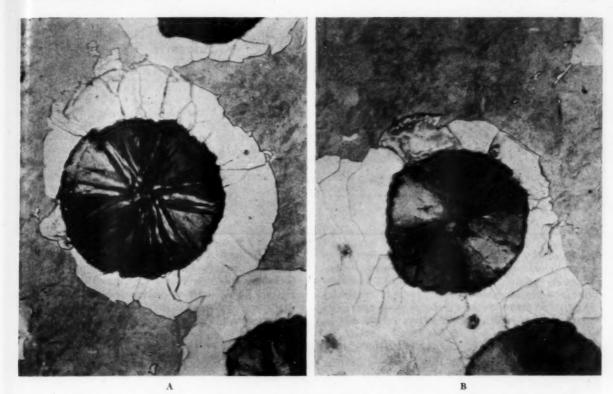


Fig. 5-Graphite spherulites in same specimen of as-cast nodular iron (treated with magnesium). Mag.1000x; Picral etch; Sensitive tint plate.

shows a spherulite formed by heat treatment in the solid state, and Fig. 5 shows spherulites formed during solidification. The crystallites of the spherulite in Fig. 5b are apparently much finer than those of either Fig. 4 or Fig. 5a.

Most of the graphite of blackheart (American) malleable iron is produced by reaction III and forms

Fig. 6-Flake-aggregate graphite nodules in commercial blackheart malleable iron. Mag.-250x; Nital etch.

as clusters of small particles of graphite which sometimes look like flakes (Fig. 6). Such clusters have been named "flake-aggregate nodules" by Morrogh and Williams.¹ Aggregate nodules may vary widely in degree of compactness, as may be seen by comparing Figs. 6 and 9, the less compact nodules being associated in general with rapid nucleation.

It is generally believed that all the graphite which forms along the Agraphite line on cooling (reaction IV) deposits on existing graphite, although this has not been demonstrated. The unusual Widmanstätten deposit of graphite of Fig. 7, however, is an example of

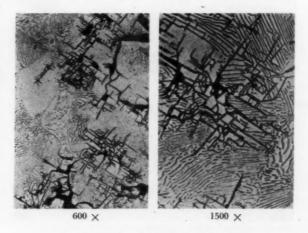


Fig. 7-Widmanstätten graphite (2).

ir th

th be

ca

th to in th

te M

ra ga wi re hi te di

SLI

of

m 1.

th cn th re dr

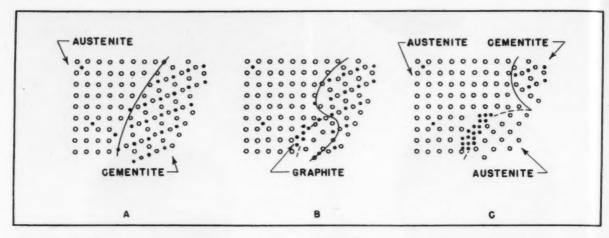


Fig. 8 – Mechanism of gamma-range graphitization (schematic), showing how growth of the graphite lattice (array of solid circles) ultimately must proceed by diffusion of carbon (solid circles) through austenite: (A) at some favorable area between austenite (left) and cementite (right), carbon atoms collect (B) to

form the nucleus of the graphite lattice; the corresponding iron atoms of the decomposing cementite form austenite which soon surrounds the graphite. Further deposition of carbon from the cementite (upper right, C), can only take place by diffusion of carbon through the intervening austenite.

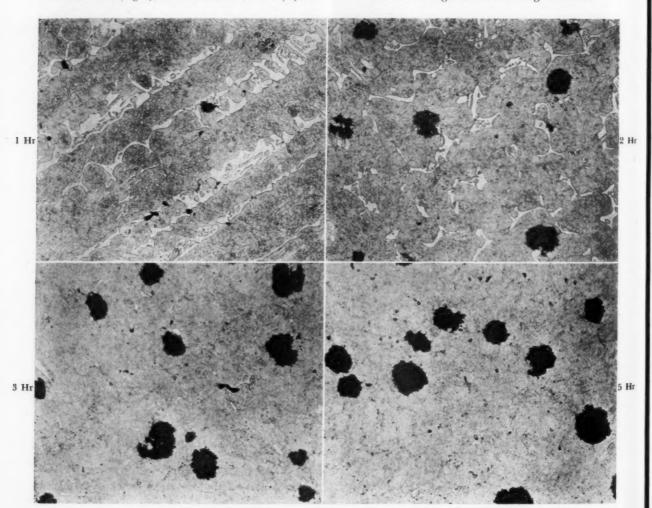


Fig. 9-Progress of gamma-range graphitization; Iron R. Mag.-250x; Nital-picral etch.

such precipitation along crystallographic planes of austenite rather than upon existing graphite.

Although most graphite formed by eutectoid transformation (reaction V) deposits on pre-existing graphite particles, the formation of new particles has been reported by Hayes et al.³ In this instance, many more graphite particles were observed in malleable iron specimens after slowly cooling through the eutectoid range than were there in the austenitic range. The metallographic evidence was not wholly convincing; however, in the current investigation it was noted that in partially malleablized specimens in which the eutectoid reaction was interrupted by quenching, there were a number of examples of what appears to be formation of new graphite particles by eutectoid reaction (c. f. Fig. 24).

Graphite which forms by the subcritical decomposition of cementite (reaction VI) in the absence of existing graphite, as in the subcritical graphitization of steel, usually deposits as an "aggregate nodule," although under appropriate conditions it may deposit along grain boundaries as a quasi-flake. In commercial cast iron (malleable, nodular, or gray) probably most graphite formed by reaction VI deposits on pre-exist-

ing graphite.

2 Hr

5 Hr

III. Gamma Range Graphitization

The process of decomposition of cementite above the eutectoid temperature (reaction III) is one in which graphite nuclei form, and carbon atoms from the decomposing cementite diffuse through austenite to precipitate on the growing graphite particles; during the reaction the iron atoms of the cementite attach themselves to the adjacent austenite lattice. This pro-

cess is shown schematically in Fig. 8.

The morphology of the graphite thus produced is controlled by the composition of the alloy and by the temperature of the reaction. In malleable iron, if the Mn/S ratio is below about 0.8 (weight per cent), the graphite particles grow as spherulites. If the Mn/S ratio is above 0.8, the graphite forms in flake aggregate nodules. The compactness of nodules may vary widely, the less compact being associated with higher reaction temperatures, faster nucleation rates, and higher Mn/S ratios. Very high gamma range reaction temperatures may give rise to graphite particles so disperse that they resemble the flakes of gray iron.

The kinetics of the gamma range reaction was studied using white cast irons in which the Mn/S ratio was kept at about 1. When such irons are graphitized at 1500 to 1750 F (816 to 954 C), the flake-aggregate nodules so produced are reasonable approximations of spheres (Fig. 9); the mechanism of growth, however, is identical with that of commercial first stage malleablization.

Measurement of Rate of Growth of Flake Aggregate Nodules

The compositions of the white cast irons used in this study are listed in Table 1. Cubic specimens 1 cm. on an edge were sawed from identical positions in the cast coupons. The specimens were heated to the reaction temperature rapidly in an atmosphere of dried argon, allowed to graphitize for a measured

TABLE 1-COMPOSITION OF IRONS USED TO STUDY
GAMMA-RANGE REACTION KINETICS

Iron	C	Si	Mn	S	P	Other	
Н	2.35 0.97	0.05	0.090	0.008	_		
E	2.35	0.98	0.07	0.082	0.012	0.17 Mo	
G	(2.35)	(0.98)	(0.06)	(0.09)	(0.01)	0.43 Cu	
F	(2.25)	(0.98)	(0.06)	(0.09)	(0.01)	0.40 Ni	
R	2.4	1.3	0.08	0.08	0.06	_	

All irons were from induction melted experimental heats and were cast in sand molds.

Iron R was from a 2-lb heat.

The others were from a single 200-lb heat divided into four fractions.

time, and then water-quenched.

This heat treatment was carried out with different specimens for various reaction times to obtain specimens representing the entire course of gamma range graphitization to complete decomposition of the cementite. This is illustrated by the photomicrographs of Fig. 9. The procedure was repeated at temperatures over the range 1499 to 1742 F (815 to 950 C).

The specimens were polished, and the diameter of the largest nodules in each specimen were measured. The growth curve for these nodules may be drawn by plotting these diameters as a function of the corresponding reaction time, as in Fig. 10. In this figure the square of the diameter is also plotted, from which it can be seen that the growth curve is approximately parabolic.

It was noted in these specimens that the carbide particles nearest to a nodule dissolved faster than those somewhat more remote, but it is significant that the more remote carbides are in the process of dis-

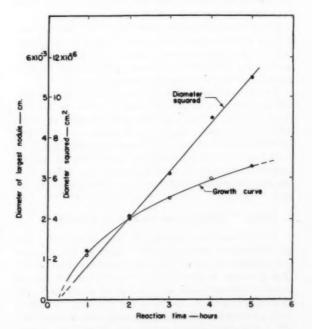


Fig. 10—Growth curve for iron H graphitized at 1661 F (905 C), showing growth time plotted against diameter and against diameter squared.

solving before the nearer carbides completely disappear. This indicates that the diffusion potential surface in the austenite must be as indicated schematically in Fig. 11. Attempts were made to obtain quantitative values of carbon concentration by observing the $M_{\rm s}$ temperatures at various points in the matrix, but the method was not sufficiently sensitive over the small distances involved. The conditions sketched in Fig. 11 indicate that while diffusion is important, it is not controlling.

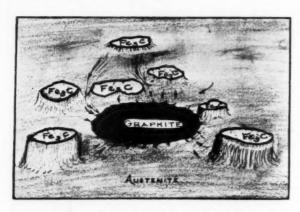


Fig. 11-Carbon diffusion potential terrain in gammarange graphitization (schematic).

Zener⁴ has given an analytical expression for the rate of growth of a spherical particle growing in a matrix originally of uniform composition. This equation is not adaptable to the problem at hand, however, unless some additional empirical term or terms be added to correct for the carbides which exist throughout the matrix and bolster its composition.

Finer carbides dissolve more rapidly because of their greater interface area and smaller radius of curvature. This is undoubtedly one reason why fine carbides, such as are promoted in castings with a small ratio of mass to surface area, graphitize more rapidly.

The growth of a nodule thus is served by carbon diffusing from a large surrounding zone or "carbonshed"; in this region the nucleation of more nodules, while still possible, appears to be somewhat inhibited by the existence of the carbon concentration gradient. Any factor which increases the possibility of nucleation in this zone will thereby compete for carbon with the original nodule and thus act to decrease its growth rate.

2. Effect of Temperature on Rate of Growth

As has long been known, the rate of gamma-range graphitization increases continuously as the reaction temperature is raised. This does not mean, necessarily, that the rate of growth is increased, for the rate of reaction depends also on the rate of nucleation. Increased temperature increases the diffusion coefficient, but increased nucleation, through the competition effect discussed above, would act to decrease growth rate. It is hence not possible to predict the net effect of increased temperature on the rate of growth.

Experiments in this investigation have shown, how-

ever, that in general the growth rate increases with increasing temperature. This increase in growth rate was observed to be discontinuous in the case of iron H (Fig. 12). This is understandable when it is realized that the curve is the algebraic sum of the effects of diffusion and competitive nucleation upon rate of growth.

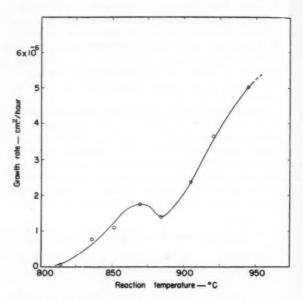


Fig. 12-Effect of temperature on rate of growth of graphite nodules in iron H.

3. Effect of Composition on the Rate of Growth

The elements which affect the graphitizing rate of a white cast iron do so primarily through their effect on the stability of cementite, according to the views set forth in a subsequent section of this paper. The changed stability may result in a change in the rate of nucleation of graphite, a change in its rate of growth, or both. Whether a given alloying element will cause an increase in rate of growth or a decrease depends, as does the effect of temperature, not only on the specific effect on an isolated nodule but also on the rate at which competing nodules are nucleated. Where growth measurements can be made before "carbonshed" spheres impinge, one should be able to detect an increase in growth rate when a graphitizer is added. What happens after impingement effects begin depends on the relative effects of the graphitizer on rates of nucleation and rates of growth, but it appears to be the rule that any alloying element which increases the overall graphitization rate increases the rate of nucleation in such proportions that the rate of growth of an individual nodule is slower than that for the unalloyed iron. As a corollary of this, the elements which speed graphitization decrease the size of and increase the number of final nodules. This effect is especially striking in irons containing nickel, copper, and high silicon.

These conclusions are illustrated in Fig. 13 which shows growth curves for several irons from the same

Fittion 16

DIAMETER OF LARGEST MODULES - CM

sp th ra so the

ra pi of po ot

ter of co.

in

be are car exp graten the

car I ticl

an

tia

d

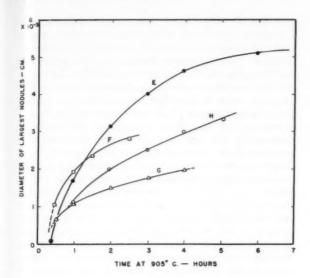


Fig. 13—Growth curves for irons of several compositions from a split heat graphitized isothermally at 1661 F (905 C). E, with Mo addition; F, with Ni addition; G, with Cu addition; H, unalloyed.

split heat. Iron F contains nickel and reacts faster than iron E, which contains molybdenum. At first the rate of growth of iron F is faster than that of E, but so is its rate of nucleation; and as more nuclei form, the rate of growth in iron F drops off until it is the slower.

Iron G is a copper-bearing iron which reacts more rapidly than either E or F. However, because of impingement effects, the rate of growth of the nodules of iron G is slower (at least during the measurable portion of the reaction) than those of either of the other irons.

The effects of alloying elements on graphitization reactions are further discussed in section VI.

4. Rate of Nucleation of Graphite Nodules

The graphite lattice is so dissimilar to those of austenite and of cementite that it is unlikely that any sort of bonding could develop to permit graphite to grow coherently with either of the other phases.

The plane of polish in a metallographic specimen is sometimes observed to cut a nodule which is growing at a carbide-austenite interface. If this were the principal site of nucleation, the probability for nucleation in a given volume of iron would be expected to be a function of the total carbide-austenite interfacial area, which for a given carbon content depends upon carbide size distribution. This would furnish another explanation for the observed fact that a white iron graphitizes more readily if it has first cooled to room temperature to form pearlite and is then reheated into the gamma range than if the iron is allowed to solidify and cool only to the gamma-range graphitization temperature. In the former instance there are many more carbides left from the incomplete solution of pearlite.

Furthermore, in liquid droplets and in solid particles as well, the "vapor pressure" or "solution potential" of an element in the phase is an inverse function of radius of the bounding surface. One would therefore expect smaller carbides, such as those arising from the partial solution of pearlite and the tempering of martensite to be more active in nucleation than the larger eutectic carbides. This has been suggested by Hultgren.⁵ Type D flake graphite in gray irons, which also results from the decomposition of cementite, appears to form at carbide-austenite interfaces too, and never within the austenite. These three effects thus constitute the best published evidence, indirect as it is, on the site of nucleation and suggest that the site is at the cementite-austenite interface.

It was noted earlier that nucleation of additional nodules tends to be inhibited in the "carbon-shed" which surrounds a growing graphite particle. This may explain in part the decreased rate of graphitization observed in the partially decarburized skin of a white cast iron.

The nucleation of a flake-aggregate nodule commences with the nucleation of a flake which grows predominantly by extension of the basal plane. This appears to stimulate the nucleation of other adjacent flakes which tend to be oriented radially about the center of original nucleation. This is favorable to growth parallel to the basal planes.

Deductions concerning rates of nucleation in irons reacted in this investigation were made from studies of the change in nodule size distribution as the graphitization reactions proceeded. True size distributions were obtained by using the Scheil method of analysis^{6,7} on specimens graphitized for increasing times at temperatures yielding spherical flake-aggregate nodules.

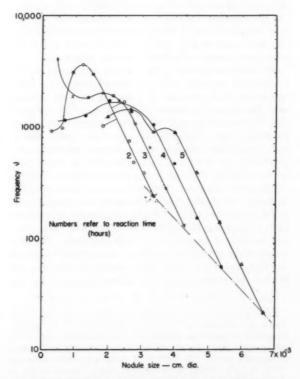


Fig. 14—Graphite nodule size distribution for specimens of iron R reacted isothermally at 1697 F (925 C).

B

be

he

tr

te

tu

of

Si

cle

m

OD

by

th

th

fer

tio

ule

sm

it

WC

CII

do

act

ser

tui

of

Fig

tain

the

get

late

tha

atu

is t

act

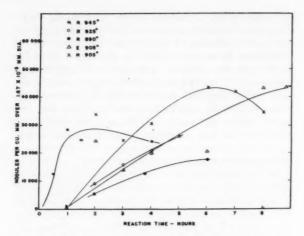


Fig. 15-Effect of reaction time on number of nodules (larger than 1.67x10-3 mm in diameter) during gammarange graphitization. Illustrates disappearance of nodules. Temperature in Centigrade.

A typical example of the distribution curves obtained is shown in Fig. 14.* There are three characteristic regions to such curves:

- A linear branch at larger nodule sizes on the semilog plot.
- 2. A minimum at intermediate sizes.
- 3. A rising curve for the smaller sizes.

One of the interesting features of these distribution curves is the progressive decrease in the number of nodules in the maximum size group as the nodules grow during the reaction. This indicates that all during the reaction competition is entering to effectively stop the growth of a large number of nodules. The measured growth curve is then the growth curve for only those nodules which grow to the largest size.

By combining data from nodule size distribution curves and growth curves obtained from the same specimens, it is possible to calculate rates of nucleation for the early period of reaction. This assumes that during this period all nodules grow at the same parabolic rate. Such calculations were made, and it was found that the rate of nucleation increases progressively faster as time proceeds; this is apparently common in reactions in solid metals. Although smooth curves were obtained, and the trend is undoubtedly as stated above, the results obtained on specimens reacted at various times differed by a factor as great as two, and hence the particular numerical values are not considered significant enough to report. The discrepancy in numerical values is undoubtedly due to the fact that, even in the early period of reaction, the growth of some nodules is slowed because of the

Another interesting observation was made which concerns the effect of reaction time on the number of graphite nodules present during the later period of reaction. Figure 15 shows the total number of nodules (of all sizes above a minimum diameter of 1.67 x 10-3 mm) per cubic millimeter in several irons as a function of reaction time. The number is seen to reach a maximum and then decrease in at least two instances; presumably this would also be true in the remaining instances if the observations had been carried out to longer reaction times. This disappearance of nodules has been reported previously by a number of investigators (e.g., Ref. 13). The explanation is probably that the solution potential of carbon at the surface of nodules of small diameter is greater than that at the surface of nodules of larger diameter. There should thus always be some tendency for smaller nodules to disappear by giving up carbon atoms to diffuse through the surrounding austenite and deposit on larger nodules.

5. Effect of Temperature on Rate of Nucleation

If the rate of nucleation were relatively unaffected by an increase in temperature while the rate of growth increased rapidly, then the first few nodules formed would consume all the carbon and the result would be fewer and larger nodules at higher temperatures of graphitization. If on the other hand, the rate of nucleation were increased relatively much more than the rate of growth, then the result would be a larger num-

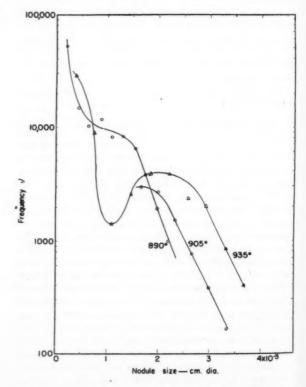


Fig. 16—Comparison of graphite nodule size distribution curves for iron E reacted at three different temperatures. Temperatures in Centigrade.

proximity of neighboring competing nodules.

^{*} In determining any one curve in plots of this type, each nodule is considered to have a diameter which is the nearest integral multiple of the smallest diameter observed. Nodules of adjacent diameters are then tabulated as differing in diameter by one cell size c. The number of nodules in each size group (cell) is thus influenced by c which varies from specimen to specimen. In order to make the various curves comparable, the number of nodules in each size group was divided by c to give the ordinate D (which has the units nodules per cm³ per cm).

ber of very small nodules. The observed behavior, however, is that there is little difference in size distribution of nodules grown over a comparatively wide temperature range; therefore, the effect of temperature on the two rates is much the same. In Fig. 16, which shows the distribution curves from three reaction temperatures, it will be noted that the slopes of the straight line branches are essentially the same. Since the rate of growth is affected by the rate of nucleation and is not the same for all nodules, nothing more precise can be said of the effect of temperature on the rate of nucleation than that it is increased by increasing the temperature of reaction just as is the rate of growth.

This conclusion also appears to be substantiated by the curves in Fig. 15 for iron R reacted at three different temperatures. To be sure, true rates of nucleation cannot be derived from these curves, since nodules below the minimum size, including the many too small to detect, have not been included in the count; it would not be expected, however, that their inclusion would radically alter the relative position of the

curves.

of

c-

s;

0

ly

of

e

d

0

e

n

d

d

1

6. Rate of Gamma-Range Reaction

Having graphite nodule size distribution curves, it is possible to compute per cent reaction as a function of time for the temperatures concerned. This was done for two of the irons studied and resulted in reaction curves of the type shown in Fig. 17. It is observed that the reaction curves for different temperatures can be superimposed by simple lateral shifting of the log time axis.

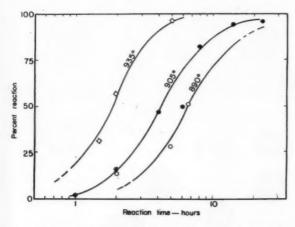


Fig. 17—Gamma range graphitization reaction curves for iron E. Temperatures in Centigrade.

Reaction data for four additional irons were obtained from the literature. These and the data from the two irons of this investigation were plotted together. Again the data would be superimposed by lateral shifting as shown in Fig. 18. It is remarkable that at least over a considerable range of both temperature and composition, the form of the reaction curve is the same. This means that for any given iron reacted at any temperature, it is only necessary to have

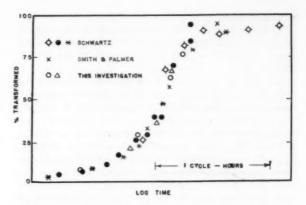


Fig. 18—Master reaction curve for gamma-range graphitization to form flake-aggregate nodules. Irons R and E from this investigation used. Other data from Schwartz¹¹ and from Smith and Palmer.¹²

one reliable point on the reaction curve plus the master reaction curve in order to know the entire course of the reaction.

IV. The Eutectoid Reaction

Austenite of eutectoid composition may, on cooling into or below the eutectoid temperature range, transform either to ferrite and cementite (pearlite) or to ferrite and graphite. Both reactions are observed in cast irons. The reaction to pearlite is found in pearlitic malleable, incompletely decarburized whiteheart malleable, and higher strength gray and nodular irons; the reaction to ferrite and graphite occurs in the production of most ferritic malleable, and in the softer grades of gray iron and nodular iron when the final structure is developed by slow cooling from the austenite range. Control of the factors determining which reaction will take place is thus of prime concern in the manufacture of gray, malleable, and nodular cast iron. The direct transformation of austenite to graphite in steel has never been reported, although there is no theoretical reason why it should not occur if the composition and heat treatment are favorable.

Isothermal transformation studies by Boyles⁸ on gray iron have shown that both the pearlite and graphite eutectoid reactions increase in rate as the transformation temperature is decreased. However, the irregular shape of the graphite of gray cast iron and of blackheart malleable iron precludes measuring the rates of growth and rates of nucleation of the graphite eutectoid reaction in such materials. Furthermore, in whiteheart malleable iron the reaction to ferrite and graphite does not take place; instead the austenite transforms entirely to pearlite (Fig. 4).

On the other hand, the eutectic graphite of nodular iron is in the form of spheres, the reaction of austenite to ferrite and graphite takes place over a wide range of temperature, and the reaction can be controlled so that the ferrite forms in concentric spherical shells around the nodules. This permits measurement of growth of the ferrite shells as well as of rate of reaction. Results of several such measurements follow:

The nodular cast iron used was made by the mag-

nesium process and contained 3.61 C, 2.2 Si, 0.45 Mn, 1.07 Ni, and 0.12 P. By quenching from successively higher temperatures, the upper limit of the eutectoid range (point S' of Fig. 1) was found to be 1520 F (827 C).

To obtain relatively homogeneous austenite of eutectoid carbon concentration, the austenitizing treatment selected consisted of heating to 1742 F (965 C) for 1 hr, then cooling slowly to 1526 F (830 C) and holding for 1 hr. Specimens so austenitized were then isothermally reacted in molten salt at various temperatures for various lengths of time following which they were quenched to room temperature. The specimens were then polished, the thicknesses of the ferrite shells were measured, and the per cent of transformation of the eutectoid austenite to ferrite and graphite was measured by lineal analysis.⁹

The progress of this transformation is depicted in Fig. 19 which shows the structures of specimens quenched from the reaction bath after various reaction times. This figure shows the complete trans-

formation of austenite directly to ferrite and graphite. Since the austenite is of eutectoid composition, ferrite, and graphite precipitate in eutectoid proportions, the carbon diffusing through the ferrite and precipitating on the graphite spherulites. Figure 20 shows the thin layer of pro-eutectoid and eutectoid graphite which has deposited on the original eutectic spherulite.

In addition to the ferrite which forms around the spherulites, there is always a tendency for some ferrite to form at austenite grain boundaries. This tendency becomes more pronounced as the temperature of the reaction is increased, until at the highest temperatures the ferrite is virtually all of the grain boundary type, as illustrated in Fig. 21.

One interesting aspect of the structure in Fig. 21 is that an unusual type of divorcement of eutectoid phases is exhibited. In normal divorcement the two product phases precipitate in massive form in juxtaposition, but in Fig. 21 the two product phases have precipitated in such way that they are separated from each other by untransformed matrix. Figure 22 shows

the

toi ite in yie ite gra fro

fer

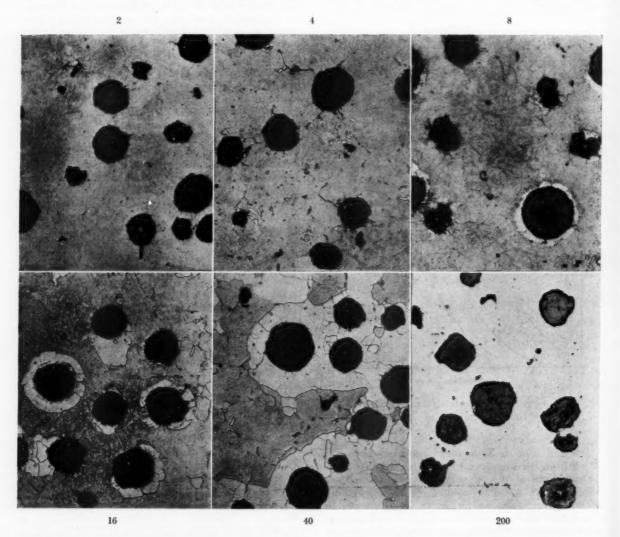


Fig. 19—Process of transformation of austenite in nodular iron P at 1346 F (730 C). Time in minutes. Mag. -250x; Nital etch.

h

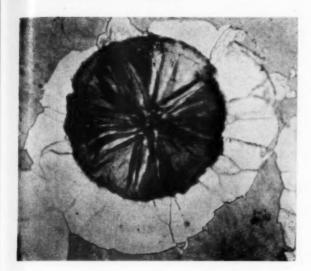


Fig. 20—Spherulite showing outer layer of graphite added during cooling along the agraphite line and during eutectoid transformation. Mag-1000x; Nital etch; Sensitive tint illumination.

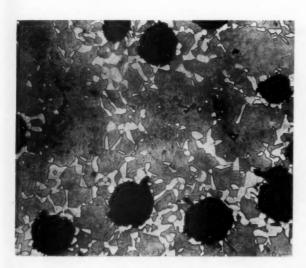


Fig. 21—Grain boundary ferrite in specimen of nodular iron partially transformed at 1400 F (760 C). Mag. —250x, Nital-picral etch.

the mechanism of the reaction which produces shells of ferrite around the spherulites.

Before assuming that all the graphite of the eutectoid divorces by carbon diffusing to the existing graphite spherulites, a careful search was made for graphite in small quantities in the ferrite. An extensive search yielded only one instance of separate eutectoid graphite (Fig. 23). A second manifestation of eutectoid graphite was observed in a commercial iron quenched from the middle of the second stage (Fig. 24). Here apparently some relation between the austenite and ferrite lattices has promoted the formation of graphite at certain interfaces where the diffusion distance to

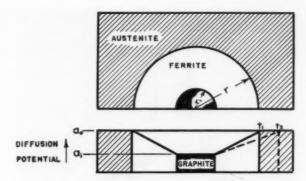


Fig. 22—Mechanism of formation of ferrite shells during the transformation of eutectoid austenite to ferrite and graphite.

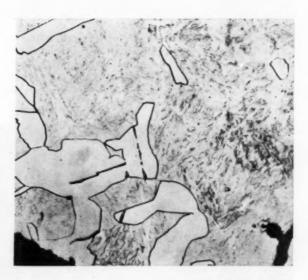


Fig. 23—Graphite deposit (thin dark bands in ferrite) which appears to be eutectoid graphite which formed at a ferrite-austenite interface. Field also includes portion of graphite spherulite and graphite which formed in a crack. Mag. 1600x; Nital-picral etch.

the nearest nodule is too great to accommodate the reaction.

Above 1382 F (750 C), within the $\gamma + a$ + graphite region of the phase diagram for the alloy, the transformation of austenite to ferrite and graphite does not go to completion, but instead the three phases co-exist in equilibrium; the higher the temperature, the less austenite can transform.

At reaction temperatures from 1382 F (750 C) down to about 1332 F (725 C) the reaction products are ferrite and graphite. Below this temperature, reaction to ferrite and graphite begins, but before the austenite has completely transformed, pearlite begins to form and the two eutectoid reactions continue simultaneously.

As the temperature is lowered further, this pearlite interruption commences sooner, proceeds faster, and produces more pearlite. This interrupted reaction is

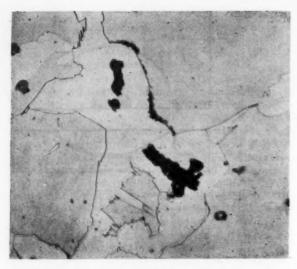


Fig. 24-Formation of eutectoid graphite at an austenite-ferrite interface in a commercial blackheart malleable iron specimen quenched before completion of second stage malleablization. Mag.-750x; Nital-picral etch.

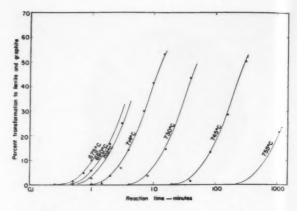
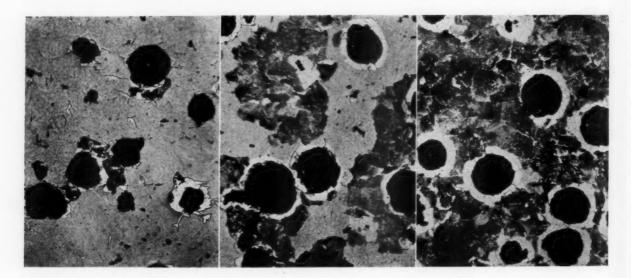


Fig. 26-Ferrite-graphite eutectoid isothermal reaction curves.

Above

Below

Fig. 25-Pearlite interruption of the ferrite-graphite eutectoid reaction in a specimen of nodular iron transformed at 1247 F (675 C). Mag. 250x; Nital-picral etch.



illustrated in Fig. 25. At about 1166 F (630 C) and lower, the austenite transforms almost exclusively by the pearlite reaction.

The amount of transformation in each isothermally reacted specimen was measured by lineal analysis.9 The curves describing the isothermal reaction are shown in Fig. 26. It is observed that reaction curves for various temperatures can be superimposed by a simple lateral shift of the log-time axis. This means that, having a general reaction curve, a single reliable point at a given temperature is adequate to determine the entire course of the reaction at that temperature.

By plotting the time for one per cent transformation ("beginning") and 99 per cent transformation ("end"), versus reaction temperature, a TTT-curve type representation of the reaction was constructed

(Fig. 27). It should be emphasized that the TTTcurve reproduced here represents the particular iron studied only and that such surves are affected not only by composition but also by nodule size distribution and thermal history.

In addition to these transformation rate data, growth rate data were obtained by measuring the thickness of the ferrite shells as a function of time. The resulting curves are plotted in Fig. 28. Conclusions regarding the limiting factors in ferrite growth rate can be drawn from the shape of these curves. This is discussed in the following paragraphs.

If the decomposition of the austenite is limited by a diffusion process in which one assumes a diffusion potential & and a linear diffusion gradient along which carbon moves through the ferrite to plate out REACTION TEMPERATURE

of mai

L

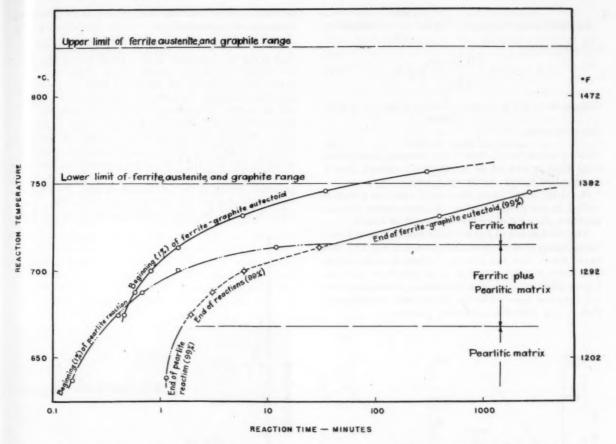


Fig. 27-Transformation diagram for austenite → graphite + ferrite and austenite → pearlite reactions in a nodular iron.

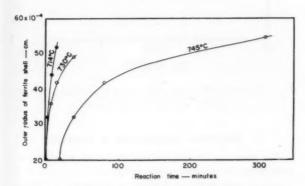


Fig. 28—Growth curves for ferrite shells formed isothermally from austenite.

on the graphite spherulite, then the rate of growth of the spherical ferrite shell can be analyzed mathematically as follows:

Let r = outer radius of spherical ferrite shell (in Fig. 22)

> r₁ = inner radius of spherical ferrite shell, constant within the limits of accuracy of this measurement.

a_o = diffusion potential of carbon in ferrite at outer surface of shell

a_i = diffusion potential of carbon in ferrite at inner surface of shell

m = mass of carbon moving into the shell

D = coefficient of diffusion of carbon in ferrite, assumed constant over the carbon gradient, which is considered linear.

The rate — at which carbon moves through the dt outer shell of thickness dr can be written:

$$\frac{\mathrm{dm}}{\mathrm{dt}} = \left[\begin{array}{c} A \end{array} \right] D \left[\begin{array}{c} \mathrm{da} \\ \mathrm{dx} \end{array} \right] = \left[\begin{array}{c} 4 \pi r^2 \end{array} \right] D \left[\begin{array}{c} a_o - a_i \\ r - r_i \end{array} \right]$$

The eutectoid composition is estimated to be 0.65 per cent C. The decomposition of the eutectoid to form dm grams of carbon is accompanied by the appearance of a mass of ferrite $\frac{dm}{0.0065}$ which appears in a spherical shell of mass $4 \pi r^2 \rho$ dr. Equating these two expressions for the mass of ferrite,

$$\frac{\text{dm}}{0.0065} = 4 \pi r^2 \rho dr,$$

whence

$$\frac{\mathrm{dm}}{\mathrm{dt}} = \frac{0.0065 \, 4 \, \pi \, \mathrm{r}^2 \, \rho \, \mathrm{dr}}{\mathrm{dt}}$$

Equating these two expressions for dm/dt transposing

and integrating,

$$r^2 - 2r_i r = \frac{2D (a_o - a_i)}{0.0065} t + C_o.$$
 (1)

Writing $r^2 = 2r_i r$ as ϕ (r), $\frac{2}{0.0065}$ as β , and $a_o - a_i$ as Δ ,

this reduces to

 ϕ (r) = B Δ Dt + C_o. (2) This equation means that if the reaction is diffusion controlled, a plot of ϕ (r) against t should give a

straight line for an isothermal process.

If, on the other hand, the rate of reaction was controlled by the rate of passage of atoms across the γ - α interface, the plot of r versus t should be linear.

The thicknesses of ferrite shells were measured and the resulting growth curves are plotted in Fig. 28. This figure shows that the plot of r versus t is not linear and hence is not controlled by an interface reaction. Figure 29 shows plot of ϕ (r) versus t, indicating that the function appears to be linear and consequently that it is a diffusion controlled process.

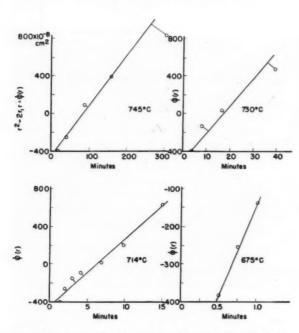


Fig. 29—Plot of ϕ (r) versus time for ferrite shells formed isothermally from austenite at four different temperatures.

The right hand member of equation (2) contains only two terms which are temperature-dependent, Δ and D. D is known to decrease with temperature while Δ , the "instability," increases with decreasing temperature. Whether the rate of growth will be faster or slower at lower temperatures depends upon which of these two has the larger temperature coefficient. The data of Figs. 28 and 29 show that the temperature effect on the "instability" factor Δ is greater than on the diffusion coefficient, and the

growth of ferrite proceeds more rapidly as the reaction temperature is lowered.

Rather accurate estimates of D_c^a can be made, so that Δ can be evaluated. If in a given system this can be treated as simply concentration difference, the value can be plotted on the phase diagram to check for reasonableness. Inasmuch as this is merely a check on the order of magnitude, the plot is made on the binary iron carbon diagram. The two "concentration" values defining this potential are (1) that of the carbon in ferrite in equilibrium with austenite, which is given by the line GP,* extended if necessary, and (2) that of the carbon in ferrite in equilibrium with graphite, which is the ferrite solvus line. In Fig. 30 the values of Δ are plotted from the solvus line. The locus is seen to be a reasonable extension of line GP.

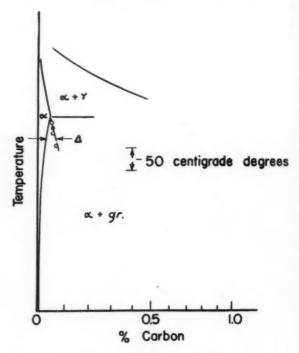


Fig. 30-Diffusion potential values plotted on the binary iron-carbon diagram.

V. Subcritical Graphitization of Cementite

The reaction for the decomposition of cementite at subcritical temperatures is written:

cementite → ferrite + graphite (VI) This reaction is responsible for most graphitization phenomena in steel and is essential for the "full annealing" of pearlitic nodular cast iron and gray cast iron. The expansion which it entails accounts for some, though by no means all, of the growth observed in pearlitic gray cast irons heated to temperatures just below the critical. The reaction could be used in the production of malleable iron, although it is not economical to do so.

Mechanistically, reaction VI proceeds by the de-

Fig.

B.

COL

dif

wi

geo

pre

san

tra

hel

at

roc

gra

in

que

var

gra gra

fre

tur

from grow sphe the the rite

of repear noderapit and havi

[•] See Fig. 1.

composition of the cementite lattice, the carbon atoms diffusing through ferrite to deposit on growing graphite particles, and the iron atoms remaining behind with the ferrite of the matrix.

Experimental Work—Again because of the favorable geometry of the graphite, nodular iron was selected in preference to gray cast iron or malleable cast iron for studying subcritical graphitization. Specimens of the same nodular iron previously used for studying the transformation of austenite were austenitized in dry helium at 1652 F (900 C) for 3½ hr, reacted in lead at 1098 F (595 C) for 5 min, and slowly cooled to room temperature. The specimens then consisted of graphite spherulites in a matrix of fine pearlite.

The specimens were then graphitized isothermally in dry helium at various subcritical temperatures, quenched from the graphitization treatment after various reaction times, and polished for metallographic examination. The structure of the partially graphitized specimens was one of graphite spherulites surrounded by growing shells of ferrite comparatively free from carbide, these shells being surrounded in turn by the untransformed pearlite (Fig. 31).

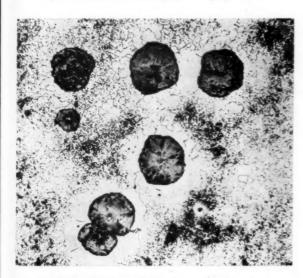


Fig. 31—Subcritical graphitization of pearlite in nodular iron. Specimen heated 40 min at 1328 F (720 C).

Mag.—300x; Nital-picral etch.

No new graphite nuclei formed; instead the carbon from the decomposed pearlite diffused through the growing shell of ferrite to precipitate on the graphite spherulites. Not every graphite spherulite nucleated the reaction at once, a situation parallel to that when the austenite in nodular iron was transformed to ferrite and graphite.

The fact that no new nuclei form under such conditions has important implications in the metallurgy of nodular and gray cast iron. It demonstrates why pearlitic matrices in nodular irons having many small nodules are decomposed to ferrite and graphite more rapidly than in those irons having a few large nodules and why a similar situation will prevail in gray irons having many well distributed graphite flakes as com-

pared with those having a few large flakes.

The formation of the shell of ferrite entails complete recrystallization of the existing ferrite of the pearlite, as well as the addition of iron from the decomposing cementite, in such a way that a radial columnar structure tends to form, as in Fig. 31. While the border between this shell and the undecomposed (but partly spheroidized) pearlite is not a sharp one, by selecting an arbitrary boundary the per cent of decomposed pearlite can be measured with fair accuracy, and growth rates for the spherical ferrite shell can be estimated.

The reaction curves thus measured for specimens reacted at several temperatures are shown in Fig. 32. It should be noted that the reaction proceeds faster as the temperature is increased. Figure 33, in which is plotted the time required to form 50 per cent ferrite by the two methods, shows that above about 1357 F (736 C) (for this iron) the subcritical graphitization of pearlite is the faster.

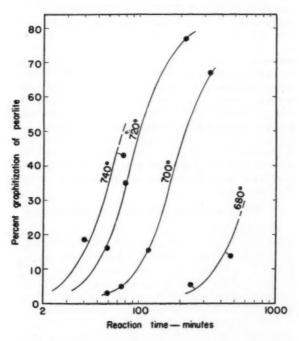


Fig. 32—Isothermal reaction curves showing per cent of pearlite converted to ferrite and graphite. Temperature in Centigrade.

A metallographic study suggests that the cementite nearest graphite spherulites dissolves rather completely before that more distant commences to dissolve, which would indicate that the reaction is diffusion controlled and that the growth of ferrite shells should therefore be described by an equation similar to equation (2) of the previous section except for a new diffusion potential Δ' .

Measurement of the growth of the ferrite shells around the graphite spherulites shows that if $r^2 - 2r_1r$ is plotted against t, as in Fig. 34, the locus is a reasonably straight line. The diffusion potential difference Δ' derived from the slopes in Fig. 34 represents the

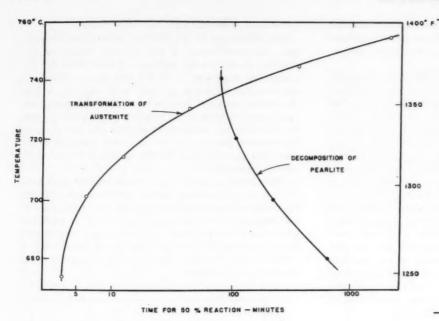


Fig. 33-Effect of temperature on time to graphitize by two different reactions.

TEMPERATURE

difference between ferrite in equilibrium with graphite and ferrite in equilibrium with cementite. These values are plotted in Fig. 35; in this plot, which represents one corner of the iron-carbon diagram, the curvature of the various lines is disregarded. The important point is that the diffusion potential does not vary appreciably over the temperature range studied.

In general the effects of alloying elements on the rate of subcritical graphitization of cementite are qualitatively the same as on the gamma-range graphitization reaction.

VI. Effect of Alloying Elements on Carbide Stability

Alloying elements such as chromium, copper, and sulphur, profoundly affect the rate at which cementite decomposes to form graphite, whether in the gamma range or at subcritical temperatures. Table 2 is a summary of the effect of the more common alloying elements which may be found in cast irons. These effects are so important and often so surprising that a study of graphitization kinetics would hardly be complete without some speculation regarding their causes.

TABLE 2—EFFECTS OF ALLOYING ELEMENTS ON STABILITY OF CEMENTITE

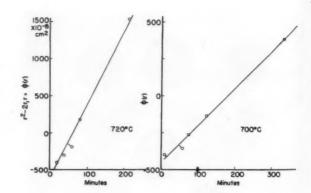
Graphitizers

Co, Ni, Cu, Si, and probably Pt, Pd, Ir

Carbide Stabilizers

Al, Cr, Mn, Mo, Sb, Sn, Ti, Zn, Bi, H, O, S, P, N

Austin¹⁰ concluded that the orthorhombic cementite lattice is essentially one having metallic bonding between the iron atoms, and resonance bonding between each carbon atom and its six near-neighbor iron atoms. The resonance bond holding the carbon is inherently weak and would be further weakened by being stretched. Using this model of the cementite



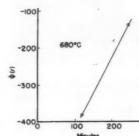


Fig. 34—Plot of ϕ (r) versus time for ferrite shells formed by isothermal decomposition of pearlite at three different temperatures.

lattice, as shown schematically in Fig. 36a, an explanation of the mechanism underlying the effect of alloying elements is suggested as follows:

a. Atoms held by metallic bonding with iron:

If such atoms enter the lattice, distending it presumably by furnishing more electrons in the metallic bonding system, the resonance bond holding the carbon is weakened by stretching and the cementite becomes less stable (Fig. 36b). Examples of such alloy atoms are cobalt and nickel. On the other hand, those atoms which enter the lattice causing it to contract by virtue of reducing the number of electrons in the metallic bonding

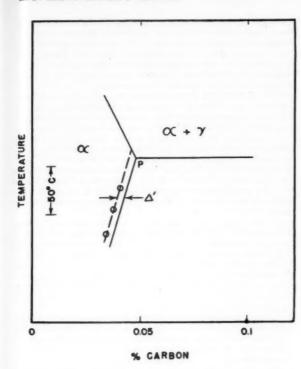
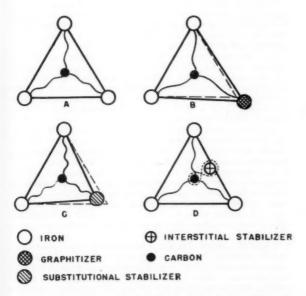


Fig. 35-Corner of binary iron-carbon diagram with plot of diffusion potential values for decomposition of pearlite to ferrite and graphite.



ŧŧ

Fig. 36-(A) Cementite lattice (schematic). (B) Effect of a substitutionally dissolved graphitizer. (C) Effect of a substitionally dissolved carbide stabilizer. (D) Effect of an interstitually dissolved carbide stabilizer.

system, strengthen the carbon bonds and cause the cementite to become more stable (Fig. 36c). Chromium is an example.

b. Atoms which are not held by metallic bonding:

All such atoms which go into solid solution in cementite cause the cementite to become more stable, presumably by forming with the carbon some subsidiary bonding reinforcing the existing resonance bonding (Fig. 36d). Such elements as oxygen, sulphur, phosphorus, nitrogen, hydrogen, and boron are presumed to fall in this category.

It should be emphasized that the foregoing to hypothesized on the basis of very scant evidence regarding the structure of alloyed cementite; many precision studies of the lattice of cementite as affected by small amounts of alloying elements are desirable to check the hypothesis. A point in favor of the theory is that it fits most existing data on the effects of alloying elements on the rate of graphitization. It also resolves the anomalous behavior of certain elements, such as aluminum, boron, and manganese which in small quantities act as graphitizers and in large quantities act as carbide stabilizers. Thus aluminum and boron act first to remove oxygen, a strong carbide stabilizer, and in doing so increase the rate of graphitization. As aluminum or boron is added in larger amounts, the intrinsic effect of these elements, both carbide stabilizers, begins to be felt, and with sufficiently high additions the effect is to increase the stability of the cementite and hence decrease the rate of graphitization. The action of manganese is undoubtedly parallel, except that it combines with sulphur instead of oxygen. It is almost certainly true that other "scavengers" such as Zr, Ce, Sr, perhaps Ti, and many others act the same way.

Qualitatively the effects of alloying elements are in general the same in all the reactions by which graphite is formed, at least insofar as they have been

Acknowledgments

This study was made possible through the grant of a fellowship by the American Brake Shoe Company who also graciously supplied most of the irons used. Appreciation is due Mr. Raymond Schaefer, Director of Research, and other members of the staff in this regard. Other irons were kindly supplied by Dr. Harry Schwartz, Mr. J. E. Rehder, and the International Nickel Company Research Laboratory. The constant advice and encouragement of Dr. R. F. Mehl of Carnegie Institute of Technology is gratefully acknowledged.

References

1. H. Morrogh and W. J. Williams, "Graphite Formation in Cast Irons and in Nickel-Carbon and Cobalt-Carbon Alloys,' Journal of the Iron and Steel Institute, vol. 155 (I), p. 322

2. H. Hanemann and A. Schrader, Atlas Metallographicus, vol. 2, Part 1 (1931), pp. 19-22. Published by Gebrüder, Born-

traeger, Berlin.

3. Anson Hayes, W. J. Diedrichs, and H. E. Flanders, "The Mechanism of Graphitization of White Cast Iron and Its Application to the Malleablization Process," Transactions, A.F.A., vol. 32, p. 601 (1925).

4. Clarence Zener, "Theory of Growth of Spherical Precipitates from Solid Solution," Report on Office of Naval Research

Contract NR 015-018 (1949)

5. Axel 'Hultgren and Olov Edstrom, "Graphitization of Martensite on Heating," Jernkontoret Annaler, vol. 126, No. 3, p. 83 6. Erich Scheil, "Die Berechnungder Anzahl und Grössenverteilung Kügelformiger Kristalle in undurchsichtigen Körpern mit Hilfe der durch einen ebenen Schnitt erhaltenen Schnittkreise," Zeitsch. für Anorg. und Allgemeine Chemie, vol. 201, p. 259 (1931).

7. H. A. Schwartz, "Metallographic Determination of the Size Distribution of Graphite Nodules," Metals and Alloys, vol. 5,

No. 6, p. 139 (1934).

 Alfred Boyles, The Structure of Cast Iron (1947). Published by the American Society for Metals, Cleveland, Ohio.

9. R. T. Howard and M. Cohen, "Quantitative Metallography by Point Counting and Lineal Analysis," *Transactions*, AIMME, vol. 172, p. 413 (1947).

 J. B. Austin, "The Effect of Chauges in Condition of Carbides on Some Properties of Steel," Transactions, ASM, vol. 38,

p. 28 (1947).

11. H. A. Schwartz, "Graphitization at Constant Temperature," *Transactions*, American Society of Steel Treaters, vol. 9, p. 883 (1926).

12. Cyril S. Smith and Earl W. Palmer, "Some Effects of Copper in Malleable Iron," *Transactions, AIMME*, vol. 116, p. 363 (1935).

13. W. D. McMillan, "The Effect of Composition on the Annealing of White Cast Iron," Transactions, A.F.A., vol. 50, pp. 30-38 (1942).

DISCUSSION

Chairman: R. Schneidewind, University of Michigan, Ann Arbor.

Co-Chairman: V. A. Crosby, Climax Molybdenum Co., Detroit.

H. A. Schwarz (Written Discussion): The authors have chosen to present a study of an extremely large field in a somewhat brief form. There are a considerable number of statements here and there in the authors' paper given as factual without very much supporting data and which one can either take on the authors' authority or leave. Such items are, of course, entirely beyond the possibility of discussion unless the discusser possesses original data in the same field.

This discusser hardly agrees with Drs. Brown and Hawkes that their reaction V is the most important one in producing ferritic malleable iron, or indeed ferritic iron of any kind, by heat treatment. However, most commercial operations are designed to use reaction VI in the manufacture of malleable iron. Reaction VI is, therefore, of greater importance to that industry

than the authors suggest.

Reaction VII, although it should exist, is probably of very

little consequence.

The present commentator, perhaps for lack of enough experience, does not attach the same confidence to following the deposition of graphite by metallographic means that the authors do. This is particularly true of nodules of the "flake aggregate"

type as compared with the truly spherulitic ones.

We, in this laboratory, have made some attempt to evaluate the precision of nodule counts in alloys of the malleable iron type. We have thus become convinced that it is necessary to count about 36 square millimeters of actual specimen area at a magnification of 100 diameters, to get an average result whose standard deviation is of the order of 5 per cent of the nodule number. Since we do not know how many fields the authors counted we can form no image of the degree of reproducibility of their results. The mere counting, without classification of size of so large an area is so laborious that we have, in general, adopted Saltykov's generalization that the number of nodules per unit volume is 2.38 times the 1.6 power of the number per unit area. This, however, gives no information as to size distribution.

One of the figures given by the authors shows the numerical distribution of the nodules in terms of particle size as a function of the time of heat treatment from 2 to 5 hr. If the nodules grow during this process it is, of course, reasonable that the longer the time the larger the maximum nodule size.

The authors' curves also show that small-sized nodules have disappeared as time goes on, at least to some extent. It would be rather interesting to see a plot correlating the frequency of occurrence with nodule volume, and hence determining the total volume of nodules as a function of time. Presumably this is the basis for the curve showing an agreement between the authors' data, Smith and Palmers, and the present discusser, as to the form of the graphitization curve. Can the authors give us from their original data of the frequency-nodule size curves, their estimate of the total nodule number as time goes on? This is difficult to get from the published graph because of its logarithmic scale and the absence of cross ruling.

The authors refer to the fact that this figure shows a "characteristic region" having a minimum at intermediate sizes. Of the four curves shown, the 4-hr ones show no obvious minimum nor does the 2-hr one. The 3-hr one shows one only because of the exceedingly high frequency of very small nodules. It becomes a question whether this point is or is not reliable. There is nothing in this paper to permit the reader to form an independent judgment. The 5-hr curve shows a minimum which is not too strongly marked and the commentator himself would say that a smooth curve passing through all the points is more probable than a curve of simpler form curving steadily downward.

In the figure correlating nodule number per unit volume with reaction time in hours with a separate curve for each of five heat treatments, the commentator finds himself in a similar state. The authors conclude that there is always a drop in nodule number although only two of five curves go far enough to show that drop. Here again it is a question of the reproducibility of results whether or not there is a maximum in these curves or whether it merely happens in some cases that the last point has been relatively low.

As an indication of how far off these observations can get, attention is invited to curve H 905° at the 2-hr time limit (Fig. 15). The fact that the number increases in earlier times may be real or may be merely that the number of visible nodules in-

creases although the total number does not.

The authors' demonstration of the increase in nodule number with time is based on what happens originally in graphitization. We have rather repeatedly, in this laboratory, attempted to see if the nodules of malleable iron would be made to coalesce by reheating into the gamma range. We have never found this to be possible by studies of nodule number and size. The results of repeated experiments merely scatter so as to make them incapable of useful interpretation. It is possible that the authors are justified in ignoring Bunin and Danil'chenko's observation that the governing factor in graphitization may be the rate of self diffusion of iron atoms.

In a very brief personal statement Drs. Brown and Hawkes told the writer that they had discussed this feature and rejected it. An expansion of the reason for this rejection, having in mind that the migratory path of carbon need not be the same in length as the migratory path of the iron atom, which has to go away in order to leave room for a graphite lattice to form,

would be highly instructive.

The two authors, in view of their connection with an institution where diffusion problems have been so frequently studied, should be in an excellent position to enlighten us further in a field which seems to be perhaps somewhat complicated.

The preceding discussion is, of course, based solely upon the evidence submitted in the published paper. It is quite possible that in the original doctorial thesis more substantiating data were available than were included here. If that is the case, then the present discussion is not a criticism of the correctness of the views expressed but of the material selected in support of these views.

There is a single categorical statement in the paper which seems particularly glaring in this respect. It is stated that the authors' reaction III is responsible "for the graphite in most mottled irons." This seems to be a direct negation of Boyles' classic work on the formation of primary graphite and is also contradicted to some extent by observation in the discusser's laboratory that mottling tendency is related to the time of the so-called eutectic hold.

It would be interesting if the authors gave their reasons for not accepting Boyles' views with regard to the formation of graphite during freezing.

Dr. Brown's Reply

Mr. Brown (Authors' Reply to Dr. Schwartz): Dr. Schwartz suggests that reaction V is of great importance in producing ferritic malleable iron. This is a point of some consequence, although more so to any consideration of future research than to the present paper. In the course of this investigation, speci-

¹ Manager of Research, National Malleable & Steel Castings Co., Cleveland.

ON

ors'

the

om

eir

s is

ar

Of

um

of

be-

ere

de-

uld

ore

VIII-

me

of

lar

in

igh

ese

ast

ret.

ig.

be

in-

on.

see

by

to

lts

in-

ors

of

es

ed

in

ne

to

m,

d.

a

he

le

ta

se,

rt

ch

ne

st

38

SO

ne

of

mens of one commercial iron were quenched at various stages during a standard blackheart malleabilization anneal. It was observed in this experiment that the deposition of eutectoid carbon took place exclusively by reaction V. It would be very useful to know just in how many cases this is true in industrial practice, and what the evidence is. In reading the literature on this point one should accept with reserve the conclusions from specimens which have been malleablized (partially or wholly), cooled to room temperature, and re-austenitized; for it has been shown that at least in some circumstances the transfer of carbon across the metal-graphite interface is profoundly affected by thermal history.

In making a measurement of true size distribution of nodules by Schwartz's modification of the Scheil method, it was found necessary to measure the diameters of at least 300 nodules per specimen if they were good spheres and 1,000 or more per specimen if they were not good approximations of spheres. Needless to say these measurements are tedious, and they rapidly depreciate if the metallography should be so poor that some of the graphite is pulled out.

The total nodule number (above an arbitrary minimum size)

as a function of reaction time is given in Fig. 15.

The thesis from which this paper was extracted contains 22 size distribution curves. In these, 15 show minima, 4 show points of inflection, and 3 show only straight lines. Hence the generalizations about the characteristic region which is understandably not convincing from the 4 curves available to Dr. Schwartz.

The data in the paper on the total number of nodules per unit vol- as a function of reaction time (Fig. 15) were the results of incidental observations; perhaps they were more fragmentary than really to justify publication. The conclusion associated with them, however, to the effect that the total number of nodules does decrease after reaction is complete is based on three pieces of evidence:

1. That the two curves extending well beyond completion of

reaction show a convincing drop.

2. That in the many specimens studied for other purposes it was observed that the myriad of extremely small nodules, at and below the minimum measurable size, decreases perceptibly in specimens held at gamma range reaction temperatures.

3. Computations from the observations reported by McMillan* showed that out of 10 specimens studied for nodule counts after the gamma-range reaction was completed, 9 specimens showed a decrease in total number of nodules per unit volume, most of them as much as a tenfold decrease.

If extended growth curves on completely isolated nodules were available, it might be possible to demonstrate from those data alone the role (or lack of role) of the self diffusion of iron in governing the reaction. In all the alloys studied, however, the rate of nucleation of interfering nodules was such that significant measurement of G for isolated nodules was not possible. If the rate of self diffusion of iron were governing, however, then one would find (1) that there would be no preference for the solution of carbides nearer the graphite nodules, and (2) that there would be a preference for reaction near a free surface (in the absence of a structure or composition differential). Since these are not observed, it was concluded that although the self diffusion of iron is a participating phenomenon, it is not a controlling one.

Nothing in the paper was intended to be in conflict with Boyles' views on the formation of graphite during freezing. The phrase in question would perhaps be clearer if it read "some of the graphite in most mottled irons, namely graphite in that part of the mottled zone adjacent to the clear chill." Such graphite is frequently A. S. T. M. type D which, as shown by the work of Eash and of Morrogh and Williams, forms by decomposition of cementite. This decomposition may be very fast and may take place before all the eutectic has solidified.

J. E. REHDER: 2 A paper of this type is a little difficult to discuss. The field covered by the paper is very large. Although a great deal of work has been done in the past, I think that the condition of our knowledge can be described as a state of

ignorance rather than a state of knowledge as far as graphitization is concerned.

Dr. Brown has made what I think is a worthwhile attempt to contribute something to it. There are a few points, I think, where I may have slight differences of opinion and some questions I would like to ask. I noticed that Dr. Brown based a considerable amount of the work on growth of nodules and in doing so, chose the size of the largest nodule as a criterion. In our laboratories we have had the opportunity of measuring large numbers of nodules in nodular iron, counting and measuring nodules as they appear in the microstructure and then separating the nodules and recounting and remeasuring the diameters. The important point is that there is no one particular size, the sizes of nodules being a frequency distribution that may be flat or sharp, depending on the conditions of manufacture, which does not lend much support to the use of the largest-appearing nodule as a criterion.

Considerable weight is given by Dr. Brown to diffusion considerations. I think we are all agreed that diffusion plays a major part in graphitization phenomena, but some puzzling evidence exists. If a partially annealed white iron is examined metallographically, it is not difficult to find instances where small pieces of undecomposed carbide extend from the matrix into the side or interior of a graphite temper carbon nodule. If diffusion of carbon were a controlling factor, this would seem to be impossible; therefore, some other factor must be operating.

There is another point. It is a common idea that the rate of graphitization is related more or less to nodule counts. There is no question that there is a relation of some sort but I am reasonably well satisfied at the moment that it is not a simple, direct one. If a white cast iron is quenched from above the critical temperature range before annealing, the annealing rate in the first stage is increased by a factor of two to five times, and the nodule count and surface area of graphite are increased by about 100 and 10 times respectively. However in the second stage of anneal the annealing rate is increased about 100 times, with only a very broad correlation with nodule count.

I think Dr. Brown justifiably used the word "speculation" in the last part of the paper, dealing with the functions of the alloving elements. It is noted that silicon was considered as a graphitizing agent only because of its neutralization of a supposed carbide stabilizing effect of oxygen. Even if oxygen is a carbide stabilizer, it is fantastic to think that a few hundredths of a per cent of oxygen requires 1 to 2 per cent of silicon to neutralize it in the form of SiO₂. Furthermore, I have yet to see a quantitative and verifiable demonstration of oxygen as such as a carbide stabilizer.

T. E. EAGAN: 8 In the first reaction shown, Dr. Brown stated that austenite decomposed into ferrite and graphite. In your statement did you say that Type D graphite was formed at that point?

Dr. Brown: No.

MR. EAGAN: I thought that the graphite was formed just before the iron became solid.

I have seen the TTT curve for nodular iron that Dr. Brown included in the paper. It is about the first one we have seen. We tried to duplicate it using larger pieces and various nickel contents, and the curves that we get confirmed Dr. Brown's curves so well that we did not even publish them. There may be a little shift, one from the other.

DR. BROWN: Distribution measurements were made on all these irons and typical examples of size distribution in various stages of reaction are shown in Figs. 14 and 16.

It is quite true and a very important consideration that some nodules are large and some are small. Presumably the very largest ones-that is, the ones representative of the largest nodules in each specimen-were the ones nucleated first, and that is why they do give a true growth curve for the largest nodules.

Figure 14 is the type of size distribution curve we get where the frequency is plotted to the left and nodular size is plotted to the right. The numbers of those curves refer to reaction time in hours. As Dr. Schwartz says, the number of small nodules goes up very rapidly (note this is log scale); you will appreciate that eventually, when you get down to small enough sizes, you can have a number so large that counting eventually becomes impossible. Both Dr. Schwartz and Mr. Rehder are presumably entering your exceptions without benefit of reading the paper

 ^{*}W. D. McMillan, "The Effect of Composition on the Annealing of White Cast Iron." A. F. A. Transactions, vol. 50, p. 30 (1942).
 *Foundry Engineer, Bureau of Mines & Technical Surveys, Ottawa,

Ontario, Canada.

⁸ Research Metallurgist, Cooper-Bessemer Corp., Grove City, Pa.

the

cas

of

is

pa de W at us us

fl

which shows those distribution curves.

I think I may have somewhat undersold my opinion c. the validity of the explanation for the effect of alloying elements on graphitization. We have much data and sound information on the location of alloying elements in cementite—not nearly as much as we would like to have—but this explanation was not drawn completely out of thin air.

The question of silicon arises. It has been rather definitely shown by Hulgren and by Owen (Journal of the Iron and Steel Institute, vol. 167, p. 117, 1951) that silicon is not located in cementite. I believe there is a good chance it may be written off as a deoxidizer. The evidence to support the idea that oxygen is a carbide stabilizer is not clean-cut, but it is extensive and

hard to rationalize any other way. It would be well to have some direct confirmation.

I believe another statement was misunderstood, in which it was said that the analysis of our data showed that the first-stage reaction is not diffusion controlled, but that the second-stage reaction is diffusion-controlled. That is a rather interesting bit of evidence that Mr. Rehder presents in discussing this, where points out that the change in graphitization time is an order of magnitude different in second stage from first stage.

In the master reaction curve of Fig. 18, the length of a log cycle is labeled in hours. This could also represent a log cycle in any other (linear) time units, as for example the span from 1 to 10 min, or from 100 to 1000 sec.

ILL EFFECTS AND USEFULNESS OF GASES IN METALLURGY

By E. Spire*

ABSTRACT

After discussing briefly the theory of gas absorption in metals, the ill effects of gases like hydrogen and nitrogen are reported. Inert gas flushing is a good method for removing dissolved hydrogen in molten metals. However, the technique used up to now can be improved.

The author has built a type of ladle in which gases can be introduced through a porous refractory material. This ladle can be used not only for degassing by inert gas flushing, but also for other metallurgical treatments when mechanical agitation is desired.

This process seems to offer possibilities for new applications of gases in metallurgy.

THE THEORY OF ABSORPTION OF GASES in metals is now well established. The general principles of it will facilitate the understanding of the rest of this paper. We shall discuss a gas which is considered detrimental in most common metals, i.e., hydrogen. We will also discuss nitrogen, which can be undesirable, but on the other hand, sometimes also quite useful. Finally, we shall discuss how gases can be used successfully in bringing improvement in metals with the help of a new technique. These new possible uses of gases is what we have called "Gasometallurgy."

I-Absorption of Gases in Metals

The mechanism by which gases are absorbed by metals, has been divided by physicists in three main processes: adsorption, diffusion, solution.

Adsorption is the combination or condensation of gases at the surface of solid or liquid metals. Gases are adsorbed before being absorbed! In certain cases, the volume of gas adsorbed by the surface of metals is quite large. To give you an example, we could quote this: a French scientist, Mr. Chaudron, has discovered sometime ago that, by cathodic bombardment, he could extract about 200 cc of gas from 100 grams of a flat aluminum sample, and only 20 cc by vacuum fusion after cathodic bombardment. Most of the 200 cc of gas were gases adsorbed on the surface and not really adsorbed by the sample. Adsorption phenomena are quite complex and have been studied thoroughly.

After being adsorbed on surface, gases travel through metal by diffusion.

Solution of Gases in Metals

Adsorption and diffusion are secondary phenomena. If gases were only adsorbed on the surface of metals, or diffused through, the problem would be rather simple.

The complication arises from the fact that certain gases can actually go into solution in metals. They behave as alloying elements.

There are two ways of studying the solubility of gases in metals:

- 1. Solubility versus temperature at constant pressure.
- 2. Solubility versus pressure at constant temperature.

Solubility of Gases vs Temperature at Constant Pressure

Figure 1 represents schematically the apparatus used by the German metallurgist Sieverts who, around 1900, did remarkable work.¹

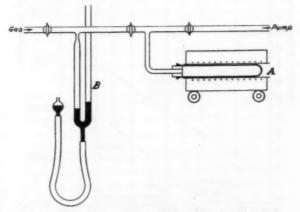


Fig. 1-Apparatus used by Sieverts for the study of gas absorption in metals.

Figure 2 shows three different types of gas solubility curves in metals. Curve A (Fig. 2) shows the case where, at the melting point, there is a sudden

ve

ge ge bit re

^{*}Chief Metallurgist, Canadian Liquid Air Co. Ltd., Montreal, Canada.

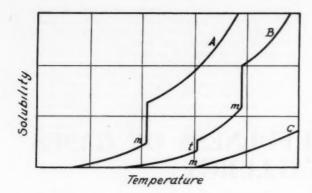


Fig. 2—Typical curves of gas solubility in metals vs. temperature at constant pressure.

increase of solubility. Curve B (Fig. 2) shows the case where there are two breaks in the solubility curve. One at an allotropic transformation point "t," one at the melting point "m." Curve C (Fig. 2) shows the case where the solubility curve shows no discontinuity.

Figure 3 shows the variation of hydrogen absorption in nickel, aluminum and copper with temperature.

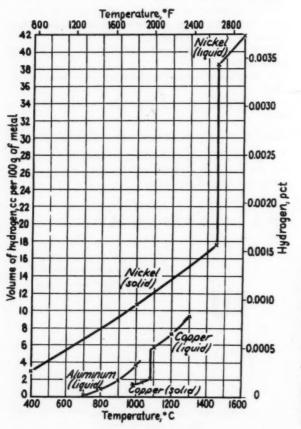


Fig. 3—Variation of the solubility of hydrogen in Ni, Cu, and Al at 1 atm pressure vs. temperature.

Figure 4 shows the variation of nitrogen and hydrogen absorption in iron with temperature.

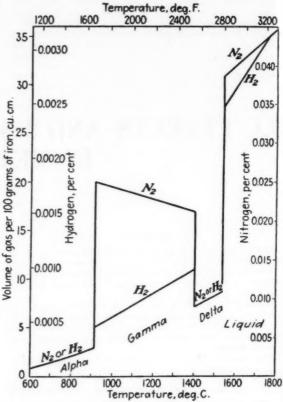


Fig. 4-Variation of nitrogen and hydrogen solubility in iron vs. temperature at one atmosphere pressure.

Solubility of Gases vs Pressure at Constant Temperature

Sieverts found an interesting fact which is now known as Sieverts' law.

$$S = K\sqrt{P}$$

The solubility, S, of biatomic gases in metal is proportional to the square root of the pressure, P, K being a constant which varies with temperature.

Figure 5 shows in the case of hydrogen and various

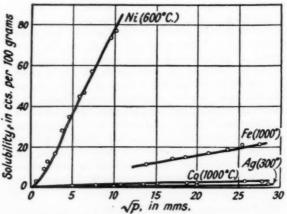


Fig. 5-Variation of hydrogen solubility vs. pressure in Ni, Fe, Ag, Co at constant temperature.

fol pur hy ha

90

E.

of

hy mo no

ap

prodet all fro

siv tio lish of is

has the ficu hu

ysis tha not (is iro

tur

me sep hyc

wat

metals the actual variation of S versus the square root of pressure. It is a straight line.

An important consequence of Sieverts' law is the following: suppose we have, in a crucible, 1000 lb of pure molten copper at 1100 C in the presence of pure hydrogen at atmospheric pressure. This copper will have dissolved around 1 cu ft of hydrogen.

Suppose now, instead of pure hydrogen, we have a mixture containing only 10 per cent hydrogen and 90 per cent of an inert gas. The actual pressure of hydrogen above the copper will be 1/10 of an atmosphere. The quantity of hydrogen in solution will now be proportional to:

$$S = K \sqrt{\frac{1}{10}} = K \frac{1}{3.16}$$

This is roughly, one third of a cubic foot, still an appreciable quantity.

III Effects of Gasses in Metals

In metal making and at metal melting temperatures, the number of gases which we have to deal with is rather limited.

First of all, we have oxygen. Then, we have the products of combustion CO, CO₂, which are never detrimental normally, and water vapor which is usually dissociated; hydrogen which can come from water, from hydrocarbons, from ferro-alloys, and finally nitrogen from the air.

The completely undesirable element in metals is hydrogen followed at a fair distance by nitrogen who

redeems itself by certain good points.

These ill effects of gases have been studied extensively for the last 50 years. We should, however, mention that the various papers of Sims and Zapffe published in the last decade have been our main source of information. The work of these two metallurgists is particularly remarkable in this field.^{2,3,4,6}

Influence of Hydrogen in Metals

It is probably only during the last 15 years that it has been recognized how bad hydrogen was. One of the reasons is that hydrogen analysis in metals is difficult. It is a matter of a few cubic centimeters per hundred grams, and that does not weigh very much.

Twenty years ago, oxygen was considered detrimental. Everybody was concerned about oxygen analysis in metal, especially in steel. We think that at that time, piping pure oxygen in a steel bath would not have been accepted. Now, it is standard practice.

Coming back to hydrogen, you must know that it is undesirable in most of the common metals like iron and steels, copper and its alloys, for two reasons:

1. Its solubility varies a great deal with temperature, and suddenly.

2. It does not form hydrides easily.

Take oxygen, or even nitrogen. You have in molten metals the formation of oxides, even of nitrides, that separate out. You do not have that in the case of hydrogen.

Sources of Hydrogen

The origin of hydrogen in molten metals is usually water vapor. This water exists everywhere:

- -Water of crystallization
- -Water in the refractories, the fluxes, the fuels
- -Water in the products of combustion
- -Water in the air.

Even a fairly large amount of hydrogen can exist in ferro-alloys like ferromanganese or ferrosilicon.⁵

Bardenheuer has pointed out that there is a great amount of water contained in rust: 1 per cent addition of rust to a 10-ton bath of steel introduces about 900 cu ft of hydrogen, measured under standard conditions (20 times the volume of metal).

At the temperature of molten metals, H₂O is dissociated in hydrogen and oxygen according to the equation:

 $2 H_2O \rightleftharpoons 2 H_2 + O_2$

This reversible reaction is governed by the law of Mass Action.

Example: In a closed vessel evacuated from air, suppose we have molten electrolytic iron held at a constant temperature. Suppose now, that we introduce in this vessel a small quantity of water. It will immediately be transformed into water vapor. If the iron was beforehand very oxidized, the oxygen formed in the above reaction will not be able to combine with it; accordingly, the reaction will not proceed very much to the right, and very little H₂ will be liberated to go in solution in the iron.

If, on the other hand, the iron were not oxidized at all, as soon as the $\rm O_2$ from the water is available, it will combine with the metal; more water will decompose with the result that more and more hydrogen will be available and go in solution into the metal, and cause troubles at the solidification.

The practical result of this is important. The more deoxidized a metal is, the more prone it is to absorb hydrogen from water decomposition.

This is true of many common metals. It is why the handling of metals is especially difficult when they are ready to pour, i.e., fully deoxidized, and also why metals are usually melted in oxidizing conditions to avoid hydrogen absorption.

Influence of Dissolved Hydrogen on Solidification

If the solidification of ingots were taking place slowly, there would be no problem. In practice, it is different. During the cooling cycle, there is a progressive increase in the concentration of hydrogen in the core that solidifies last, up to a point, where there may be a sudden expulsion of this gas which will either expel what remains of molten metal, or form blowholes.

It is mainly in the case of ferrous alloy ingots that a late exudation of liquid metal, called bleeding or rising, is taking place, due to the vertical drop of solubility of hydrogen (not to be confused with the carbon-oxygen reaction of rimmed steel).

In the case of non-ferrous metals, the separation of hydrogen will produce unsoundness in the form of blowholes or pinhole porosity.

Influence of Hydrogen on Solid Steel

For many years, a good explanation has not been given for shatter cracks or flakes. In 1935, a German

E

pi

in

th

lis

ga

Hudrogen content of melt relative volum

metallurgist, Houdremont, discovered the conditions under which flaking occurs which led to the discovery that hydrogen was really responsible for it.

We can put it this way: if during cooling the concentration of hydrogen has not been sufficient to produce rising or bleeding, there may be enough gas remaining in the metal to separate out in the solid state. This remaining hydrogen will produce locally tremendous inside pressure exceeding the ultimate strength of the metal, and will actually disrupt it internally on certain spots. Shatter cracks or flakes, or fish-eyes have a white appearance on a fractured piece of steel.

Zapffe and Sims have explained a few years ago why these flakes appear bright in contrast to the surrounding silky grains, by the fact that during the rupture of the specimen there is no deformation of the grains along the faces of the cracks, while on the contrary, the surrounding grains are plastically deformed.⁶

In order to avoid shatter cracks in forging, slow cooling after annealing is recommended to allow slow diffusion of hydrogen out of the metal. However, the larger the section the less efficient this treatment is. The latter is used for certain types of steel forgings and also for rails. The first thing to do is either to avoid hydrogen absorption in the liquid metal or to remove it just before pouring. How this can be done efficiently will be discussed later.

The Nitrogen Story

In aluminum and copper alloys, nitrogen is almost completely unreactive. In liquid iron and steel it is different. As shown previously (Fig. 4), the solubility curve of nitrogen in iron is similar to the solubility curve of hydrogen. Consequently, if liquid steel contains too much nitrogen which is not taken care of by some nitrides forming elements at the solidification, blowholes and bleeding can appear easily. We have made this observation several times. However, in cast iron, we did not find the same tendency.

In solid steel, a minute amount of nitrogen is responsible for strain aging which is a definite increase of hardness of cold worked material like low carbon steel after a period of storage at room temperature.

The rock-candy type of fracture is also attributed to nitrogen.⁷ However, nitrogen is not always an undesirable element. It is being used now in certain type of low carbon steel to increase its hardness.⁸ It is also used in various types of corrosion resistant chromium steels with the following results:

Increase in hardness and strength with no reduction in elongation and toughness.

-Decrease in grain size.

 Increase in resistance to grain growth at elevated temperature.

The mechanism of these improvements is only hypothetical. It is probably due to the formation of chromium nitrides or iron-chromium nitrides. These nitrides are readily soluble in steel at high temperature and act similarly to carbon.⁹

II—Usefulness of Gases

Surprising enough, it is possible to eliminate, at

least partially, the influence of gases like hydrogen in liquid metals by a rather simple procedure called "flushing." It consists in bubbling through the molten metal, another gas preferably inert. As the bubbles pass through the melt, hydrogen evaporates into them and is carried away just as water evaporates in atmosphere when the humidity is low. Eastwood in his book 10 said: "the function of the bubbles is merely to provide a melt surface to carry the released gas out of the melt."

In order to go into more details, suppose for a moment that you have a single bubble of argon inside your molten metal containing a certain amount of dissolved gases and especially hydrogen. The hydrogen existing in the neighborhood of this bubble diffuses through the metal at high temperature rapidly, will be adsorbed at the outside surface of the bubble and penetrate inside of it until its partial pressure, in the bubble, is in equilibrium with its partial pressure in solution. However, as said before, the partial pressure of the gas in the metal decreases according to Sieverts' law; consequently, at constant temperature, the next bubble will take out less gas than the first one and so on. However, if during the flushing operation there is an appreciable temperature drop, the amount of hydrogen removed by each bubble may remain constant and even increase.

In other words, at constant temperature the amount of hydrogen removed by flushing with an inert gas is bound to decrease as a geometrical progression; at continuously decreasing temperature, the conditions are quite different due to the decrease of hydrogen solubility with temperature. You can obtain a good result in the flushing operation only if you expel more hydrogen molecules than the number that is coming in. It is a kind of race between the degassing due to flushing and the gassing due to outside influence. This possibility, we think, has not always been taken into consideration too seriously.

The flushing technique, as a matter of fact, is not new.

The carbon boil is nothing else than flushing out hydrogen from a bath of steel with CO.

However, using gases other than CO formed by a chemical reaction, is relatively new.

Tullis did the first work in aluminum alloys in 1928¹³ and Allen in copper in 1930.¹⁴

Swinden and Stevenson,¹¹ in the early 30's used argon in steel. Their striking results have been reproduced in numerous papers. Since then, a number of other metallurgists have been working along the same line, and among them the names of Baker and Child,¹⁵ A. J. Phillips,¹⁶, Bever,¹⁷, Boak and Allen,¹⁸ Hulme,^{19,20,21} Kurzinski,²² Crockett,²³ Eselgroth,²⁴ Geller,²⁵ Baer and Loring,²⁶ Grand,²⁷ should be mentioned.

Flushing is being used now, to a certain extent, perhaps more in the non-ferrous than in the ferrous field, because the technique is simpler. Many gases can be used provided they are not contaminated by hydrogen or water vapor. Dry nitrogen is suitable for copper and aluminum alloys, chlorine for magnesium alloys; argon and helium for ferrous metals. Dry air can be used also in certain cases provided the

en

led

ten

les

em

OS-

his

ely

gas

in-

nt

ro-

if-

ly.

ole

re,

es.

al

ng

a-

ie

ıg

p,

le

nt

as

at

18

11

d

el

is

H

t

a

presence of oxygen and nitrogen is not objectionable. The type of gas used and its composition may have a great influence on the results of the degassing treatment.

Influence of Size of Bubbles on Efficiency

The diffusion of H₂, out of the bath, is of course proportional to:

a. Total surface of gas bubbles used for flushing.

 Time of contact of these bubbles with the metal to be degassed.

It is easy to demonstrate that for a same flow of the flushing gas, the total surface of the bubbles is inversely proportional to their diameter. The smaller the bubbles, the larger the total surface.

In order to flush a metal efficiently, you must divide the gas as much as you possibly can.

Battelle Memorial Institute Work

In 1946, an extensive research was started at Battelle Memorial Institute, sponsored by the Air Reduction Co. This work conducted by Sims, was published, ¹¹ and is extremely interesting, because it gives quantitative data on the influence of flushing, in degassing steel with inert gases.

Fig. 6-Results obtained by Sims on the removal of hydrogen in steel by argon flushing.

The Battelle Memorial Institute was well equipped to do this type of work, having at hand an elaborate set-up for hydrogen analysis.

Figure 6 represents some of the results of this author which are for us of fundamental importance. Sims has proven by actual quantitative analysis that the hydrogen content of a steel can be reduced appreciably by flushing with argon.

Sims also proved that shatter cracks obtained by artificially increasing the amount of the hydrogen in an SAE 4330 steel 3-in. billet can be prevented by flushing the molten metal to lower the hydrogen content below a critical value.

III-Injecting Gases in Metals

After carefully studying most of the papers published on this subject, we noticed that no satisfactory technique has been found for injection of the flushing gas into molten metals.

Almost everybody was using a pipe protected or not with a refractory coating. Some have closed the end and drilled small holes in it; some have used a porous graphite plug at the end of a graphite tube, which is certainly an improvement but not a simple device to use in a large bath of metal.

None of these methods are really practical for various reasons, especially when treating steel.

We think that in order to achieve the maximum removal of hydrogen in a molten bath of metal, the following rules should be observed closely:

 The degassing operation should be made with a finely divided gas.

2. Degassing should be made immediately before pouring into the molds, preferably in a ladle and certainly not in the furnace.

3. Degassing will be more efficient if, during the operation, the metal is cooling down because the solubility of hydrogen is decreasing.

4. It is when metals are deoxidized that they should be subjected to a degassing operation because their sensitivity to hydrogen absorption is maximum.

We believe that it is because a good method for flushing metals was not available that this process, now known for over 20 years, has not been applied more generally. Many have tried it, but not doing it at the right time and at the right place, the results were inconclusive.

Thinking it over, we came to the conclusion that if it were feasible to degas metals in the pouring

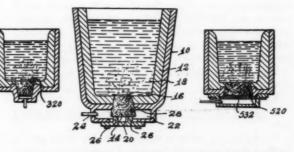


Fig. 7-Various types of ladles with porous refractory plug for gas injection into molten metals.

B

of

ca

T

pa

al

tro

he

ladle by introducing the flushing gas through the refractory bottom of this ladle, we would have the best possible conditions.

We designed such a ladle represented in Fig. 7, as it has appeared in our patent applications covering this new process of injecting gases in metals. We call this process the "gasal" process.

Not knowing too much about refractories, we were greatly helped by the Canadian Bureau of Mines in Ottawa, to whom we want to express our gratitude for the facilities offered to us for our work.

Dispersion of Gas by Porous Refractory

To show you how different the diffusion of a gas can be when injected in a liquid through a tube or through a refractory, we made the experiment in a tank of water as represented in Fig. 8. It can be seen that the reduction of the size of the bubbles is important. Consequently, for the same flow, the total surface of the bubbles, which is what counts, is increased tremendously.

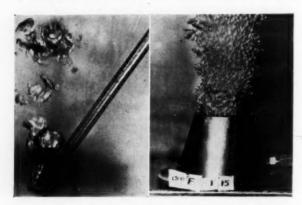


Fig. 8—Experiment showing how different in sizes gas bubbles can be if injected on a liquid through a tube or a porous refractory medium.

The refractory plug used now is completely impervious to metals although, as shown in Fig. 9, fairly large flow of gas for its size can pass through it. It is 33/4 in. in diam at one end, 5 in. in diam at the other end and 31/2 in. thick.

Other Applications of the Process

We soon realized that this type of ladle had other applications due to the fact that with a fairly small flow of gas, any type of boil was obtainable. It must be remembered that at the temperature of molten steel, the volume of any gas introduced in the metal increases around seven times by thermal expansion.

The list of some of these applications is given below:

- 1. Degassing.
- 2. Treat chemically a molten metal with an active gas.
- 3. Agitate a molten metal with an inert gas in contact with a specially prepared slag.
- 4. Agitate a molten metal with an inert gas in order to activate the solution of ferro-alloy additions.

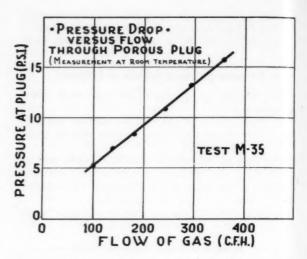


Fig. 9-Variation of the flow of air passing at room temperature through a porous refractory plug with increasing pressure.

5. Homogenize chemically and thermally a heat of metal.

One can also introduce various gases in succession. All these possible treatments constitute the new field which we have called "Gasometallurgy," entirely open to new researches.

Applications of Gasal Process

1. Cooling Effect of Flushing Gas—Injection of gas through a porous refractory medium does not increase the cooling speed of metal. To prove this, 300 lb of iron were treated in our ladle and the cooling cycle recorded with a thermocouple. The rate of drop of temperature is shown in the following table.

Temperature at start, F	Experimental conditions	Rate of drop in Temperature		
2440	No gas flow	21.5° F/min		
2510	No gas flow	24.5° F/min		
2460	Impure argon 10 liters/min	21.5° F/min		
2460	Pure argon 10 liters/min	19.0° F/min		

There is no noticeable difference in the cooling speed of this amount of metal with or without gas injection. Of course, in a larger ladle, the conditions may be different because the metal will normally cool much slower.

2. Degassing Aluminum Alloys—Several heats of 4.5 per cent copper aluminum were degassed by flushing with various gases, especially argon or nitrogen, either with a 1/4-in. diam tube or in the ladle herein discussed.

About 40 lb of metal were melted in a gas-fired furnace and loaded with hydrogen with wet lime pushed under the surface with a plunger.

We found that the speed of degassing was in favor of our porous ladle as shown on Fig. 10 which plots density versus flushing time.

3. Degassing Copper Alloys—Similar tests were conducted on a B-61 alloy containing (in percentage): Cu-88, Zn-4, Sn-6, Ni-1, Pb-1.

h

of

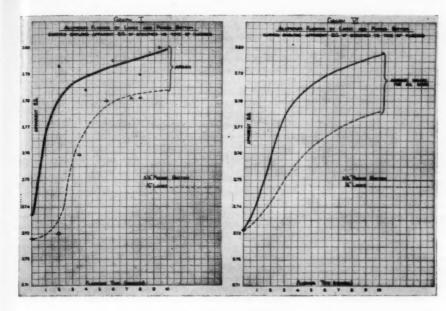


Fig. 10-Variation of density of aluminum alloy containing 4.5 per cent Cu vs. flushing time using either a tube or a porous bottom ladle.

One heat was cut in half, one portion of it being submitted to a degassing treatment in our ladle for 21/2 min while the other portion was not treated. Both portions were cast into keel blocks at the same temperature.

The following results, which are typical, were obtained:

No. of sample	Ultimate, psi	Yield, psi	Elonga- tion, %	Density	Treatment	
286	42,500	18,625	625 35.5	8.78	No degassing	
284	42,875	20,500	37.5	8.79	No degassing	
287	43,000	18,125	57.5	8.89	Degassing	
285	44,750	18,000	65.	8.89	Degassing	

There is a remarkable improvement in the ductility of this alloy due to the efficient method of degassing. The grain size was much smaller too.

4. Degassing Ferrous Alloys—Applied on several occasions on various heats of carbon steel or iron, the gasal process was always successful provided argon was used and not nitrogen. We found that when nitrogen was used the metal was full of blowholes. The nitrogen pickup was quite noticeable. In one particular instance of heats having the following analysis (in per cent): C, 0.75; S, 0.75; Mn, 0.40; P, 0.030; S, 0.050, the oxygen and nitrogen content were analysed by the U. S. Bureau of Standards.

The following table gives the results:

Test No.	Sample No.	O ₂ % Wght.	N ₂ % Wght.
M 7-1	221	0.0031	0.020
	222	0.0089	0.020
M 8-1	225	0.0028	0.022
	226	0.0018	0.019
M 9-1	231	0.0048	0.019
	232	0.0055	0.028

This seems to prove that the abnormally high nitrogen content is responsible for the gassiness of these heats. 5. Desulphurization of Steel—As we have mentioned before, this process can be used also outside of degassing, to other operations due to the fact that it can induce any desirable agitation to a bath of molten metal. But these applications are interesting only if they are obtainable in an industrial-size heat. We have not yet results to present, as tests in a 5-ton ladle have just started. However, we have run a number of experimental tests in our 500-lb capacity ladle to see if by agitation of a melt of steel in contact with a slag, desulphurization could be obtained and to what extent.

Figure 11 shows the general procedure followed. The prepared slag is placed in the preheated ladle and the molten metal poured into it. Although the ladle is lined with acid refractory bricks and the slag is highly basic, a noticeable sulphur drop can be obtained in a short period of time. Figure 12 gives some typical results obtained in medium-carbon steels. The time of treatment was limited in all our tests because only 350 lb of metal at the most could be processed in the ladle.

Using a weight of slag which is only 2 per cent of

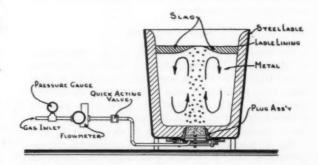


Fig. 11—Schematic layout used for desulphurizing steel by agitation of the metal and slag interface with an inert gas injected through a porous refractory plug.

E.

Ch

in

the

wil

bee

and

bir

Th

tui

eni

the

inc

tio

am

the

lad

W

and

the

flu

W

In

too

SOI

hy

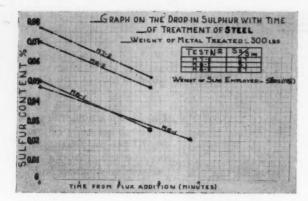


Fig. 12-Results obtained in desulphurizing medium carbon heats by the gasal process.

the weight of the metal, it is of interest to be able to lower 20 to 30 points the sulphur content of a steel in less than 4 minutes. Furthermore, the metal is well degassed by this treatment and the non-metallic inclusions finely dispersed.

Plug Assembly in the Ladle

The actual pressure of metal against the plug is small. For instance, in a 5-ton ladle, the height of steel is about 3 ft. As the cross-sectional area of the plug is only 10 sq in., the total pressure on it is about 105 lb. For a 20-ton ladle, according to the width of it, the total pressure is less than 200 lb. There is no problem to design a plug holder capable of withstanding this pressure with complete safety.

Conclusions

1. Of all common gases available, hydrogen due to its great variation in solubility with temperature is an undesirable element in most industrial metals.

2. Flushing to remove hydrogen dissolved in metals is an interesting process, provided this treatment is applied at the right place and at the right time.

3. A pouring ladle has been developed in which gases can be introduced into molten metals through a porous refractory medium. This ladle constitutes a tool for a number of metallurgical treatment, because it is simple, efficient and inexpensive. This tool, which is in the development stage, offers possiblities and new uses of gases in metallurgy.

Acknowledgments

We want to express our thanks to the Department of Mines and Technical Surveys, formerly the Canadian Bureau of Mines, especially to Mr. C. S. Parsons, who authorized most of this work to be performed in their Laboratories. We wish also to acknowledge the technical aid and kind cooperation of Mr. S. Gertsman, Dr. A. Prince in this research.

Mr. R. Lee who supervised and conducted the experiments, Mr. B. Gladu who made the chemical analyses have to be thanked for their excellent work.

We are finally very grateful to Canadian Liquid Air Co. Ltd. for the permission of publishing this report.

Bibliography

1. A. Sieverts, "The Solubility of Gases in Molten Metals," Stahl und Eisen, vol. 29, p. 1248 (1909).

2. C. A. Zapffe and C. E. Sims, "Hydrogen Embrittlement. Internal Stress and Defects in Steel," *Trans.* AIME, vol. 145, p. 225; *Metals Technology*, vol. 8, Technical Paper 1307 (1941).

3. C. E. Sims and C. A. Zapffe, "The Mechanism of Pinhole Formation," Transactions, American Foundrymen's Association, vol. 49, p. 255-270 (1941).

4. C. A. Zapffe and C. E. Sims, "Hydrogen and Nitrogen as Causes of Gassiness in Ferrous Castings," Transactors, American Foundrymen's Association, vol. 51, p. 517 (1943).

5. Sir Andrew McCance, "Gases and Steel," Foundry Trade Journal, June 8, 1950, p. 619.

 C. A. Zapffe and C. E. Sims, "Hydrogen, Flakes and Shatter Cracks Metals and Alloys, vol. 11, p. 145-171; vol. 12, p. 44-51 (1940).

7. C. H. Lorig, "Cause and Prevention of Intergranular Fracture in Cast Steel," A.I.M.E., *Proceedings Electric Furnace Steel Conference*, p. 234 (1945).

8. S. C. Faddis, "Effect of Nitrogen in Improving Physical Properties of Low Carbon Steel," A.I.M.E., *Open-Hearth Pro*ceedings, vol. 32, p. 260 (1949).

9. Nitrogen in Chromium Alloy Steels, Published by Electro Metallurgical Company, New York (1941).

 L. W. Eastwood, Gas in Light Alloys, John Wiley & Sons. Inc. (1946).

11. T. Swinden and W. W. Stevenson, "Some Experiments on Gases in Iron and Steel and Their Effects on the Solidification of Ingots," Sixth Report on the Heterogeneity of Steel Ingots, Iron and Steel Institute, Special No. 9, pp. 137-150 (1935) and also Special No. 16, pp. 139-142 (1937).

12. C. E. Sims, "Flushing Molten Steel with Neutral Gases." A.I.M.E. Electric Steel Furnace Conference, p. 302 (1949).

13. D. R. Tullis, "Notes on the Treatment of Aluminum and Aluminum Alloys with Chlorine," J. Inst. Metals, vol. 40, pp. 55-61 (1928).

14. N. P. Allen, "Experiments on the Influence of Gases on the Soundness of Copper Ingots," *J. Inst. Metals*, vol. 43, pp. 81-124 (1930).

15. W. A. Baker and F. C. Child, "The Removal of Gases from Molten Bronzes," J. Inst. Metals, August 1944 and December 1944.

 A. J. Phillips, "The Separation of Gases from Molten Metals," Metals Technology, June 1947.

17. M. B. Bever, "Gases in Cast Metals," Iron Age, April 22.

18. A. C. Boak, and G. A. Allen, "Flushing Aluminum with Nitrogen Gas," Canadian Metals—Metallurgical Industries, Sept. 1946.

19. P. M. Hulme, "Fluxing Aluminum with Dry Nitrogen," Materials and Methods, vol. 22, p. 1435, Nov. 1945.

20. P. M. Hulme, "Use of Inert Gas to Flush Dissolved Gas from Molten Steel," A.I.M.E., Proceedings Electric Furnace Steel Committee (1945).

21. P. M. Hulme, "Removing Dissolved Gases from Molten Metals, Steel, Oct. 7 and 14, 1946.

22. E. Kurzinski, "Degassing Non-Ferrous Metals," The Foundry, Dec. 1948 and Jan. 1949.

23. J. M. Crockett, "Some Industrial Uses of Nitrogen and the Rare Gases," Metal Progress. Dec. 1948.

the Rare Gases," Metal Progress, Dec. 1948.
24. Eselgroth, "Nitrogen Degassing of Non-Ferrous Metals."

Metal Progress, June 1949. 25. W. Geller, "Theory of the Degassing of Metal Baths," Zeitschrift Fur Metallkunde, vol. 35, no. 11 (1943).

26. W. H. Baer and B. M. Loring, "Effects of Melting Atmosphere, Time at Temperature and Degasification on Properties of Valve Bronze," Transactions, American Foundrymen's Society, vol. 57, pp. 257-262 (1949).

L. Grand, "Le dégazage des alliages d'aluminium,"
 XXIeme Congres de l'Association Thermique de Fonderie.

DISCUSSION

Chairman: F. KIPER, Ohio Steel Foundry Co., Springfield, Ohio.

Co-Chairman: G. W. Johnson, Vanadium Corp. of America.

Chicago.

C. A. Faist: 1 What hydrogen level is considered detrimental in a steel? From what material is that porous plug made?

MR. SPIRE: The hydrogen level which would be detrimental to steel would be approximately 0.3. If you have nickel you can go up to 0.4 or 0.5. We have been using mostly basic material for the plugs.

WM. HARBRECHT: Did the author find that hydrogen lowered the carbon content? Is this a killed steel the author was working

with?

p

11

35

le

Mr. Spire: The carbon content was not lowered. We have

been using killed steel.

J. R. GOLDSMITH: ⁸ What effect did the various sizes of bubbles have as they were admitted? We have tried similar experiments and I think our failure was that our bubbles were too large. Was the author able to control bubble size to get better scrubbing?

MR. SPIRE: It is desirable to have as fine bubbles as possible. The reason why the bubbles increase in size is that the temperature is increased. The gas comes out of the bottle and when it enters the metal it heats up and the volume increases. Due to the fact that the porous refractory is hot, the gas is heated as it passes through, so there is less chance of having the volume increased due to temperature.

J. A. Rassenfoss: ⁴ I am interested in getting more information on this desulphurizing operation. The author used an amount of flux or slag equivalent to 2 per cent of the weight of the metal. I would like to know what that flux is and whether it was at room or elevated temperature when it went into the

ladler

Mr. Spire: We have been experimenting with various fluxes. We have been using fluxes up to 75 per cent lime with fluorspar and even titanium oxide. We want a high basicity. The higher the basicity and the lower the FeO content, the better. The fluxes were in powder form, premixed and preheated, preferably. We preheated the flux to about 1000 C and placed it in the ladle. In about 1½ min the flux was molten and started reacting. In certain cases the flux was even semi-solid and still active, which is rather surprising, but it is a fact.

The ladle was acid lined and we did not want to have a flux too fluid to avoid introduction of silicon dioxide. We did get

some, however.

MR. RASSENFOSS: I presume you had something to keep the

acid slag from getting into the ladle?

MR. SPIRE: We have not done that yet. In these experiments we had a melt of 500 lb and in the heats we are about to run we will get rid of the acid slag one way or another.

Mr. RASSENFOSS: You have not really made these desulphurizing experiments with this large ladle?

Mr. Spire: No.

MEMBER: The author told of an interesting way to remove hydrogen from steel. He also mentioned that people have been

doing this for a long time by causing oxygen to go through the metal and absorb the hydrogen. Are there any particular kinds of steel where he thinks flushing argon gas would be preferable to using oxygen in the more usual way?

MR. SPIRE: It resolves itself into a question of when you do it. Oxygen is all right in the furnace, but when your metal is in the ladle and ready to pour you could not use oxygen very well. So what is practical about the way we do it is that you can do it while you carry your ladle from the furnace to the pouring platform. That is why it is interesting, and it is something different from the use of oxygen while it is in the furnace.

As far as the type of steel which should be more interesting to degas I would mention the high-alloy type. You do not have to worry much about the hydrogen when your steel is well made. If you just melt down and do not boil the metal, you may use

hydrogen.

On some of the analyses we have seen, the nitrogen gas should never be introduced because of nitride formation. Probably if you could denitrogenate the metal you may be able to add less aluminum because you take away the nitrogen before you take away the aluminum, so that is something else to consider.

J. B. Caine: ⁵ I would like to ask the author a question regarding sulphur removal. Has he checked for sulphur reversion,

in relation to time in his acid-lined ladle?

Mr. Spire: We have been aware of sulphur reversion so we slagged off before pouring because if the silicon dioxide increases the sulphur may revert. The thing to do is to get rid of the slag before pouring your casting, if you can do that, or thicken it up with some slag-forming material which would not be detrimental to sulphur removal.

MR. CAINE: Is there a temperature loss during the gas bub-

bling and the slagging off?

MR. SPIRE: The temperature must be constant all the time.
MR. CAINE: What gas did the author use when desulphurizing?

Mr. Spire: Argon.

J. M. CROCKETT: 8 I would like to emphasize that by developing the porous plug method of ladle agitation, Mr. Spire has provided a positive means of obtaining good agitation and gasmetal contact independent of chemical reactions. This technique should be investigated by foundries having melting or refining problems involving elimination of dissolved gas, promotion of faster slag metal reactions, or floating out nonmetallics.

¹ Research Metallurgist, Burnside Steel Foundry Co., Chicago.

² Research Laboratory, Electro Metallurgical Co., Niagara Falls, N. Y.

³ Metallurgical Engineer, Crane Co., Chicago.

⁴ Research Metallurgist, American Steel Foundries, E. Chicago, Ind.

⁵ Foundry Consultant, Wyoming, Ohio.

Asst. Mgr. Technical Sales Dept., Air Reduction Sales Co., New York.

A NEW METHOD OF EVALUATING COSTS IN JOBBING FOUNDRIES

By
Dr. A. William Schneble, Jr.* and Clyde E. McQuiston**

"A busy shop can go broke as quickly as an empty shop—and more easily, in fact, if costs are not known." This is the statement of Walter L. Seelbach in his editorial, "The Road Ahead" published in December 1950 issue of the American Foundryman. In these times of extended and expanded production, it is essential that an adequate costing method be used in the foundry particularly to disclose which jobs appear to be profitable, but actually are unprofitable. Such a cost system must fulfill the purposes of costing in the foundry, which, in addition to the formation of a basis for pricing products, are:

a. To determine actual costs

b. To control expenditures

To provide a basis for the formulation of operating policy.

In order to determine costs, more accurately, a study was instigated from which a new method of costing in jobbing foundries was devised. By use of this method, the foundryman will be in a better position to determine which jobs are profitable and/or which jobs are desirable considering the cost standpoint.

An analysis was made of the common jobbing foundry costing methods wherein individual jobs are costed by dividing the costs of the Mold and Core Departments for a specific period of time (including their share of the general expenses) by the number of productive molding and coremaking hours, respectively, for the same period of time. This results in the mold and core cost rates to which is added the prorated clean and ship cost rate.† Thus by multiplying these mold and core rates (and clean and ship rate) by the molding and coremaking hours of the individual job, the cost of these operations are obtained. In order to obtain the total cost of the job it is necessary to add to these costs, the cost of the Melt Department. This cost is generally obtained by dividing

the total costs of the department (for a given period) by the number of pounds of castings produced which gives a rate in dollars per pound. This rate is multiplied by the weight of the casting.

There are certain serious inadequacies with the common method of costing. For one, it does not consider the fact that some castings require more floor space during production than others while employing the same amount of productive labor. This places a serious limitation on the productive capacity of the foundry because it is not possible to increase labor in proportion to the increased space required.

Another consideration which is omitted in the present method is that the overall employment of direct producers (molders and coremakers) is limited by the type of castings and, therefore, the productive capacity in hours is likewise limited. Moreover, the commonly used system reflects only the productive hours spent on the job and ignores the capacity limiting characteristics of individual jobs. Jobs vary greatly in the amount of floor space which they require. However, the big area jobs are not necessarily the worst "space offenders." The ratio which best describes the capacity limiting characteristics of a job is that of the floor space required for a job to the number of producers that can efficiently be employed on the job. As a result of employing the commonly used costing method it is impossible to define the total capacity of the foundry. Hence it is impossible to determine the sales, cost and profit relationships except by the employment of average productive labor data. The danger of using average figures is that it tends to hide the jobs which are apparently unprofitable or undesirable (i.e., capacity limiting jobs).

The New Method of Costing

The method of costing proposed herein is based on the hypothesis that each job should pay for its limitations on the maximum production capacity of the foundry in proportion to the amount of this limitation. This method recognizes that each job has three distinctive characteristics. These are:

- a. Evaluated average floor space required by the job,
- b. Ratio of mold direct labor hours to core direct

th

af

dif

^{*} Plant Manager, Advance Foundry Co., Dayton, Ohio.

^{**} Instructor, Industrial Engineering Department, Ohio State University, Columbus, Ohio.

[†] In many foundries the Clean and Ship Department's costs are handled in the same manner as the Mold and Core Department's costs.

labor hours for the job,

c. Average number of molders on the job per day. It is by these characteristics that each job is defined.

The evaluated average floor space required by the job in the Mold Department is determined from two factors: (1) the area of the drag and cope that is required and (2) the total time that the drag and cope are on the floor. This time element is very important since often a job must sit on the floor several days after it has been poured in order to cool. Obviously, when cooling this job is displacing other jobs on which productive work could be done. It is logical then that a job which has a long cooling time costs more to produce than another with a shorter cooling time requiring the same area.

The ratio of the mold direct labor hours to the core direct labor hours is a measure of an amount of productive work required within the Mold and Core Departments. From a theoretical standpoint then, this mold to core direct labor hours ratio would effectively expand or reduce the Mold and Core Departments for each job produced.*

Number of Men on Job

8

a

The average number of men on the job pertains to the Mold Department only and is the average number of molders employed on a job per day. This figure is based upon effective utilization of molders on the job and represents the maximum number of molders which can be used efficiently on a job.

Certain assumptions are necessary before the theory of the new method is given in mathematical form. The first of these assumptions is that only one kind of job is produced and that it is produced at the maximum capacity rate of the foundry. The limitations on the equipment and the general arrangement of a given foundry are considered in the determination of the maximum capacity rate of the foundry. Although it is undoubtedly obvious to the reader, this assumption is necessary in order to reduce the complexity which would exist if all the combinations and permutations of the various jobs produced were considered at any one time.

Another assumption is that the new method of costing is independent of the kind of management operations employed in a given foundry; the new method is "to stand on its own merits." The third assumption is that the method is applicable to all job shop foundries. However, various details of the method will vary from foundry to foundry.

The last assumption made is that the working area per coremaker is a constant. From past experience it has been found that the working area per coremaker for most foundries varies over a small range for various jobs. Thus it is possible to determine empirically this constant. The variance of the working area per coremaker is small enough that it does not seriously affect the outcome of the calculations or the method.**

It is now possible to present the theory of the new method which has been developed from the description of the method and assumptions given above. The evaluated average square feet of floor space taken up by the job, J is defined as:

$$B = A_d t_d, \tag{1}$$

where B = sq ft days of drag,

 $A_d = drag working area,$

t_d = time in days the drag or pit is on the floor (including cooling time);

$$D = A_c t_c, (2)$$

where D = sq ft days of cope,

 $A_e = cope$ area,

t_e = time in days cope is on the floor.

Thence by adding equations (1) and (2) and dividing by t_d , J becomes,

$$J = \frac{B+D}{t_a}.$$
 (3)

By using t_d as the diviser in equation (3), J becomes the effective amount of floor space (sq. ft) that the entire job uniformly occupies for the entire time it is in the shop. This area is entirely imaginary since it includes the area of the cope, when it is occupying floor space, and averages it with the drag working area.

The average number of men on the job per day, N,

$$N = \frac{H_{M_J}}{8t_A} \tag{4}$$

where H_{MJ} = total mold hours of a specific job. The mold and core areas required for a specific job are obtained in the following manner. The maximum number of mold hours, H_M, in the foundry per day is

$$H_{M} = \frac{8MN}{I}$$
 (5)

where M = mold floor area, sq ft.

The maximum number of core hours, Hc, in the foundry per day is,

$$H_C = 8CL$$
 (6)

where C = core floor area, sq ft,

L = number of coremakers per sq ft floor space.

By dividing equation (5) by (6) the ratio of mold to core hours becomes

$$\frac{H_{\rm M}}{H_{\rm C}} = \frac{8MN}{\frac{\rm J}{8CL}} \quad \text{or } \frac{MN}{\rm JCL}. \tag{7}$$

The ratio of mold direct hours to core direct hours on a H_{MJ} specific job, --, must balance such that and thus this ratio can be substituted into equation (7) and by rearranging terms becomes,

$$\begin{split} M/C &= \frac{H_{^{M}J}CL}{H_{^{G}J}N} \\ H_{^{G}J} &= \text{total core hours on the specific job.} \end{split} \tag{8}$$

where H_{c_J} = total core hours on the specific job.

Since M + C = T

where T = total mold and core floor space, sq ft it is possible to determine the maximum number of

^{*} This assumes that the mold and coremaking areas are more or less interchangeable.

^{**} This is illustrated by the fact that a large casting will generally have a number of small core boxes and hence is not much different than a smaller casting as far as the coremaking is concerned.

tl

H

tl

if

th

m

de

bi m va Co

co

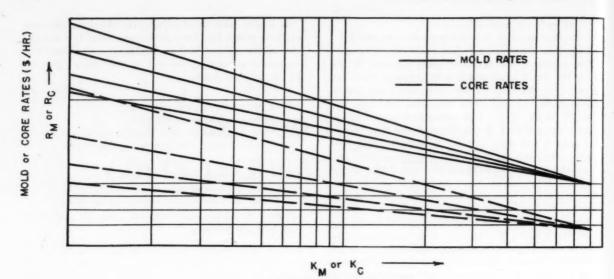
th th sa

ca jo

ca

do

th



MOLDERS or COREMAKERS

Fig. 1-Mold and core cost rates (New Method)

molders, K_M , that can be employed in the shop when molding a specific job:

$$K_{M} = \frac{NM}{I} \tag{10}$$

and in a like manner, the maximum number of core-makers that can be employed on a specific job, $K_{\rm C}$, becomes,

$$K_c = CL.$$
 (11

The constant, L, includes both coremaking and core assembly areas.

The cost rate per hour for the specific job is then determined as:

(a) Mold Rate

$$R_{M} = \frac{O_{M}}{H_{M}} \tag{12}$$

where O_M = the cost of operating the Mold Depart-

Fig. 2-Mold and core cost rates (Common Method)

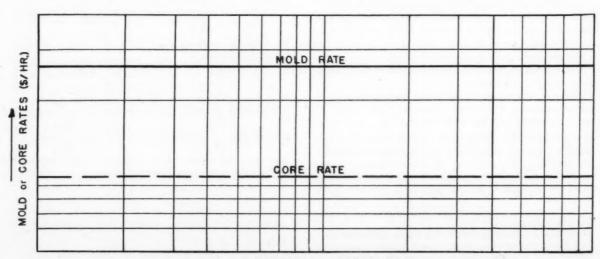
ment per day including the Mold Department's share of overhead items.

$$H_M = 8K_M$$
 (b) Core Rate

$$R_{\rm C} = \frac{O_{\rm C}}{H_{\rm C}} \tag{13}$$

where O_c = the cost of operating the Core Department per day including its share of the overhead items.

Knowing the characteristics of the job, J, H_{MJ}/H_{CJ} , and N it is possible to calculate K_M and K_C , the maximum number of molders and coremakers that can be employed on a specific job by using equations (8) through (11). By combining K_M and K_C with the cost data, the rates R_M and R_C , are obtained in cost per hour of the job. When these cost rates are multiplied by the hours required to perform the molding and coremaking on the specific job and by adding the cost of the alloy (which includes metallic components of the charge), commissions, and patterns, etc. the total



MOLDERS or COREMAKERS -

cost of the job is determined.

In Fig. 1 and 2 a comparison is given of the rate R_M and R_C with the mold and core rates determined in the usual manner. Figure 1 gives family of cost rate curves for varying molders and coremakers.

Cost Data Required for New Method

It is beyond the scope of this paper to evaluate, in detail, accounting procedures of a jobbing foundry. It is, however, pertinent to the new method of costing to discuss briefly certain accounting procedures. The new method is based on the hypothesis that jobs limit the capacity of the foundry because of the space they require and the limitation they place on the employment of direct producers. When this hypothesis holds, the cost data must be accurate if the objective of the new method is to be fulfilled—to determine the profitableness of each job. It is quite clear that the cost data represents an important role in the overall method. However, the new method illustrates a new principle in costing which is independent of the accuracy of the cost data used.

The presently accepted procedure of allocating costs to departments is considered satisfactory in general. However, to use the new method, costs must be extrapolated to various levels of operations. In many cases the levels of operation for which the cost data must be secured will be beyond the practical level of operation of a given foundry. These cost data will be purely hypothetical in nature and it will challenge the foundrymen's imagination in considering it as an actual practice. However, for the specific job in question these data represents what would actually take place if the foundry were operating at capacity making only

this one job.

The cost breakdown for the new method will be by the Mold, Core, Clean and Ship, and Melt Departments on items which are directly chargeable to these departments. The overhead and general expense items will be compiled on the basis of combined departmental costs (excluding the Melt Department) varying by men in the Mold and Core Departments where expenses vary according to men employed and combined departmental expenses (excluding Melt Department) varying by floor space where expenses incurred vary according to floor space used in the Mold and Core Departments. The Clean and Ship Department costs will be added to the combined departmental costs varying by men. This enables the management of the foundry to maintain some control over the costs of the Clean and Ship Department while still prorating the costs to the Mold and Core Departments in the same manner as currently used in many foundries. In case a foundry prefers to obtain cleaning costs on each job individually, it would require only slight modification of the new method.

The costs, in addition to the Departmental breakdown, are also separated as rates, fixed and semi-fixed costs. The semi-fixed costs are determined at various levels expressed in terms of men. The interval for these costs is chosen such that they reflect all changes in costs of the items concerned.

Actual costs by departments for use in the new

method are determined in the following manner. The direct Mold Department costs for a period are determined by adding all rates (which are based on \$/molder-hour). This total rate is multiplied by the number of mold-hours for the period. To this is added the semi-fixed costs for the same period giving the total direct Mold Department costs. The Mold Department's share of the Clean and Ship Department, overhead and general costs is determined by adding all fixed and semi-fixed combined departmental costs varying by men.

A portion of these costs are prorated to the Mold Department by multiplying these costs by the ratio of

 $\frac{K_M}{K_M + K_C}$. This ratio is used because it prorates the

overhead and general items varying by men on the basis of the number of men employed on an indi-

vidual job.

The Mold Department's share of the combined departmental costs varying by floor space is determined by adding all items in this category for the given period and multiplying by the ratio, M/T. This ratio

is used for a similar reason that the $\frac{K_M}{K_M + K_C}$ ratio is

used; namely, it prorates these costs to the Mold Department on the basis of floor space used by the individual job.

The Core Department costs are obtained in a like manner. The Melt Department costs are determined in the customary manner as described before.

Comparison of Results of the Two Methods

In order to obtain a comparison of the two methods, a limited number of jobs were costed by both methods. The results are given in Table 1. It can be observed from Table 1 that due to the capacity limiting efforts of each job, costs determined by the new method vary from 75 to 120 per cent of the costs considered accurate as determined by the commonly used method. Furthermore, Table 1 indicates that seven jobs out of 37 have a cost ratio (new method to commonly used method) greater than one. As a consequence it can be seen that a foundry costing their castings by the present method would determine costs and set prices, on the majority of their castings, substantially higher than required to make a fair profit.

There are two contributing factors to this condition. First, those jobs which have a cost ratio greater than 1.0 are the jobs with a high J/N ratio. These jobs, therefore, limit the employment of direct producers to a few men. Any job requiring a large amount of floor space for each molder employed on the job will obviously be one that is bulky in size and occupies a fairly large amount of total floor space. This type of casting will require a foundry with suitable large flasks, pits, overhead cranes, pattern layout men etc., in order to be able to cast it.

In contrast, a foundry which does not have these facilities will be forced to produce predominately the low J/N type of castings. On this type of work the commonly used costing method will give a lower cost

TABLE 1-EXPERIMENTAL RESULTS

Cus-		Weight,	K _M	Kc		Ratio of Cost of Job New/Old	Space	J/N
A	1	780	35.3	35.3	70.6	0.83	90	160
В	2	37900	9.7	8.5	18.2	1.20	10400	1200
C	3	25500	35.0	16.6	51.6	0.92	2100	269
D	4	22430	13.7	17.6	31.1	1.05	8000	700
E	5	4900	42.2	25.9	68.1	0.85	480	183
F	6	1080	38.4	27.4	65.8	0.88	84	193
F	7	1280	43.9	27.4	71.3	0.88	84	168
G	8	1312	34.9	59.4	94.1	0.83	154	385
G	9	1063	26.1	28.9	55.0	0.89	162	270
E	10	1000	46.7	34.4	81.1	0.79	147	124
E	11	400	41.4	22.3	73.7	0.82	141	205
E	12	980	41.1	30.3	71.4	0.82	164	162
E	13	300	66.6	6.7	73.3	0.76	90	181
E	14	300	52.5	3.2	55.7	0.80	153	242
E	15	350	57.9	49.3	107.2	0.72	126	216
H	16	2790	36.4	34.2	70.6	0.84	262	161
H	17	3780	31.4	12.2	43.6	0.94	388	346
H	18	745	22.1	40.7	62.8	0.88	80	198
H	19	3520	28.0	17.4	45.4	0.94	387	346
C	20	15650	53.7	16.4	70.1	0.89	794	185
C	21	10040	37.7	31.6	69.3	0.89	510	170
C	22	16540	47.8	16.9	64.7	0.92	652	205
1	23	2500	51.5	10.3	61.8	0.92	109	217
1	24	2500	13.3	32.9	46.2	0.97	202	460
K	25	1100	32.8	27.3	60.1	0.89	150	217
L	26	1200	78.7	0	78.7	0.81	105	172
	27	19000	10.6	29.1	39.7	1.003	657	657
	28	1350	57.6	0	57.6	0.89	103	237
	29	2600	76.0	0	76.0	0.88	110	177
0	30	6510	58.3	9.7	68.0	0.87	438	195
0	31	660	38.8	0	38.8	0.92	173	348
E	32	7600	50.0	26.2	76.2	0.80	1240	153
	33	8200	40.2	28.7	68.9	0.85	870	176
	34	17500	14.5	10.4	24.9	1.10	5630	772
-	35	15460	12.1	20.0	32.1	1.05	4170	750
-	36	21000	11.1	19.9	31.0	1.06	4600	817
	37	12000	7.3	20.6	27.9	1.12	3140	1225

rate since the large majority of castings produced will be of the higher molder and coremaker capacity type.

As a result the first foundry equipped to make all types of casting will of necessity quote a higher price on the low J/N jobs than the second foundry. This will tend to drive the low J/N class of work from the first foundry and make them specialist in the high J/N castings only.

It will also cause the first foundry to have an apparent increase in costs per labor hour since the low J/N jobs are no longer available to absorb part of the expenses of making the high J/N jobs.

As a second factor, it will be noted that in general the high J/N castings require a large amount of mold floor space days. Table 1 shows that as many as 100 of job No. 26 could be molded in the space as one of job No. 2. This will explain why only seven of the 37 jobs listed had a higher cost than determined by the commonly used method. This would tend to indicate that a reduction in overall total costs has resulted by use of the new method. This, however, is not the case since all costs were derived for the same source, and, thus, the total of all costs of all 37 jobs is the same whether costed by commonly used or new method. Actually the new method gives a more accurate means of distributing costs to the various jobs.

Most jobs which are available to the jobbing foundry are similar to the jobs listed in Table 1 with a cost ratio of less than one. Any foundry limiting their operations to only these jobs of the low capacity type would be faced with erratic operations during certain periods and would be limited to a small number of customers. It would be impossible for such a foundry to charge sufficiently high prices when this work was available to maintain profitable operations when work was not available.

In summary, a competitive jobbing foundry pricing their work selectively by the new method would find themselves in a healthy competitive position at all times. Under normal conditions the limited amount of competition on the low capacity type of castings will allow the increase price required and still not limit their orders on this type of work appreciably. As a result the foundry's operations will be stabilized by a large variety of types of castings and a large number of customers. A large variety of products and customers are paramount to the success of a jobbing type industry. The same foundry when operating under high business activity conditions will be in the position to selectively choose those jobs, which in addition to providing a good profit, can be chosen from the ease of manufacture, delivery date and other such factors.

A captive foundry which cannot normally select its jobs, can by use of the new method more accurately determine costs of jobs which ultimately will have an effect on the price of the end product manufactured.* When a captive foundry is in the position where it cannot produce all the castings required, the new method will indicate those jobs which can be most profitably "farmed out" either from the cost or ease of manufacture standpoint.

Bibliography

- 1. C. E. McQuiston, "Analysis of and a Proposed Solution to the Costing Problems of a Job Shop Foundry," Ohio State University, Columbus, Ohio, 1950.
- 2. L. P. Alford, Principles of Industrial Management, New York, 1940.
- 3. Cost Manual No. 1, A Simple Cost System for Gray Iron Foundries, Gray Iron Founders' Society, Inc., Cleveland, June 1941
- P. J. Esling, Basis of Cost Distribution in Malleable and Gray Iron Foundries, National Association of Cost Accountants' Bulletin, March 15, 1948, pp. 853-8.
- 5. H. L. Dyer, Costing for Pricing and Control in Gray Iron Foundry, National Association of Cost Accountants' Bulletin, March 15, 1948, pp. 847-52.
- R. L. Lee, "Controlling Foundry Costs," American Foundry RYMAN, July 1948, pp. 34-35.
- 7. O. L. Ducharme, "Know Your Cost Factors," American Foundryman, June 1948, pp. 57-9.

DISCUSSION

- Chairman: R. L. Lee, Grede Foundries, Inc., Milwaukee.
 Co-Chairman: G. Tisdale, Zenith Foundry Co., Milwaukee.
 Robert Gregg: This paper, according to the authors, discusses a new method of foundry cost accounting.
- If jobbing foundries would clean up their plants and not use their molding space for flask storage, there would not be much justification for this new method.

This analysis is particularly helpful in costing repair parts which will not be the same as the average cost of all the castings required in the end product.

Most jobbing foundries base their cost finding on direct time of molding, core making, and cleaning, and leave out a very important cost item, namely, flask equipment and special rigging Flask equipment and rigging should be charged to the job and not charged into overhead or Capital Investment; and what

should be foundry profit, often lays wasting in the flask yard. Jobbing machine shops charge for tooling, jigs, and fixtures

to do the job; so why should not the foundry charge directly for flask equipment and other rigging?

IFS

id-

010

eir

pe

in

of

IV

as

rk

ng

 $^{\rm 1d}$

ıll

nt

28

ot

d

11-

e

21

n

e

ts

n

it

ŧ

MR. McQuiston: It would be simple with this new method to charge the customer for rigging and flask equipment, since you are costing each job individually. It is merely an accounting

procedure to get those costs worked into the system.

CHAIRMAN LEE: We do charge for special flask equipment and have been doing so for years. I agree with Mr. Gregg that a foundryman would go bankrupt if he would spend a lot of money in rigging his customers' patterns to special equipment.

H. R. WILLIAMS: 2 I think the principles herein discussed should be considered in cost accounting. I agree that space requirements in a shop have a definite affect on the profits a

foundry can earn on a job.

This seems to be a rather complicated way of getting at this cost factor, but perhaps that is the way it has to be done if we

are seeking accurate cost accounting.

Mr. McQuiston: The formulae presented in the paper are not beyond comprehension of most foundrymen. Actually they are simple and calculations can be worked out which can be done rapidly with clerical help. All equations are first order equations and should be simple to handle.

CHAIRMAN LEE: I agree with you. The equations, however, will require some study on the part of the foundryman.

W. J. MACNEILL: 8 Were these calculations carried out to the point where K equaled one molder and M was only a fraction of the total area of one molding floor? In other words, has this been projected to include molding of small castings which are shaken out soon after pouring, with the result that a given unit of floor space may accommodate two or more molds during a given work period?

Mr. Schneble: The calculations are such as to apply to a flask which would occupy the molding floor an hour or so. This method would find limitations in a production foundry that

had fixed equipment.

A foundryman operating a production shop should study this method carefully and possibly find some application of it in his own foundry with variations, of course.

MR. McQUISTON: I think the conception will still hold, but the applications are different.

C. E. Westover: 4 This is a good thought provoking paper based on a certain amount of theory which should raise questions of practical application. It seems impractical to consider filling a commercial castings shop with any one job on the theory that making that job stand on its own, would produce a profit for the plant, make full use of its facilities and enable one to be competitive. We must recognize jobbing is a setup that is not practical for 100 per cent operation on any one job and it can be profitable only if a good per cent of its facilities are used.

Mr. McQuiston: Actually, it is only a relative matter and if you try to handle all the combinations and permutations of the jobs, it would be hopeless mathematically. It would not be worthwhile, by the reason that we make the assumption that only one job fills the shop, which never actually happens-it did not happen when we checked the data-is to simplify the calculations. We want to get a relationship between jobs.

MEMBER: Could not something simpler be devised taking into consideration tonnage of metal poured per square foot of floor area? This would involve size of flask and weight of metal

¹ Reliance Regular Div., American Meter Co., Inc., Alhambra, Calif.

MR. McQuiston: Of course, once you know the size, you are talking about floor space again.

MEMBER: I think it is simpler than the complicated method

presented in this paper.

Mr. McQuiston: The approach of using man-hours was to make a transformation of what is normally done day by dayyou cost on the basis of direct producers in most cases, and so we carry through that some process.

I think you would encounter difficulty if you were to use the weight basis. You get into complicated jobs requiring more

than one molder, complicated shapes, etc.

MR. SCHNEBLE: The difficulty you would encounter using a weight basis would be similar to that if you used weight as a basis for regular costs. It would be identical whether you use hours or a weight basis for the common denominator. Whether you use this method or the orthodox method of just considering each job, that would all be the same.

There is little relationship between the weight of a casting and the cost required to produce it other than the metal itself that you pour into it. I think most foundries have discontinued

costing their jobs on a weight basis.

MEMBER: What I had in mind was the side floors. You could make definite assumptions as to what would be profitable in oounds per square foot of side floor area which has to be handled mainly in pouring.

MR. WESTOVER: Floor space is one of the cheapest items in the foundry. It is a factor to be considered, however. Our main item of cost is labor, usually about 50 per cent of total costs. It seems to me too much emphasis is being placed on floor space as a factor of cost in determining selling price.

MR. McQuiston: We are still using all the labor data that we used heretofore. We are simply using the floor space as a basis of proportioning where floor space is a factor, but we are

not disregarding the labor cost data.

CHAIRMAN LEE: Some of you foundrymen who have been in the foundry industry for a long while can recall some of our old cost manuals and ways of figuring costs. A Cleveland man, a Mr. Runge, at least 20 years ago started this floor space idea and it has been used in many foundries. The purpose of presenting this paper was to provoke some thought on the part of foundry-

MR. SCHNEBLE: Floor space is not an item to be charged for as such under this system. The entire idea of the floor space is to express the capacity in the foundry. To think of charging for floor space so many dollars per square foot is erroneous. In order to determine a selling price, a foundry must charge enough to make a profit at somewhat less than 100 per cent operation. The question immediately comes up as to what is

100 per cent operation.

That difficulty is what provoked the thought that started the work on this paper and this provides a means of determining just what is 100 per cent capacity. You can set your break-even point at 100 per cent so that you can earn no profit. When you do not make a profit operating full force, or 50 per cent, or perhaps 80 per cent, all this does is give you a guide post and it is not the same for all castings. It is the same for those castings that use the same amount of space per molder. The physical size of all plants is limited by the four walls around them and if you put twice as many men in the same spot, they are going to share the same overhead items that are fixed, the insurance, the rent, etc., and the cost of operations on an hour basis will be reduced by increasing the number of productive people inside the same four walls of the building. That is the thought which prompted work on this paper. Do not get the idea that floor space is being charged for as such. It is not. It is only to determine what is 100 per cent of capacity.

MEMBER: What adjustment would you make on that capacity during boom times when you cannot get the molders to fill your floor space and in slack times when you cannot get the work to

fill your floor space?

MR. SCHNEBLE: The system is independent of management of the foundry. Whether you have enough orders to fill the foundry or not has nothing to do with this method of costing.

² Williams Management Engineering Co., Milwaukee.

⁸ Acme Aluminum Alloys, Dayton, Ohio.

⁴ Westover Engineers, Milwaukee.

DESIGN OF CORE BOXES AND DRIERS FOR USE ON CORE BLOWING MACHINES

By E. A. Blake*

THE CORE BLOWING MACHINE was first introduced to the foundry core room about 40 years ago. At that time, and for many years thereafter, the foundryman who decided to use the blowing principle of core production had to work out his own technique for core box rigging and sand preparation since the existing sand mixtures and core boxes were in most cases unsuitable.

In the early days of development of the core blower, the various ingredients of the core sand mixture were not highly developed. The usual procedure was to mix a linseed-base core oil with an unbonded sand, a little kerosene, a dry binder (mainly dextrine) and some bank sand. Whatever may have been the combination that finally worked, it was considered confidential information.

The prepared sand mixture was shoveled into the hopper of the core machine. The machine was designed so that by proper manipulation of the valves, the sand in the reservoir was blown by compressed air through suitable holes into a core box.

Certain preparations had to be made in the core box such as locating and drilling of the blow holes and providing means to allow escape of the air trapped in the box. Most of the venting done in those days was called joint venting. This was accomplished by scoring the box joint with a fine file to provide openings for the air to escape, but small enough so the sand would not go through.

Progress was made slowly. The early successful users of core blowers found this method of making cores economical and efficient. They carefully guarded the technique they had learned the hard way. Those who were unsuccessful considered the core blower a failure.

With the constant improvements made in machines and materials, ideas and methods were exchanged more freely among foundrymen. The art of blowing cores grew slowly but surely in the period from about 1920 to 1929.

The progress that was made prompted manufacturers to explore the possibilities of core blowing as it

was apparent that the field had hardly been touched. In the meantime, there was a growing interest in machines that would produce larger cores than those previously blown.

In 1933, several makes of core blowers were available in different sizes. During the last 15 years, the greatest development in the use of core blowers has taken place. Today machines are available to blow cores weighing from a few ounces up to several hundred pounds. These machines are for manual, semi-automatic, or completely automatic operation, and include mechanical handling of the core box as well.

Automatic core blowers are in operation in several foundries on high production jobs. The trend is for equipment that is efficient and economical—machines that will relieve the operator of heavy work or operations requiring skill.

To accomplish this, core boxes and core driers must be designed and constructed properly. The purpose of this paper is to point out the more important design requirements for equipment used with core blowing machines.

During our early experiments in blowing cores weighing from a few ounces to approximately 20 lb, we found that certain fundamentals had to be followed, such as:

1. A sufficient supply of dry air through a line of ample capacity at a pressure, in general, not less than 80 psi was necessary.

2. Properly prepared sand with controlled moisture was essential.

Machine clamps, both vertical and horizontal, must be of sufficient capacity to clamp the box tightly during the blow cycle.

 The machine should be equipped with a sand reservoir of suitable size for the core being blown. Large reservoirs for blowing small cores were not practical.

5. The blow plate should be properly drilled with holes of proper size and to accurately match those in the core box.

6. The core box should be machined parallel top and bottom and if clamped from the side, both sides should be parallel.

7. The joint should be accurately fitted to prevent

^{*} The Osborn Manufacturing Co., Cleveland.

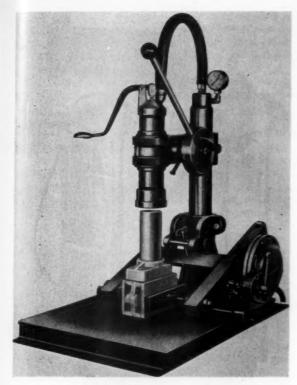


Fig. 1-Bench-type core blower

leakage of sand.

st

n

d

8. A guide rail and stop should be provided so the holes in the box will line up with holes in the blow plate each time the box is placed in the machine.

9. Blow holes should be drilled and spaced so the sand will enter the box and pack the sand uniformly from the bottom up.

10. Air release vents have to be inserted at various locations to divert the sand stream to pockets and corners or around bosses. We found that vents in the lower part of the box were necessary, but they lose their efficiency when covered with sand; so, therefore, more vents are required in the upper half of the core box.

We found that by following the above and analyzing a core before deciding how to blow it, and using good judgment as to where to place the blow holes and air release vent in the core box, we could produce satisfactory cores. With little extra work on the core box or core sand mixture, cores were made that were firm and would draw clean from the box. They would hold their shape without sagging when going through the baking cycle.

Accompanying illustrations show the different types of core blowers commonly used, and illustrate the conditions that take place when a core is blown.

Bench-Type Core Blower

The bench-type core blower (Fig. 1) first appeared about 1943. Due to its simplicity of design and ease of core box rigging, it was promptly accepted by foundrymen. It is recommended for blowing small

cores ranging from a few ounces to about 2 lb with a surface area of approximately 24 sq in.

The operation consists of filling the cartridge with sand by hand, placing the core box and cartridge in the machine, holding the cartridge tight to the box by applying pressure on the clamping lever, and pressing down on the blow control lever. The clamping lever must be held down slightly longer than the blow lever to allow exhaust air to escape, as this machine does not have an exhaust valve.

Diaphragm-Type Core Blower

The diaphragm-type machine (Fig. 2) is made in several sizes. The smaller machines blow cores from a few ounces to 15 lb. The larger machines blow 50 to 200-lb cores. The surface area of cores blown on the smaller machine is a maximum of approximately 140 sq in. while on the large machine it is a maximum of approximately 600 sq in.

A machine designed to blow cores larger than a few pounds and with larger surface areas becomes more

complicated in design.

A method of moving blow head or sand chamber to the blow position over the core box should be mechanical. On this machine, it is accomplished by an air cylinder with spring return.

A sufficiently large air inlet to the machine should be provided so that when more and larger blow holes are used, the pressure drop from the blow valve to the core box is held to a minimum. A 21/2-in, diam air

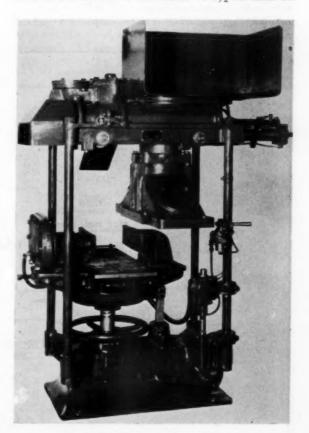


Fig. 2-Diaphragm-type core blower

E.

73

lai

0

0

sic

in

th

ha

ing

fa

be

or

ba

pl

of

th

fr

bl

th

li

inlet pipe is recommended on the smaller size blower and up to a 4-in. diam pipe on the larger size.

Air trapped in the sand chamber at the end of the blow operation has to be exhausted before unclamping the box; therefore, an exhaust valve must be incorporated in the design to do this quickly.

Power clamps are required and must be of sufficient size to keep the box tightly closed during the blow cycle.

These operations are controlled by one or two manually-operated valves or by an automatic sequence valve that will move the blow head from fill to blow position, clamp the core box, blow the core, exhaust the excess air, release the clamps, and return the blow head to the filling position.

Draw-Type Core Blower

The draw type core blower (Fig. 3) is substantially the same as the diaphragm machine except that a cylinder and piston are used to perform the clamping operation. The upper half of the core box is usually mounted on the blow plate on this machine, and the core is drawn from the upper half of the core box on the down stroke of the clamp piston.

Side clamps are not standard on this type machine, and are only used in special applications.

This machine has an automatic control valve which will put the machine through its complete cycle by moving the starter handle downward. After the cycle is completed, the starter handle returns to the up position ready for the next cycle.

Clamping pressures on both the diaphragm-type and

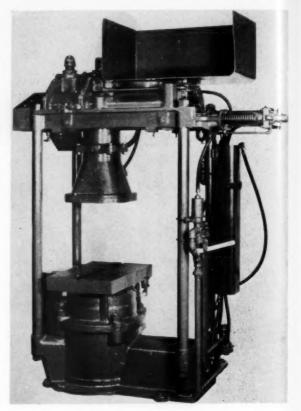
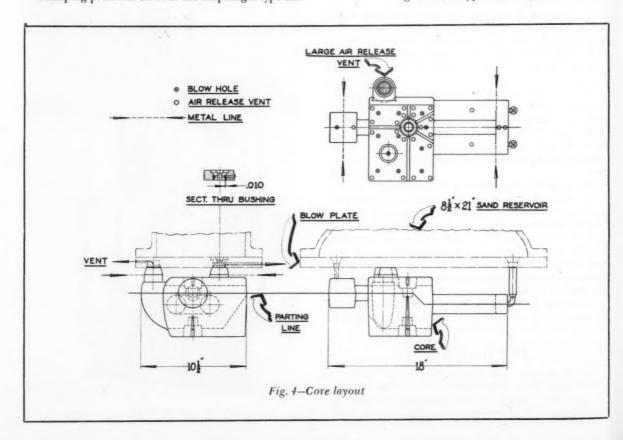


Fig. 3-Draw-type core blower



the draw-type machines will range from a minimum of 7500 lb on the smaller machine up to 55,000 lb on the larger machine. Such pressures applied on the core box must be considered when designing and building core box equipment.

Figure 4 shows a complete core illustrating various conditions typical of many cores that might be con-

sidered for core blowing.

This core has a large center section with bosses having ribs forming pockets in the main body. There are two round extensions to the right and a projection to the left of two diameters. On the upper surface, we

have two small conical core prints.

In studying this core, we find that a practical parting would be as outlined through the center of the core. To blow this core, there are two practical surfaces from which to work, either the lower face or the upper face. The upper face was selected to blow into because it is the shallower and the more practical side on which to place the drier on which this core is to be baked.

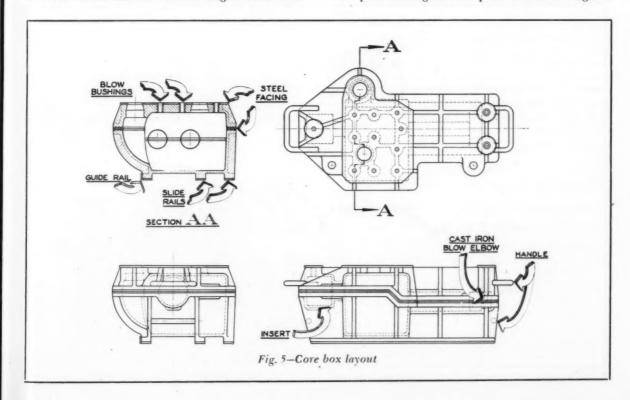
Examination of this core indicates that the best places to locate blowing holes are where the sand will readily flow into the various pockets in the main body of the core so that when the sand enters the core box through these blow holes, the core box will be filled from the lowest level upward until completely filled. A blowing hole was located in the round projection to the left outside the metal line indicated by the dotted line in order to assure proper packing at this point. The projections or baxrels to the right could be blown by vertical blow holes in the print end, but due to the extreme length of this blow hole, it was decided that it would be preferable to blow as shown. By directing a stream of sand into the barrel through a blow elbow

as indicated, the core box wear at this point will be greatly lessened,

The projection at the left with the curved surface at the bottom lends itself nicely to blowing, but the abrasive action of the sand would be quite destructive. It would be best to omit a blow hole at this point due to the fact that blowing directly on this curved surface creates a point that wears rapidly and is extremely hard to repair. Instead, a large vent was placed over this opening and as the core box is filled, the sand will direct itself upward toward this vent, packing the print and the upper pocket solidly. The other conical print at the top of the core is located over the main body of the core, and the sand directed through this blow hole strikes a flat surface at the bottom and far enough away from the blow hole that the wear at this point should not be excessive.

As previously stated, most of the air release vents are in the top surface of the core box. The only ones that the writer has indicated in this particular core in the bottom of the core box are those around the boss in the center main body of the core. Others would be added only if required. It should be kept in mind that there should be no interference between blow holes and vents. In other words, all vents should lead to atmosphere and be so arranged with relation to the blowing holes that air and sand entering the box will not interfere with escape of trapped air.

After further study, we found that it would require an 8½-in. x 21-in. inside opening of the sand reservoir to properly blow this core. The writer indicated this in Fig. 4 and has also shown the two types of vents that are placed in the blow plate rather than in the core box. To the left is the large vent over the one core print leading to atmosphere and to the right is



E.

tak

for

spa

be

me

to

ha

the

ma

cer

an

wi

WI

slie

ab

tw

the

the

dr

the blow hole through a blow bushing that has a vent area around the outside that allows the air to escape to atmosphere. This is shown in the enlarged view in Fig. 4.

Before building a core box, information should be at hand as to whether the requirements are for a few or many cores from a given box. If this core were a short run job that is run occasionally, there is a possibility that it could be blown in a well constructed wood core box. However, we are assuming that the production requirements are sufficiently high to justify careful consideration of those factors which will result in maximum life, efficient operation, and low maintenance. Our discussion of design, therefore, is based on a core box made of either magnesium or aluminum.

In the design of this core box, we considered its weight, strength, and rigidity, and have made provisions to minimize wear as well as simplify maintenance. You will note (Fig. 5) that the top and joint are steel faced. The top and bottom surfaces should be parallel to permit clamping without the possibility of sand leakage during the blowing operation. The bottom slide rails and the guide rail to one side are steel faced, and the center rails are spread out to give good support to the core box joint so that under the clamping pressure, this joint is tightly held. The joint flange should be well supported with ribs.

The top of the box is steel faced with two types of steel facing to minimize wear and reduce maintenance. At the center of the core box, it covers the entire surface of the main body due to the fact that there are a number of blow holes in it and it is more economical to use this type of steel face rather than a series of washers such as are shown at the ends. Bosses are

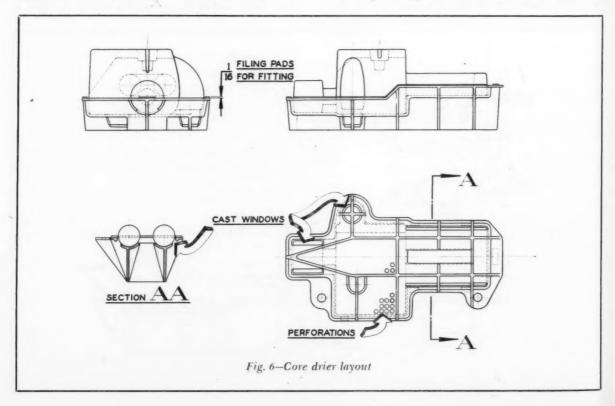
provided at those points where blowing holes are to be located. In a production core box, these holes should be bushed, the usual practice being to make the bushings of thin wall seamless tubing. The tubing can be peened over the steel facing, making a tight seal at this point. The plate should be securely fastened to the top of the core box with flat-head screws. The round washers at each end should also be secured with small screws and the blow bushings should be peened around the opening at the blow hole.

At the right hand end, a cast-iron elbow insert is provided to take the wear from the blow stream, and these parts are held in place by two screws and provision should be made so that they can be replaced when they become worn.

At the opposite end, a different condition is shown. Here we have placed an insert directly below the blow hole and this insert should be made so that it can be replaced. The inserts are usually made of cast iron or synthetic substances now available that have great resistance to abrasion.

Handles should be provided at each end on both halves of the core box.

You will note the core box pin lugs are spaced well apart and are in line and on one side of the box. This is good design in that it permits accuracy and convenience in locating the top half of the box on the lower half and also simplifies the operation of placing the drier over the core box pins in proper position to be dropped down over the top of the core. In general, you will note that the flanges are well reinforced and the bottom ribbing does not necessarily have to conform to the top ribbing of the box since in the layout of the top ribbing, the location of the blow holes was



INES

e to

oles

ake

oing

ight

ast-

ews.

red

be

t is

and

oro-

ced

wn.

OW

be

or

eat ,

oth

vell

his

eni-

wer

the

be

ral.

nd

on-

nic

vas

taken into consideration. In laying out the ribbing for the lower half of the box, the desire for widely spaced guide rails determined their location.

In the design of a core drier (Fig. 6), the aim should be rigidity and lightness. Wherever possible, excess metal should be removed. While it is common practice to use annealed aluminum castings for driers, we may have to revert, under present conditions, to cast-iron driers.

This drier is laid out showing the main flange of the drier relieved from the core box joint approximately ½6 in. with suitable filing pads for fitting at certain locations. With the ribs running lengthwise and crosswise giving rigidity to the drier itself. You will note on this drier, the ribs or rails that run lengthwise are spaced out widely as these ribs are used to slide the core and drier on to a rack in the core oven. To get lightness as well as proper heat circulation about the core, we have indicated perforations in the main body of the drier and windows going around the two core prints at the center section and windows on the two barrels and the two diameter projection at the opposite end of the core. In other words, the drier is designed so that only the surfaces of the core

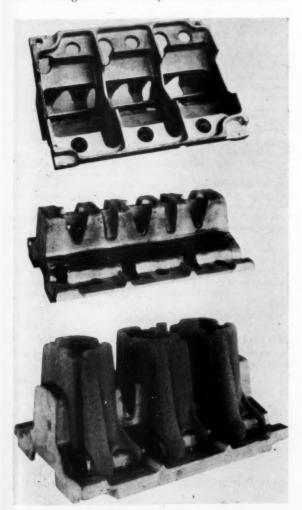


Fig. 7-Typical example of a core drier

that have to be supported are supported using as little metal as possible to maintain rigidity.

To elaborate on what is meant by cutting windows in the core drier, refer to Section AA, Fig. 6. On many driers, the drier is fitted to a point approximately 45 degrees each side of the center line, and then a relief is made running from nothing to 45 degrees to about $\frac{1}{16}$ in. at the parting. The metal following this contour was left in the drier. If we have this relief for clearance, then why is it necessary to have metal there at all? This metal can be taken out when the drier is designed and as a drier has to be handled many times during the day, lightness is an important factor.

By the same reasoning, we only supported the core prints in the upper view to the right by placing a fitted pad on the bottom only with the sides relieved entirely. The same condition exists on the projection at the left end of the core. We have supported only the bottom of the large diameter and with this diameter properly supported, the small diameter of this projection would not have to be supported at all. In general, both the design and construction of core driers is as important as the design and construction of the core box itself.

What we have done here is to take a typical core as an example, and analyzed it to determine the method of making it on a core blowing machine. We have designed a light-weight core box with strength and rigidity to stand high production use. We have incorporated in its design steel facing, steel blow tubes, and replaceable inserts to simplify maintenance.

We have tried to show by proper design that much maintenance can be eliminated and when wear does occur, the maintenance can be taken care of and the box put in proper shape with the least amount of expense.

We have then designed a drier for this core, pointing out the importance of lightness and rigidity and have shown where metal can be eliminated to reduce the weight of such driers by using perforations and windows to eliminate metal where it is not necessary.

There has been, and perhaps always will be, a never ending change of shapes and designs in castings, presenting new problems for the foundry. We hope that some of the features we have pointed out will build better core boxes for use with the core blowing machine.

DISCUSSION

Chairman: HARRY LEES, Whitin Machine Co., Whitinville, Mass.

Co-Chairman: V. C. Rejd, City Pattern Foundry & Machine Co., Detroit.

F. C. CECH: 1 How would you determine whether to use a screen or a slot vent in the core box?

MR. BLAKE: I would use the slot vent only under conditions where a scar such as caused by a screen vent would not be permitted on the core, or possibly under conditions where you would want to shape the vent to the contour of the box to get exceptionally smooth surface. The slotted vent has less than one third of the area of the screen vent, but it has definite advantages where a scar is not permitted. Wherever possible, I would use a screen vent.

W. L. Larson: What size blow holes would you recommend for blowing cores of from 5 to 40 lb, and should the holes be at a 90-degree angle to the blow plate?

MR. BLAKE: It is difficult to recommend the size of blow holes on a pound basis. Each core should be analyzed for the location and the size of blow holes, and that is what has to be done before rigging core boxes. The bond of the sand as well as the shape of the core determines the size of the blow holes.

Member: How do you prevent the sand from the blow hole becoming part of the core? Are there any provisions made for that?

Mr. Blake: There are several methods for doing that. The common method is to take a pipe tap and run a few threads in the blow hole. Then there are steel blow bushings that have a groove in them, and there are also plastic bushings. There are several methods to prevent the slug in the blow hole from becoming a part of the core.

MEMBER: What is the accepted method? Does there seem to be any?

¹Patternmaking Instructor, Cleveland Trade School, Cleveland. ²Research & Development Engr., The Dow Chemical Co., Bay City, Mich. MR. BLAKE: We are in a business where there are no accepted methods. Everybody has his own ideas on that. Particularly, I do not like the plastic blow tube. I have seen too many conditions where in blowing, the tube expands under pressure and contracts when the pressure is released. In a jacket of this type, the slug of sand from the blow hole absolutely will not disintegrate, and the slug of sand will show when the core is baked and will cause a scrap core.

Member: What method do you use? We have found this

MEMBER: What method do you use? We have found this tubing you spoke of the cheapest and best application we have been able to cover. The steel tubing is inexpensive and does not cost much to put it in.

MR. BLAKE: I believe it is the most economical. I am familiar with steel tubing with approximately $\frac{1}{32}$ in. thick wall and when this wears, it can be knocked out and replaced with little expense.

me sol me col he con of seg clu res con res rec fre en ine ler an de ne tu

un

ha se es be

M

MECHANICAL PROPERTIES OF CAST STEEL AS INFLUENCED BY MASS AND SEGREGATION*

By

John F. Wallace**, John H. Savage,† and Howard F. Taylor††

ABSTRACT

Four heats of various low-alloy steels were cast in green sand molds in the form of 5-in. diam well-fed test coupons, some solid and some with a 1-in. round, hollow cylindrical sand core in the center. Tensile and V-notch Charpy impact specimen blanks were cut from the outside and center of these coupons and water quenched and drawn in small sizes. After heat treatment the pieces were machined and tested, including complete transition curves for the impact specimens. Specimens from these locations were subjected to microscopic examination of the structure and path of fracture as well as chemical segregation analyses.

The results of the mechanical tests permit the following conclusions: (a), Increasing mass reduces the tensile and yield strength somewhat, tensile ductility considerably, and impact resistance only slightly in the solid coupons. (b) In the sand cored coupons, the effect of mass on the strength and impact resistance is eliminated and the effect of mass on ductility

T-00 er

HW. iis ve

es

nd th

> Microscopic examination of the metal and fracture surfaces from the various locations illustrates the reason for these differences. Segregated interdendritic areas and agglomerations of inclusions (frequently in a chain-like arrangement) are prevalent in the center of the solid coupons.

> The dendritic segregation is mild and the inclusions smaller and uniformly distributed in the outside metal. The segregated dendrites and agglomerations of inclusions are areas of brittleness through which the tensile tests from the solid center fracture, but these nonuniformities have little effect upon the fracture of the impact bars.

> Specimens from the center of the cored specimens have a uniform distribution of inclusions and slightly less severe dendritic segregation. Consequently, a smaller effect of mass on mechanical properties is observed. The carbon content is reduced slightly at the center of the solid 5-in. section.

Introduction

ONE OF THE CHARACTERISTICS OF CAST STEELS that has evoked considerable interest is the influence of section size or mass upon mechanical properties. It is essential for proper service behavior that the engineer be cognizant of any variation in the properties of castings used in structures or machines. Therefore, quantitative information on these variations in properties and the development of a procedure of obtaining representative test metal are of practical importance.

The effect of size or mass upon the tensile properties of well-fed cast steels were reported by several investigators. Sisco1 reviewed several of these papers and concluded that lower values of strength and ductility in the tensile test are found in large steel castings but that values approaching those attained in lighter sections can be obtained by subsequent heat treatment of heavy sections.

Merten,2 Hull,3 and an A.F.A. committee4 reported lower tensile properties (particularly lower reduction of area and elongation) in heavy than in light cast-steel sections. The first investigator2 concluded that the dimensions of the test coupon should be equivalent to the heaviest section of the casting to

obtain a representative test.

Harper and Stein,5 Lorenz,6 and Sims7 have shown that the structure differed from light to heavy sections or from the surface to center of a heavy section. Considerable work^{2,3,5,6,7} has been done to improve the properties at the center of heavy sections by various types and multiple heat treatments. Some of these treatments have resulted in appreciable improvement in the properties of heavier sections. Nevertheless, Hall⁸ stated that differences in the rate of solidification and cooling between light and heavy sections will lead to considerable differences in physical properties.

It is pointed out in a paper by Briggs and Gezelius9 on the effect of mass on the mechanical properties of cast steels in the as-cast and annealed condition that the efforts of the above investigators were concentrated on "the effect of heat treatment upon the mechanical properties in various sections rather than the effect of the mass of the section upon the mechanical properties." This latter paper is directed towards determining the effect of mass on the mechanical properties of medium-carbon and medium-manganese steels in from 1-in. to 8-in. sections.

It is concluded that "There is a loss in strength

† Major, U. S. A. Ordnance Corps.

^{*} This work was the subject of a Master's Thesis at the Massachusetts Institute of Technology.

^{**} Metallurgist, in charge of foundry operations, at Watertown Arsenal.

[†] Associate Professor of Mechanical Metallurgy at Massachusetts Institute of Technology.

J.

of

ca

T

fre

slo

eq

fac

co

ho

tri

co

TI

co

th

and ductility, measured at the center of the section, as the mass increases." Some recent papers^{10,11} have reported a slight loss of strength and considerable decrease in ductility in heavier sections for normalized as well as quenched and tempered steels.

The influence of increasing section size upon the impact resistance of well-fed cast steels, however, presents a somewhat different picture. The references that investigated impact properties as well as tensile tests to determine the influence of mass and heat treatment on the mechanical properties noted that either the impact resistance improved^{6,9} with increasing section size or that there was no definite trend^{10,12}. The data from Lorig's discussion¹² as replotted by Briggs and Gezelius show increasing impact resistance with increasing section size of the casting but the majority of the impact resistance of specimens from the light section, attached coupons and separate keel blocks are higher than those from the heavier sections.

The results of impact tests must be interpreted with care. Impact tests (either Izod, V-notch or keyhole Charpy tests) show a transition from ductile (fibrous fracture appearance) to brittle (crystalline or granular fracture appearance) as the testing temperature is decreased. The transition temperature is accompanied usually by a sharp reduction in the amount of energy absorbed during fracture. The transition temperature and energy absorbed for any given steels depends upon the sharpness of the notch and the rate of loading used in any particular test. It has been shown¹⁴ that the transition temperature is raised and the absorbed energy decreased with increasing strength. Other work¹⁶ indicates that the transition

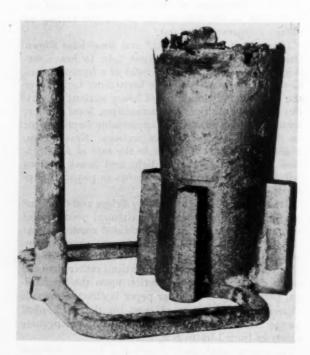


Fig. 1-As-cast test coupon casting showing gates and riser.

temperature and absorbed energy is lower in the keyhole notch Charpy impact test than either the Izod or V-notch Charpy. In the light of these effects, it is apparent, as previously shown,^{14,16} that the results of the comparison of tests taken from different types of bars, for steels of different strength, or at only one temperature are apt to be misleading.

When the impact data are re-evaluated in view of the above facts, the limited conclusions that can be drawn differ from those made previously. Since the strength of the heavier sections is generally lower, the higher energy absorbed during ductile fracture of these bars may be expected under some conditions and does not indicate intrinsically superior toughness for the metal from these sections. In fact, the only significant results¹² found show that the transition temperature of the heavier sections of a casting is probably higher than for the lighter pieces.

It is believed that the impact properties are of greater significance in determining the service behavior of machines and structures than the per cent reduction of area or elongation reported from the tensile test. This hypothesis is advanced by Herres and Jones. 15 True tensile loading occurs infrequently and large deformations before failure are extremely rare, so that little elongation under tensile loading is required. Most failures examined are usually of a brittle nature and originate at a notch, fatigue crack, or other discontinuity. Actually, steel that has the ability to deform plastically under nonaxial loads and in the presence of stress concentrators is required. This property is more nearly measured by the notch bar impact tests. Impact resistance is not essential in all structural parts but is required where stresses are concentrated, particularly at high rates of strain and low temperatures. Previous work¹³ has shown also that a higher endurance limit is associated with higher impact resistance.

The causes of this variation in properties with changing section size in well-fed castings is of interest to metallurgists. Briggs and Gezelius[®] state that "microstructure, carbon segregation, and density values are responsible for the decrease of mechanical properties as the mass increases." A considerable amount of microscopic work and some chemical segregation analyses were carried out in this paper to determine the influence of these factors.

Test Procedure

Test Coupons—To investigate the effect of section size a 5-in. diam cloverleaf type of test coupon similar to that developed by Wayne, Bishop, and Taylor¹¹ was employed. The coupon was bottom-gated and well-fed by a top gravity riser as shown in Fig. 1. The soundness of this coupon was investigated. Sections were cut through the center of a solid block, polished and subjected to magnetic particle examination. In addition prints were made from an X-ray plate taken through a $\frac{5}{16}$ -in. thick central slab.

It is to be noted that the shrinkage cavity does not penetrate into the cloverleaf coupon in any of these methods of examination. Certainly, a test as severe as the radiographic examination of a longitudinal slice EEL

ev-

or

is

of

of

ne

of

be

he

he

of

118

SS

ly

m

is

of

e

11

le

of the center of the block would reveal any shrinkage cavities that existed. The test pieces on the outside were I in. square and positioned to facilitate removal. Test bars were obtained from the center of this coupon by sectioning in a mechanical hacksaw. The test metal from the outside was quickly solidified; metal from the center was solidified at a correspondingly slower rate. The diameter of this type of coupon can be adjusted to obtain a solidification rate that is equivalent to any desired thickness of casting.

A means of determining some of the causes of the variations in mechanical properties between the surface and center of cast sections was devised during the course of this work. This consisted of inserting a hollow, cylindrical shaped core within and concentric with the 5-in. diam coupon. The location of this core is shown schematically in Fig. 2. The core was a split and pasted, wire reinforced, baked sand core composed of a sand, silica flour, and core oil mixture. The core was designed to act as a barrier to the segregation of impurities into the center of the section without appreciably affecting the solidification and cooling rate of the center of the coupon. Undoubtedly, the solidification rate was increased to some extent by the core although no quantitative data are available on this subject.

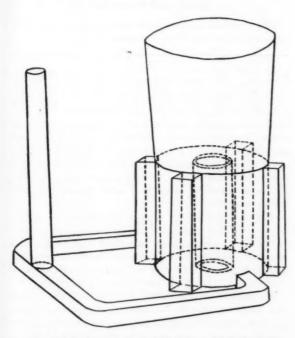


Fig. 2-Sketch indicating location of central core.

Melting and Casting—Four heats of test blocks were cast to various commonly used alloy steel analyses. The approximate compositions are listed below:

Elements, %									
Heat	C	Mn	Si	Cr	Mo	Ni	S(max)	P(max)	Al
R	0.30	0.60	0.40	0.60	0.30	1.75	0.025	0.015	2 lb/ton
S	0.30	0.80	0.30	0.80	-	1.50	0.025	0.015	1.5 lb/ton
T	0.28	0.75	0.30	1.50	0.50	-	0.025	0.015	1.5 lb/ton
U	0.30	0.75	0.40	0.60	0.30	2.00	0.025	0.015	1.6 lb/ton

Several solid and several cored 5-in. diam test coupons (to a total of 10) were poured from each heat. Ladle analyses were taken before the coupons were poured.

Four 700-lb heats were melted in a high frequency, acid-lined induction furnace powered by a 960 cycle, 330 KVA motor generator set. The charge consisted of excellent quality bar melting stock, pig iron, and ferro-alloys. Composition of the charge was controlled by weighing the constituents and allowing for losses of the oxidizable elements. No slag was added and the slag that formed was removed at convenient intervals. After a pouring temperature of approximately 3050 F (optical pyrometer measurements) was obtained the heat was poured into a lip-pour ladle and deoxidized with calcium silicon followed by an addition of commercially pure aluminum wire. The metal was poured from the ladle into the test coupon sand molds at a temperature that varied from 3000 to 2800 F depending upon the order in which the molds were filled.

The ten rammed sand molds in which each heat of test blocks were cast consisted of AFS No. 60 grain fineness washed silica sand, with a 3½ per cent moisture content and a 3 per cent Western bentonite bond. Mold surfaces were air dried overnight before pouring. The gates of the mold (see Fig. 1) were designed so that the castings were bottom poured and filled with small turbulence. Each riser was covered with an antipiping insulation to prevent the top surface from freezing before the casting was completely solidified. The behavior of the cores was observed through the open riser as each mold was poured. If any indication of break-up occurred, the coupon was discarded.

Sectioning and Heat Treatment of Test Coupons— The castings were permitted to remain in the sand molds until almost cold, then were shaken out, and the gates and risers removed. Four 1-in. square test pieces were cut from the outside and two 1-in. square

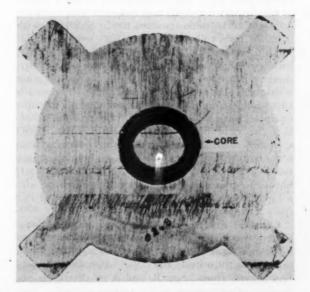


Fig. 3—Saw cut cross-section of casting with central core showing coupon location.

J.

R-1

R-2

R-3

R-9

R-4

R-5

R-6

R-7

R-7

R-8

R-8 R-9

R-1

R-1

R-1

R-1

R-2

R-5

R-6

R-7

S-1-

S-3-

S-3-

S-4-

S-5-

S-5-

S-6-S-7-

S-8-

S-8-

5-9-

S-10

S-10

hea

res

the

tice

ret

fair

sho

ste

dir

son

He

Bri

and

bars were cut as close as possible to the center of each solid test block; two outside pieces and the 1-in. diam bar of metal from the central core were removed from the cored blocks. The center pieces in the cored castings were completely free when the riser was removed (see Fig. 3).

Test bars were normalized after holding at 1950 I for 6 hrs, austenitized at 1650 for 1 hr, water quenched and tempered as shown in Table 1. The austenitizing temperatures were selected by experiment to yield a fine grain size and satisfactory carbide solution. The different analyses required various tempering treatments to obtain the desired hardness level.

TABLE 1-HEAT TREATMENT AND MICROSTRUCTURE

Heat				ASTM Grain Size Number	
R	1200	2	253	7-9	95
S	1150	1	257	7-10	85
T	1280	1	231	7-9	100
U	1215	2	253	8-10	95

It is to be noted that all specimens were cut from the blocks before heat treatment and all pieces heat treated were approximately 1-in. square by 5 in. long. This procedure was followed in order to subject the various sections of the test coupon to identical heat treatments and to determine the intrinsic quality of the steel from these different locations. The properties obtained for the metal from the center of the coupon should not be considered as those that would be obtained were the test block heat treated in its original size.

Mechanical Testing-Each outside piece was machined into one tensile or two V-notch Charpy impact tests; the two bars from the solid center of the coupon were used for both tensile and impact bars; and the cored center piece was finished into either a tensile or two impact bars. All test specimen bars were the ASTM 0.505-in. diam standard tensile specimens with threaded ends. The impact test bars were the usual 0.394-in. square 25/32-in. long V-notch Charpy bars with a 0.010-in. radius notch. The two impact bars from each coupon were located one above the other within the 5-in. long heat treated piece. The tensile bars were fractured at room temperature in a machine equipped with an autographic strain recorder to obtain the yield point. The V-notch Charpy impact specimens were broken at temperatures ranging from room temperature down to -195 C to obtain a transition of the fracture appearance from completely fibrous to granular. An impact testing machine with a capacity of 217 ft-lb and a striking velocity of 16.8 fps was employed.

Chemical Analysis and Microscopic Examination— Chemical analyses and microscopic examinations were made on several of the specimens. Chemical tests were taken from the outside and center of a solid and cored test broken from heats R and S to determine the amount of segregation of the various elements. Sam ples for microscopic examination were removed from the outside and center of the solid test coupons from all four heats in the as-quenched condition to determine the austenitic grain size and percentage of martensite after quenching. Photomicrographs of representative specimens from the outside, solid center, and cored center were taken at 100 and 1000 diameters magnification in the tempered condition to determine the location of the nonmetallics, dendrites, and the principal microstructural constituents.

Representative tensile and impact specimens were selected after breaking from heats R, S, and T for microscopic examination of the fractured surfaces. The samples are from the outside and center of solid coupons and include tensile tests together with fibrous and granular fractured impact bars. Six specimens per heat or a total of 18 were examined. The surface was nickel plated for protection and the bars were sectioned across the center of the fracture. The sectioned test bars were mounted in bakelite and examined under the microscope at 100X, first with only a very light picral etch (referred to as unetched), and then after etching with 4 per cent picral (etched). Photomicrographs were taken of all 18 specimens in both the etched and unetched condition. The unetched samples reveal the position of the nonmetallics and the etched samples show the dendrites and some of the larger inclusions.

Results and Discussion

The chemical compositions obtained from the ladle analyses of the four heats are shown below:

Elements, %									
Heat	C	Mn	Si	S	P	Ni	Cr	Mo	Al
R	0.265	0.64	0.325	0.017	0.014	1.69	0.63	0.28	0.055
S	0.275	0.75	0.11	0.019	0.013	1.50	0.82	No.	0.032
T	0.19	0.63	0.360	0.023	0.015	-	1.62	0.495	0.029
U	0.27	0.70	0.39	0.017	0.010	2.05	0.60	0.34	0.050

With the exception of the lower carbon contents, the desired composition was obtained. The sulphur and phosphorus content is uniformly low in all heats. Ali steels possess sufficient hardenability to transform to a completely martensitic microstructure when a 1-in. section is quenched in agitated water from the austenitizing temperature.

The majority of the results were analyzed in two steps: first, the variations in the mechanical properties from the inside to center of the 5-in. section was observed; and second, the changes in the microscopic appearance and chemical segregation analyses were studied to explain these variations in properties.

Mechanical Test Results—To compare the tensile and impact properties of different cast steels, a similar heat treatment, grain size, microstructure, deoxidation practice, and hardness should occur in all steels. These variables have been shown^{13,14,17,18} to influence the relationship of hardness to ductility and impact resistance. Table 1 contains the tempering treatments, ASTM grain size, average Brinell hardness numbers, and the percentage of martensite for all four heats. Each specimen was normalized, quenched, tempered, and water quenched from the temper in small sizes.

The as quenched microstructure was essentially martensitic as required for optimum mechanical properties.¹³ A uniformly fine grain size existed in all

EL

Cr-

arrend TS ne he re or es. id us er a ·Cd ed ry n 0h d d d

e

5290

TABLE 2-TENSILE TEST RESULTS

Speci- men No.	Yield Strength 1% Offset,	Tensile Strength,	Elong.,	R.A.,	Fracture	Speci- men No.	Yield Strength 1% Offset,	Tensile Strength,	Elong.,	R.A.,	Fractur
	psi,	psi	0%	%			psi.	psi	%	%	rincia
	I. HEAT	R (CR-NI-)									
A	. Tensile Bar			stings			II. Hi	EAT S (CR-N	STEEL) (Cont.)	
R-1-1	106,250	124,500	19.0	53.8	C.P.*	В. 1	Tensile Bars f				it Cores
R-1-2	106,875	125,750	19.0	55.7	C.P.	S-1-S	101,875	122,500	16.5	43.6	C.P.
R-2-1	107,500	125,750	18.0	54.8	C.P.	S-3-S	100,000	119,750	14.0	37.8	C.P.
R-2-2	107,500	126,000	16.5	43.3	C.P.	S-5-S	108,750	125,500	16.0	41.8	C.P.
R-3-1	105,000	124,500	18.5	53.0	C.P.	S-7-S	105,000	122,750	13.5	35.6	C.P.
R-3-2	108,125	127,250	17.5	50.5	Irr. Br.	C	Tensile Bars	from Cente	r of Castin	igs with	Cores
R-4-1	104,375	123,500	18.0	51.0	C.P.	S-4-C	111,875	127,500	14.5	37.5	C.P.
R-4-2	108,750	126,750	18.5	55.1	C.P.	S-6-C	108,750	125,500	14.5	39.3	C.P.
R-5-1	105.625	124,500	16.5	44.8	C.P.	S-8-C	116,250	131,500	14.0	37.2	C.P.
R-6-1	106,875	125,500	17.0	51.0	C.P.	S-9-C	110,000	127,750	14.5	37.8	C.P.
R-6-2	105,000	124,250	16.5	41.5	Irr. Br. **		III.	HEAT T (C	R-Mo STE	EL)	
1-7-1	107,500	127,000	17.5	47.7	Irr. Br.			Bars from			S
2-7-2	105,000	124,000	16.5	49.1	C.P.	T-1-1	92,500	111,250	20.0	59.8	C.P.
1-8-1	105,625	125,000	17.5	49.4	C.P.	T-1-2	95,000	113,250	21.0	57.5	C.P.
1-8-2	107,500	126,750	17.5	48.8	C.P.	Т-3-1	93,750	111,250	21.0	63.2	C.P.
2-9-1	106,875	125,250	14.5	30.7	Irr. Br.	T-3-2	95,000	112,750	20.5	61.3	C.P.
2-9-2	106,250	124,250	18.0	53.5	Irr. Br.	T-5-1	91,250	110,000	21.5	63.2	C.P.
-10-1	105,625	124,750	17.5	45.9	Irr. Br.	T-5-2	88,125	108,250	21.0	63.9	C.P.
-10-2	110,000	127,500	14.5	36.3	Irr. Br.	T-6-1	94,375	112,250	20.5	61.5	C.P.
	Tensile Bars f					T-6-2	95,000	113,500	20.5	61.0	C.P.
t-1-S	101,250	120,250	13.0	24.7	Irr. Br.	T-7-1	88,375	108,000	21.5	60.8	C.P.
1-9-8	103,750	121,750	12.5	24.0	Irr. Br.	T-7-2	90,000	109,000	22.0	65.3	C.P.
1-10-S			14.0	26.4		T-8-1	92,500	111,000	21.0	60.3	C.P.
	101,875	121,250			Irr. Br.	T-8-2	92,500	110,750	21.0	59.5	C.P.
1-2-C	Tensile Bars		16.5	41.2			ensile Bars fr				
	105,625	123,500 125,250			C.P.	T-1-S	90,625	108,250	18.5	42.4	Irr. Br.
R-4-C	107,500	125,200	15.5	38.7 40.3	C.P. C.P.	T-3-S	88,375	107,500	19.5	48.5	Irr. Br.
1-5-C	106,875		15.5	39.0		T-5-S	92,500	109,500	19.0	46.9	Irr. Br.
R-6-C	106,250	124,350	15.5	1000	Irr. Br.	T-7-S	86,250	105,250	18.0	49.9	Irr. Br.
1-7-C	106,875	125,250	16.0	38.4	Irr. Br.		Tensile Bars				
C-8-C	106,250	123,500	17.0	46.2	Irr. Br.	T-6-Q	91,250	108,250	19.5	58.5	C.P.
	II.	HEAT S (C				T-8-C	88,750	108,000	15.5	49.3	C.P.
		Bars from				1-0-0		IEAT U (CR-			C.F.
-1-1	109,375	127,500	18.0	49.3	C.P.	Α	Tensile Bars				etinas
-1-2	110,625	127,750	17.5	47.2	C.P.	U-2-1	106,250	123,250	18.5	54.6	C.P.
-3-1	107,500	125,000	18.5	51.9	C.P.	U-2-1 U-2-2	100,230		19.5	55.1	
-3-2	106,875	124,750	16.5	46.9	C.P.	U-5-1		123,750 121,750	18.0		C.P.
4-1	111,250	128,500	17.0	48.5	C.P.		104,750			54.1	C.P.
4-2	110,000	126,250	17.5	48.5	C.P.	U-5-2	104,500	121,500	17.0	49.6	C.P.
5-1	108,125	125,500	16.5	49.6	C.P.	U-6-1	107,500	123,750	18.5	54.1	C.P.
5-2	106,875	125,250	17.5	49.6	C.P.	U-6-2	106,250	123,000	18.0	51.3	C.P.
-6-1	114,375	130,000	15.5	46.9	C.P.		ensile Bars fr		0		
-7-1	112,500	128,500	17.5	49.6	C.P.	U-5-S	98,750	117,500	8.5	35.3	Irr. LGC
7-2	111,875	128,000	17.5	51.9	C.P.		Tensile Bars				
8-1	114,375	130,000	16.5	49.3	C.P.	U-2-C	107,000	123,500	16.5	45.1	Irr. SGC
8-2	112,500	128,000	16.0	49.9	C.P.	U-6-C	108,750	125,250	16.0	44.5	Irr. P.
9-1	115,000	131,250	16.0	48.2	C.P.	* Cuppe	ed and Pitted				
9-2	113,125	130,000	16.5	49.1	C.P.	** Irregu	lar Break				
10-1	108,750	128,000	17.5	47.1	C.P.	† Irregu	lar, Large Ga	s Cavities			
10-2	113,750	129,750	16.5	48.0	C.P.		lar, Small Gas				

heats—an additional requirement for optimum impact resistance.¹⁸ All tempering temperatures were above the reported embrittling range. Similar furnace practice was employed and the percentage of aluminum retained as residual aluminum metal was within a fairly narrow range and close to that which has been shown to be preferable for good toughness in cast steels.¹⁹ The hardness of heats R, S, and U were directly comparable. This permits a direct comparison of the mechanical properties of these three heats. Heat T, however, has a hardness of from 20 to 25 Brinell Hardness Number less than the other three and, therefore, a slightly higher ductility and level of

absorbed energy during fibrous failure.

The complete tensile test results for heats R, S, T, and U are listed in Table 2. The specimens have been arranged in three groups from each heat: the first group consists of those tensile bars taken from the outside coupons; the second, the bars taken from center of castings without cores; and the third, the tensile bars taken from the center of the centrally located sand core. The following averages of yield and tensile strength, per cent elongation and reduction of area for the three groups of specimens in each heat are listed in the following table.

	No. of	0.1% Offse	et	%	%
Heat R	Tests	Y.S. psi	T.S. psi	Elong.	R.A.
Outside coupons	19	106,800	125,000	17.3	48.0
Solid-center coupons	3	102,300	121,000	13.0	25.0
Cored-center coupons	6	106,560	125,000	16.0	40.7
Heat S					
Outside coupons	17	111,000	127,400	17.0	49.0
Solid-center coupons	4	104,000	122,000	14.9	38.0
Cored-center coupons	4	112,300	129,000	14.0	38.0
Heat T					
Outside coupons	12	92,200	111,000	21.0	59.7
Solid-center coupons	4	89,500	108,000	18.7	47.0
Cored-center coupons	2	90,000	108,100	17.5	53.9
Heat U					
Outside coupons	6	106,200	122,800	18.3	53.1
Solid-center coupons	1	98,750	117,500	8.5	35.3
Cored-center coupons	2	107,875	124,375	16.3	44.8

These results show clearly that the outside coupons exhibit the best tensile properties in every heat. The coupons from the center of the solid cast test block had a somewhat lower yield and tensile strength in all cases than those from the outside. The ductility was markedly less at the center of the solid 5-in. section. The reduction of area value at the center decreased from 80 per cent to nearly 50 per cent of the value at the outside on the different heats. It is notable that the ductility of specimens from the outside of the various heats are similar for all heats except for the variation caused by the lower strength of steel T.

The tensile test results from within the sand core were affected by the presence of this core. The yield and ultimate strengths were similar to the outside values; the ductility was intermediate between the results obtained from the outer and solid center tests.

Fig. 4-Specimens from casting without core.

Some quantitative variation in results from heat to heat is observable but the trend of results is the same for the three locations.

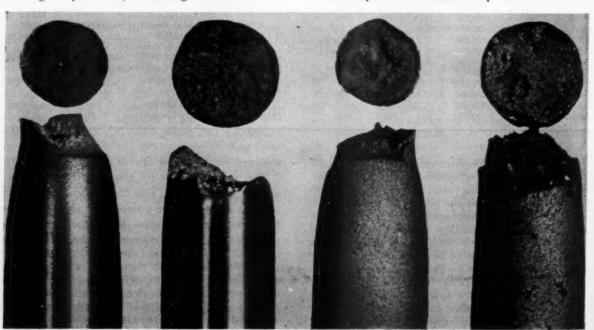
The fracture appearance of the tensile test bars is further evidence of the decrease in ductility that accompanies increasing section size. Most of the test bars taken from the outside of the castings broke with a characteristic cup and cone and a necked-down section adjacent to the fracture. Those specimens from the solid casting centers fractured with considerably less reduction of area and the breaks were generally very irregular. Many of the specimens removed from the core, however, showed appreciable necking. Typical center and outside specimens from castings without and with the central core are shown in Fig. 4 and 5 respectively.

The size of the fractured areas in these figures serves as a comparison of the reduction of area of each test. The two etched specimens in each figure were etched similarly, so that direct comparisons of the etched appearance can be made for both types of castings. It is apparent from both the fracture appearance and amount of pitting during etching that, in the case of the castings without the core, the metal taken from the outside bars is more uniform than that from the center. On the other hand, the difference in appearance from outside to center is not nearly as marked for the specimens from the cored casting.

The loss in strength and ductility at the center of the test coupon must be attributed to the quality of the metal at the center of the 5-in. section since the heat treatments received by all pieces are similar. The results are in agreement with the bulk of data reported in the literature^{2,3,4,9,10,11} although the decrease in strength and ductility is more than that reported by some investigators.^{6,12}

The impact test results are reported as transition

fr (in an fo m



EEL

to

15 at est ke wn ins erered ıg. gs ig. 'es of re of of T. in al an

T-

ot

of

of

16

ne

e-

eat

n

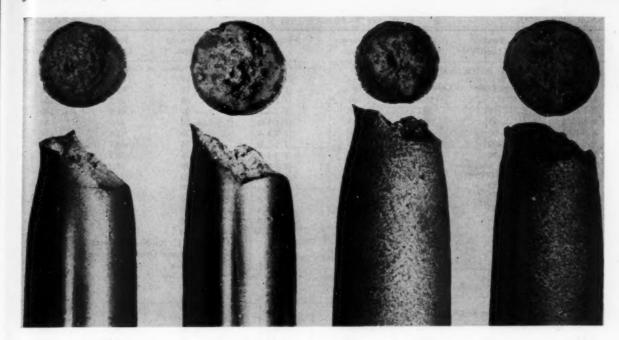


Fig. 5-Specimens from casting with core.

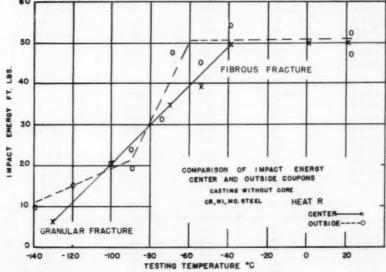
(Right) Fig. 6

curves for the four steels – i. e., graphs of energy required for fracture vs testing temperature over a sufficient range of temperatures to change the fracture appearance from fibrous (shear) to granular (cleavage). The results are listed in Table 3; the transition curves are contained in Fig. 6, 7, 8, and 9

for heats R, S, T, and U, respectively. The approximate energy required for ductile fracture and the transition temperature* for each steel are listed in the table below:

Heat	Outsid	e Tests	Solid-Ce	enter Tests	Cored-C	Cored-Center Test		
	Energy, ft-lb	Trans. Temp, °C	Energy, ft-lb	Trans. Temp, °C	Energy, ft-lb	Trans. Temp, °C		
R	51	60	50	_40	none t	aken		
S	44	-20	42	-20	none t	aken		
T	62	65	54	55	none t	aken		
U	52	-85	48	80	50	-85		

^{*} Transition temperature is defined as the lowest temperature at which a fully fibrous appearing fracture is obtained.



The tests from the outside metal exhibit slightly higher absorbed energy during ductile fracture and a slightly lower transition temperature. The over-all differences in impact resistance, however, are small.

The relative toughness of the steel from the outside and center of the coupons is best compared for the four heats by an examination of the transition curves in Fig. 6 through 9. Even with the expanded testing temperature scale used, the curves for the outside and center metal lie very close to one another. The single transition curve of samples taken from the central core of heat U has approximately the same transition temperature and absorbed energy level as that from the outside coupons. No evidence of the higher energy absorbed during fracture of the metal from the heavier sections, as previously reported, 6.9 was found. In

Specimen No.	Testing Temp. °C	Ft-lb Charpy	% Fibrous Fracture	Specimen No.	Testing Temp. °C	Ft-lb Charpy	% Fibrou Fracture
	I. HEAT R (CR-N	I.Mo STEEL		B. Specin	nens from Outside o	f Castings wit	hout Cores.
	,			T-1-3-A	40	62.8	100
A. Specim	ens from Center of	Castings with	out Cores.	T-1-4-A	- 55	60.0	100
R-1-I-A	0	50.6	100	T-7-4-A	60	67.5	100
R-1-I-B	_ 40	50.1	100	T-1-4-B	— 65	61.9	100
R-10-I-A	— 55	39.5	85	T-1-3-B	70	47.9	80
R-9-I-A	70	35.4	70	T-7-3-A	— 70	45.8	70
R-9-I-B	100	20.8	15	T-5-3-B	— 75	64.7	100
R-10-I-B	_130	6.4	0	T-5-4-B	— 75	32.2	25
				T-7-3-B	— 75	38.2	40
B. Specime	ns from Outside of	f Castings with	out Cores.	T-5-3-A	80	42.4	60
R-10-4-B	+ 22	47.5	100	T-3-3-A	- 90	29.1	20
R-1-3-B	_ 40	54.6	100	T-3-3-B	-100	21.1	15
R-10-4-A	55	45.3	100	T-3-4-A	-120	15.1	5
R-1-4-A	— 70	47.9	100	T-3-4-B	-130	13.0	0
R-10-3-B	— 75	31.4	70	T-5-4-A	_ 70	32.2	40
R-9-4-A	80	25.0	50		IV. HEAT U (CR-)		
R-9-3-B	— 90	19.4	25	Λ.			ings.
R-10-3-A	90	24.3	50	U-3-1-1	+ 21.5	54.6	100
R-9-3-A	-100	19.8	30	U-8-1-1	+ 21.5	51.9	100
R-1-4-B	_120	16.4	10	U-8-1-2	- 40	50.1	100
R-9-4-B	_140	10.0	0	U-3-1-2	40 40	53.7	100
K-3-1-B		10.0		U-8-2-1	70 70	49.2	100
	II. HEAT S (CR	-NI STEEL)		U-3-2-1	- 70 - 70	51.9	100
A. Specim	ens from Center of	Castings with	out Cores.			46.2	100
S-3-I-A	0	42.4	100	U-9-1-2	- 80 - 80	49.2	100
	20	42.4	100	U-4-1-2			
S-1-I-B			100	U-4-2-2	— 85	46.6	95
S-7-I-A	30	37.0	==	U-9-2-1	_ 90	44.1	100
S-1-I-A	- 40	33.0	70	U-4-2-1	90	42.4	90
S-7-I-B	— 60	19.4	20	U-10-1-1	- 95	39.1	70
S-3-I-B	— 70	16.8	30	U-4-1-1	-100	32.6	70
S-5-I-A	-100	7.5	0	U-9-1-1	100	31.8	60
S-5-I-B	130	4.8	0	U-8-2-2	120	19.4	30
D Specime	ne from Outside of	Castings with	out Cores	U-3-2-2	-120	23.6	40
	ns from Outside of			U-9-2-2	150	14.8	10
S-3-3-A	+ 24	43.2	100	U-7-1-1	-150	17.4	10
S-1-3-B	20	44.1	100	U-10-1-2	195	11.2	0
S-1-4-A	30	39.5	100	U-7-1-2	-195	9.2	0
S-1-3-A	_ 40	37.0	90	B. Specia	nens from Center of	Castings with	out Cores.
S-5-3-A	60	20.8	25	U-I-S-1	+ 21.5	52.3	100
S-5-3-B	— 60	26.5	40	U-3-S-2	80	42.4	. 100
S-1-4-B	— 70	21.8	30	U-7-S-1	90	40.7	70
S-3-3-B	100	7.5	0	U-3-S-1	-100	23.6	50
S-3-4-B	130	5.8	0	U-1-S-2	-120	18.1	30
				U-7-S-2	150	10.6	10
	III. HEAT T (C)	R-MO STEEL)			imens from Center of		
A. Specime	ens from Center of	Castings witho	out Cores.	U-4-C-1	+ 21.5	52.8	100
T-1-I-A	_ 40	51.4	100	U-4-C-2	40	48.8	100
T-3-I-A	- 40 55	55.9	100	U-9-C-1	80	43.2	100
T-7-I-B	- 60	31.1	50	U-10-C-1	85	42.4	100
				U-9-C-2	_ 90	42.8	95
T-7-1-A	— 65 70	47.5	95				80
							40
							20
T-1-I-B T-3-I-B T-5-I-B	— 63 — 70 — 90 —120	38.2 19.1 9.4	85 20 0	U-8-C-2 U-8-C-1 U-10-C-2	—100 —120 —150	35.0 24.0 17.7	

this case the differences in strength from outside to center were small and the better toughness was exhibited by the outside specimens.

The similarity of the impact resistance of the specimens from the outside and center of the test coupon is surprising in view of the great differences in the tensile ductility of these steels. The amount of energy absorbed during fibrous fracture is believed to be associated with the per cent reduction of area obtained in the tensile test.²⁰ There appears to be a slight correlation between the tensile ductility and impact resistance; heat R, with the greatest spread in reduction of area from center to outside, shows the

greatest difference in transition temperature; and heat S, with the least difference in ductility, has the smallest difference in the transition temperatures.

Generally, the correlation between the two is much smaller than anticipated. In steel R, for example, where the per cent reduction of area at the center is reduced to almost one half of the value at the outside, the impact resistance is similar at both these locations. The same lack of agreement between tensile ductility and impact resistance is evident also in heats U, S and T. Apparently, whatever reduced the tensile ductility of these cast steels at the center of the coupon has considerably less effect on the impact resistance.

d

h

is

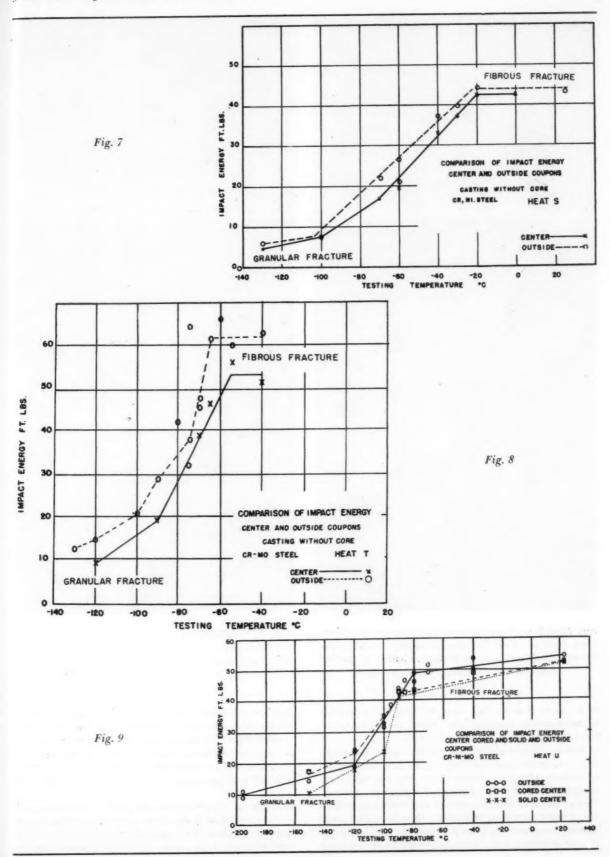
,

1-

e

e

n



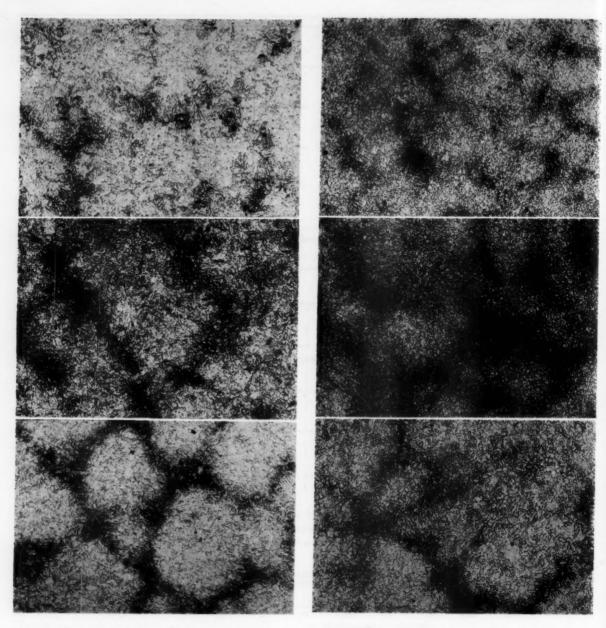


Fig. 10-Photomicrographs of heat treated specimens from Heat R. Top, Outside of casting; Middle, Center of casting with core; Bottom, Center of casting without core. Picral etch. Mag. 100x.

Fig. 11—Photomicrographs of heat treated specimens from Heat S. Top, Outside of casting; Middle, Center of casting with core; Bottom, Center of casting without core. Picral etch. Mag. 100x.

It is noted that, in this case, the lack of correlation between tensile ductility and impact resistance is not the result of different heat treatments.

The toughness of the four steels is generally good. The energy absorbed during ductile fracture is commensurate with the hardness level and the transition temperatures are low. This good toughness is shown by the steels from the center of the section as well as the outside metal. It is evident, therefore, that satisfactory toughness can be obtained in the center of slowly solidified 5-in. sections as well as at the more

quickly solidified surfaces, providing this metal receives adequate heat treatment and feeding during solidification. This adequate heat treatment can be secured by employing steels of sufficient hardenability to quench to an essentially martensitic microstructure with the proper precautions to prevent embrittlement.

De

ot

th

m

ste

ar

in

dr

fo

ot

Some of the differences in impact test results are of interest. The high energy absorbed during fibrous fracture of heat T is undoubtedly the result of the low strength of this steel. The higher transition tem-

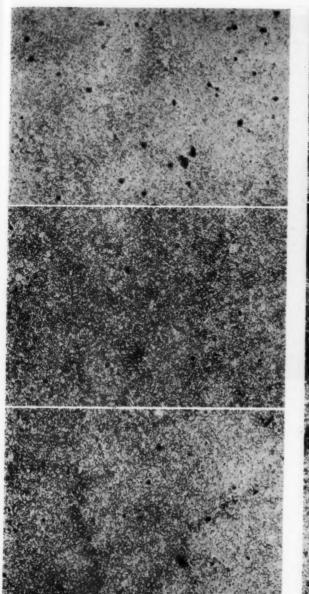


Fig. 12—Photomicrographs of heat treated specimens from Heat T. Top, Outside of casting; Middle, Center of casting with core; Bottom, Center of casting without core. Picral etch. Mag. 100x.

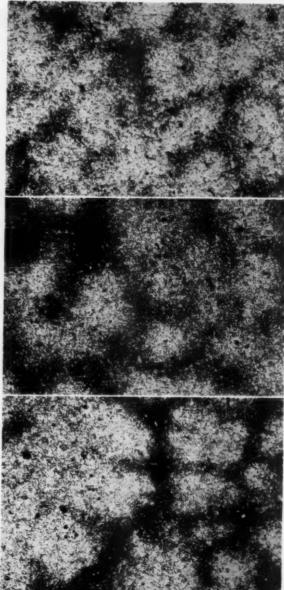


Fig. 13—Photomicrographs of heat treated specimens from Heat U. Top, Outside of casting; Middle, Center of casting with core; Bottom, Center of casting without core. Picral etch. Mag. 100x.

perature of heat S (0 per cent Mo) compared to the other heats lends further weight to the evidence²⁴ that molybdenum is a desirable alloy to obtain optimum toughness. Some of this decreased toughness of steel T, however, may have been caused by the larger amount of nonmartensitic microconstituents present in this heat.

Microscopic Examination—Examination of the dendrites and inclusions at the various locations in the four steels reveals considerable differences from the outside to center of the coupon. The etched photo-

micrographs at 100X of the specimens from the outside and center of the solid and cored coupons are shown in Fig. 10, 11, 12, and 13 for steels R, S, T and U, respectively. The tempered martensitic structures of the four heats are similar. The dendrites are larger and more pronounced in the slowly solidified samples from the solid center than in the quickly solidified outer layers. The metal from the cored central sample has a dendritic structure that was intermediate in intensity between the other two. The inter-dendritic material is apparently that which exists between the



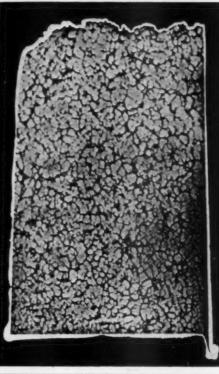


Fig. 14—Photo of etched split half of tensile bar from outside (left) and from solid center (right). Mag. 5x.

cross sections of the dendrite arms as reported by Hawkes and Brown. 22

The inclusions are smaller and uniformly distributed in the metal at the outside but in the center of the coupon have collected into the inter-dendritic areas, frequently in a chain-like arrangement. The size of the inclusions at the center has increased and the number decreased by agglomeration. This concentration of nonmetallics at the inter-dendritic boundary has been observed by Harper and Stein.5 As noted by these authors, no amount of homogenization, annealing, or other heat treatment will reduce this concentration or improve the ductility of the metal. The metal from the cored central tests shows a slight tendency for this agglomeration and segregation of the inclusions but the majority are uniformly distributed. The photographed areas were selected to illustrate the distribution and nature of the inclusions and contain more than the average amount of nonmetallics that appeared in the complete cross-sections.

The inclusions present in the steel are the irregularshaped clusters referred to as Type III by Sims and Dahle.²³ This type of inclusion is undoubtedly the result of using a sufficient addition of aluminum to avoid the collection of sulphides at the primary grain boundary. This distribution permits the attainment of good ductility.

Examination of the fracture path in the test specimens indicates the manner in which these dendrites and inclusions affect the behavior of the tensile and impact bars. Representative photographs of the etched half of tensile bars from the outside and center of the coupon are contained in Fig. 14. These photomacrographs show the fine, light etching dendrites of the ductile tensile bar on the outside and deeper etching, coarser dendrites of the relatively brittle center tensile specimen. Photomicrographs of unetched specimens illustrate the relationship between the fracture

sio

obs

side

TABLE 4-SEGREGATION ANALYSES COMPARING CENTER AND OUTSIDE COUPONS

				Ele	ment, %						
Heat	Location	C	Mn	Si	S	P	Ni	Cr	Mo	V	Al
R	Center bar with core	0.275	0.55	0.315	0.018	0.012	1.73	0.60	0.305	_	0.055
R	Outside bar	0.275	0.54	0.315	0.018	0.012	1.71	0.60	0.305	-	0.05
R	Center bar without core	0.260	0.54	0.320	0.017	0.013	1.69	0.60	0.30	_	0.060
R	Outside bar	0.270	0.55	0.315	0.017	0.013	1.73	0.61	0.305	-	0.062
S	Center bar with core	0.29	0.78	0.36	0.021	0.012	1.50	0.82	_	_	0.032
S	Outside bar	0.29	0.79	0.365	0.019	0.012	1.50	0.81	_	_	0.031
S	Center bar without core	0.285	0.78	0.36	0.019	0.012	1.50	0.82	_	_	0.032
S	Outside bar	0.29	0.80	0.365	0.021	0.012	1.52	0.82	_	-	0.032

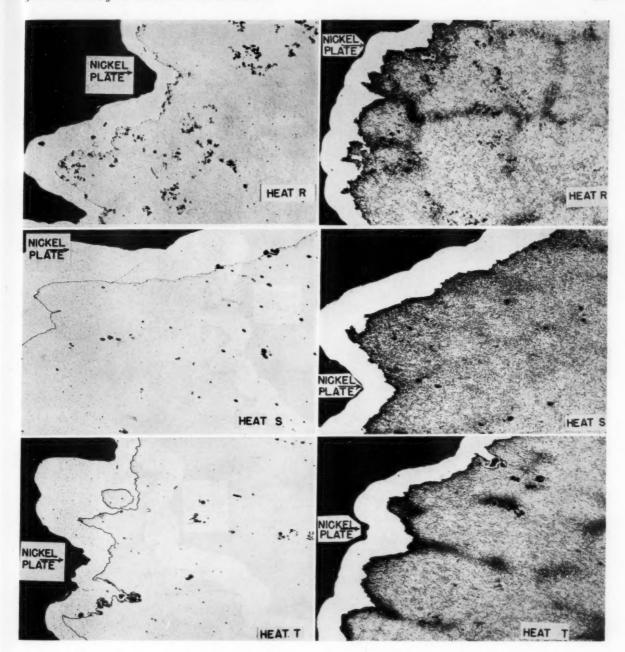


Fig. 15—Solid metal taken from solid center of coupon. Photomicrograph at 100x of tensile test fracture and surrounding metal. Unetched.

Fig. 16—Test metal taken from solid center of coupon. Photomicrograph at 100x of tensile test fracture and surrounding metal. Picral etch.

path and the inclusions; the etched samples demonstrate the influence of the dark etching inter-dendritic material. Usually the etched samples show the inclusions as well as the dendrites but in a few instances the inter-dendritic area has etched so darkly that it obscures the inclusions.

Microscopic examination was conducted on the outside and solid center samples from heats R, S, and T. Figures 15 through 26 containing photomicrographs of the samples from the various locations are shown.

The nickel plate used to protect all fractured surfaces appears as a white area in the photographs.

The photomicrographs of the tensile test from the center of the coupons in Fig. 15 and 16 contain the same type of large, agglomerated inclusions and dark etching inter-dendritic areas shown previously. Examination of the path of fracture demonstrates that the tensile break passes through several nonmetallic inclusions, some of which are arranged in a chain-like form. The path of fracture follows the segregated

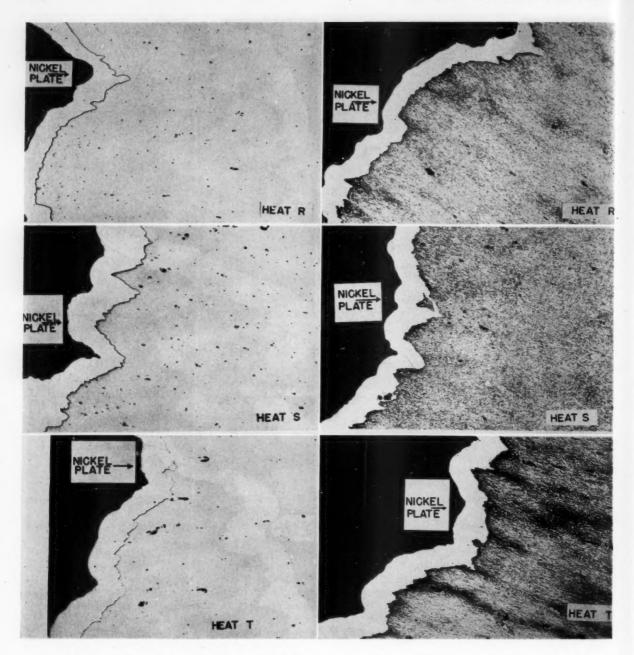


Fig. 17—Test metal taken from outside of coupon. Photomicrograph at 100x of tensile fracture and surrounding metal. Unetched.

Fig. 18—Test metal taken from outside of coupon. Photomicrograph at 100x of tensile test fracture and surrounding metal. Picral etch.

inter-dendritic areas that contain these inclusions and other impurities.

Closer scrutiny of the fracture indicates the presence of tears which occur within these inter-dendritic areas as contained in the center of the photomicrographs from heat R in Fig. 16. It appears that these inter-dendritic areas and inclusions are planes of brittleness in tensile tests cut from the segregated metal. When the tensile bar is loaded, cracks form within these areas and grow as the stress increases.

Eventually the effective load-bearing cross section of the bar is reduced sufficiently so that failure occurs. This cracking rather than ductile flow explains the poor ductility of a tensile bar from the slowly solidified central portions of a heavy casting.

di

cl

in

Obviously, the influence of mass on ductility will vary with the segregation present in the steel. Severe segregation (particularly of inclusions) as shown for heat R will cause a sharp decrease in reduction of area; milder segregation as shown in heat S will result

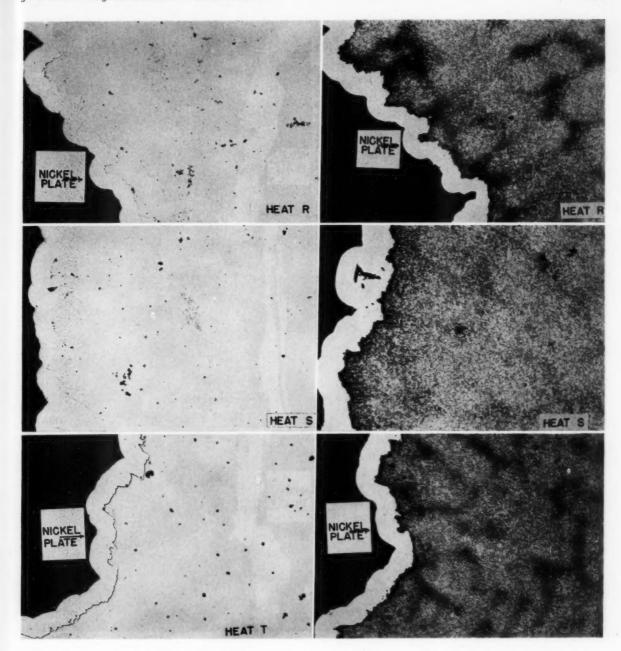


Fig. 19—Test metal taken from solid center of coupon. Photomicrograph at 100x of ductile impact test fracture and surrounding metal. Unetched.

Fig. 20—Test metal taken from solid center of coupon. Photomicrograph at 100x of ductile impact test fracture and surrounding metal. Picral etch.

in a smaller decrease in the ductility of tensile tests from the center of heavy sections.

The tensile test from the outside metal has considerably less segregation and more uniform inclusion distribution. Figure 17 contains the unetched photomicrographs from this steel and there is only a slight tendency for the fracture to follow these scattered inclusions. Some nonmetallics are found in the fracture, as shown in heat T, but there is no evidence of chain inclusions forming planes of brittleness. The segre-

gated areas of the metal from the outside of the coupon are much lighter than from the center and have appreciable tensile ductility. Considerable indication of flow of the dendritic structure in the direction of applied stress is in evidence. There is little or no indication that the inter-dendritic areas lie along the fractured surfaces or reduce the ductility.

Photomicrographs of the path of fracture of the impact bars from the solid center reveal the chain-like clusters of inclusions in the unetched specimens (Fig.

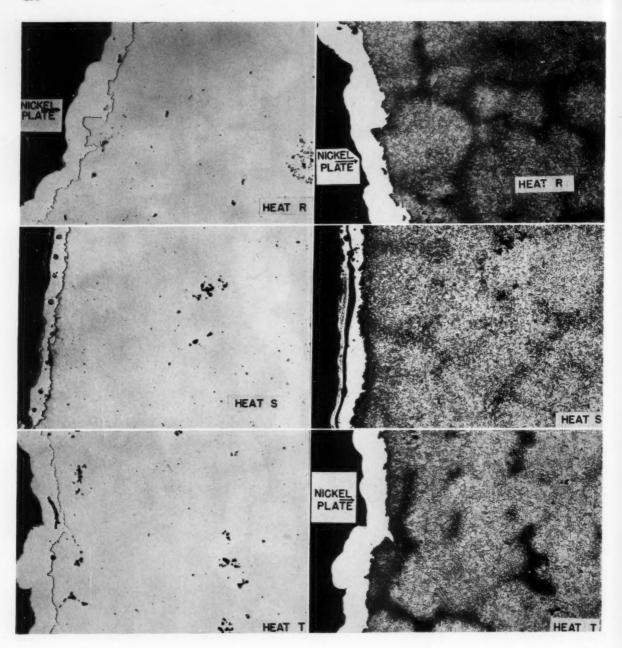


Fig. 21—Test metal taken from solid center of coupon. Photomicrograph at 100x of brittle impact test fracture and surrounding metal. Unetched.

Fig. 22—Test metal taken from solid center of coupon. Photomicrograph at 100x of brittle impact test fracture and surrounding metal. Picral etch.

19 and 21) and the highly segregated inter-dendritic areas in the etched specimens (Fig. 20 and 22). The inclusions and inter-dendritic areas do not appear to have influenced greatly the mode of fracture in either the ductile or brittle impact tests. A slight tendency of the fracture to pass through some of these inclusions and segregated areas is apparent and the impact resistance may have been influenced slightly.

For the most part, as illustrated in steels S and T, the path of rupture avoids the large areas of non-

metallics and segregates. The fracture of the impact test seems to progress so rapidly that there is no appreciable embrittling effect or stress concentration introduced by these nonuniformities. The small influence of inclusions and dendrites upon the fracture of impact bars appears to be the reason for the slight effect of mass upon impact resistance.

The photographs of the brittle and ductile impact bars from the outside of the coupon have the same uniform distribution of nonmetallics and slight tenth

m

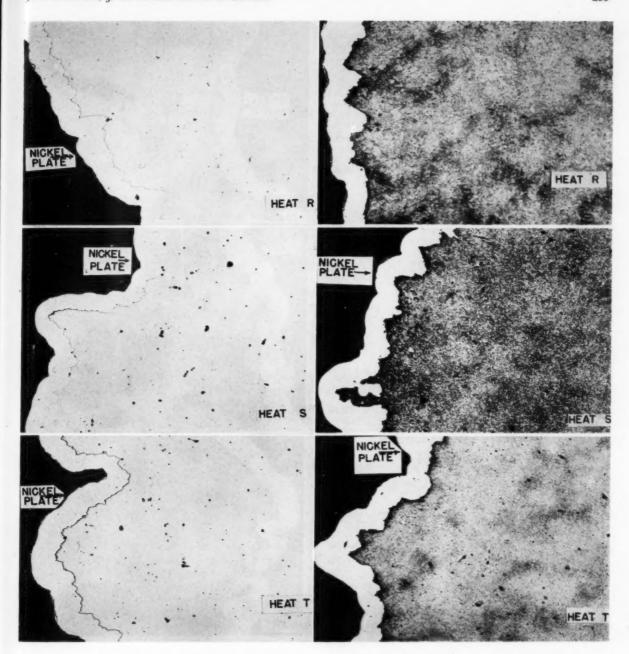


Fig. 23—Test metal taken from outside of coupon. Photomicrograph at 100x of ductile impact test fracture and surrounding metal. Unetched.

Fig. 24—Test metal taken from outside of coupon. Photomicrograph at 100x of ductile impact test fracture and surrounding metal. Picral etch.

dency for segregated inter-dendritic areas previously noted for this quickly solidified metal. Examination of the fractured surfaces (Fig. 23 through 26) makes it clear that neither the nonmetallics nor dendritic pattern influence the path of fracture significantly. In no case was any tendency for the fracture to follow the inclusions or segregated areas observed in either the ductile or brittle failures.

The different structures observed in the photomicrographs of the outside and center metal at 100X magnification is not as apparent at 1000X. Representative samples of each type of metal contained in Fig. 27 have the same finely divided, tempered martensitic microstructure at all locations. The dendritic pattern is much less in evidence at the larger magnification. The photomicrographs illustrate that the same microconstituents were obtained in all specimens.

Chemical Segregation Analyses—The results of the chemical segregation analyses from the outside and center of solid and cored coupons are listed in Table

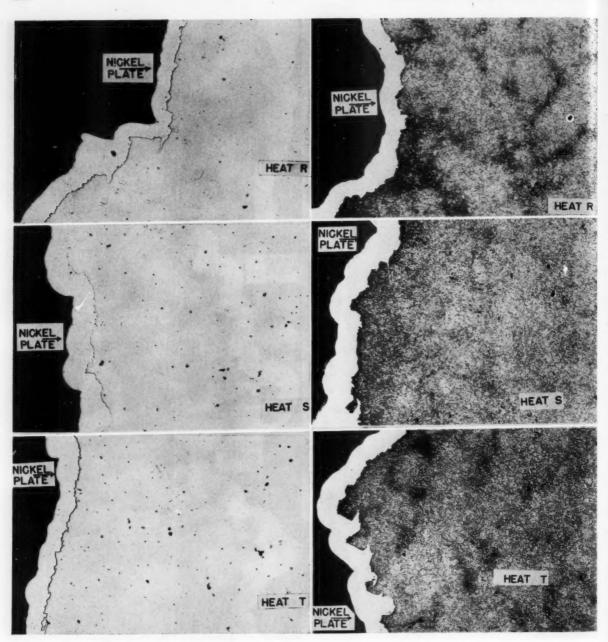


Fig. 25—Test metal taken from outside of coupon. Photomicrograph at 100x of brittle impact test fracture and surrounding metal. Unetched.

Fig. 26—Test metal taken from outside of coupon. Photomicrograph at 100x of brittle impact test fracture and surrounding metal. Picral etch.

4. The only significant segregation is a reduction in the carbon content at the center of the solid coupon. This slight carbon segregation (0.005 to 0.01 per cent) may be one of the factors, as previously reported,9 that causes the loss in density and tensile strength of the center of the coupon. This carbon segregation does not occur in the center of the cored coupon. More carbon segregation would be anticipated in heavier sections.

General Considerations

The results of the microscopic examination of various cast steels offer an explanation of the effect of mass on the mechanical properties of steel castings. The presence of heavily segregated inter-dendritic areas and large agglomerations of nonmetallic inclusions (frequently in a chain-like arrangement) in the slowly solidified metal from the center of the heavy castings results in a slightly lower strength and con-

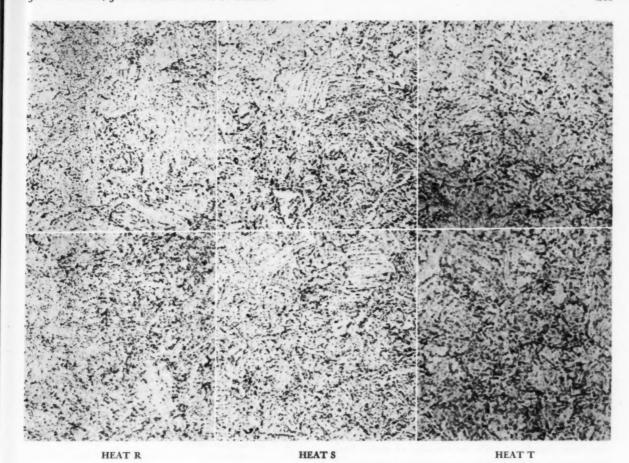


Fig. 27-Photomicrographs of outside (bottom row) and center specimens (top row) in quenched and

drawn conditions without central cores. Picral etch.

Mag. 1000x.

siderable reduction in ductility during tensile testing. The segregation of impurities and accumulation of inclusions in the metal from this location produced planes of brittleness within the test bar. Cracks form at these discontinuities as the load is applied and they lengthen with increasing load, resulting in eventual failure. Consequently, the amount of plastic flow and perhaps even the load-bearing ability of the bar is reduced.

Metal from the rapidly solidified outer portions of heavy castings or thin sectioned castings have a much finer, less segregated, dendritic pattern and a more uniform distribution of finer inclusions. Tensile tests from these locations are, therefore, not appreciably embrittled or weakened by this structure. The results of notch bar impact tests, however, are only slightly influenced by the presence of these large, segregated dendrites and chain-like inclusions in the metal from the central portions of heavy cast sections. Apparently, fracture occurs so rapidly during this type of test that tearing and cracking around these areas of brittleness does not proceed to such an extent that the impact resistance is affected greatly.

The results of this investigation indicate that sound steel castings are suitable for universal use, including highly stressed, shock loaded parts, in spite of these variations in tensile properties. Since there is evidence that notched-bar impact test results are of considerably greater significance in determining the service behavior of steel components, it is believed that heavy, well-fed castings can be produced and processed so that they can perform satisfactorily in severe service.

Perhaps one of the recognized advantages of forging—that it breaks up the dendritic structure of cast steels—is not as essential for some purposes as formerly thought. It is even conceivable in some cases that forging will influence unfavorably the serviceability of a cast part (with its nondirectional properties) by arranging the inclusions so that the part will be more apt to fail in a brittle manner transverse to the forging direction. The tendency for inferior toughness in that transverse forging direction has been reported by Herres and Lorig.²⁴

To obtain the good toughness in the center of heavy sections, such as shown in this investigation, it is necessary to employ a sound casting with the proper microstructural constituents (tempered martensite). The attainment of tempered martensite in heavy sections requires the use of steels of adequate hardenability for the section size and type of quench em-

ployed. Since such highly alloyed steels are subject frequently to temper brittleness,¹⁴ the combinations of alloys must be carefully selected, tempering treatments above the embrittling range used, and the castings quenched from the temper.

Use of the sand core in the center of the coupon provides further information on some of the causes of mass effect, particularly on the tensile results. The absence of any carbon segregation from the outside to cored center sample indicates that the segregation of this element consists of a movement of higher carbon steel from the center towards the outside. The core acted as a barrier to the motion of this metal. It is suggested that this segregation is the flow of higher carbon, inter-dendritic, molten metal through the network of dendrites to provide metal for feeding the faster solidifying outside layers. This type of segregation is similar to that reported by Bishop and Fritz²⁵ for small steel castings. It is suggested further that this segregation is related to the decreased density noted by Briggs and Gezelius9 at the center of heavy sections. In any event, the absence of carbon segregation and possibly the lack of a difference in density eliminated any differences in strength of the specimens from the center and outside of the cored coupon.

The presence of the core appears to reduce the severity of dendritic segregation to a small extent and greatly decrease the tendency for agglomeration of the inclusions in the inter-dendritic areas. No definite conclusion can be reached from the available data on the causes of these differences in structure. The core may have interfered with the transfer of metal from the center to outside of the coupon block or possibly produced a difference in the solidification rate at the center of the coupon.

It does seem, however, in view of the relatively heavy inter-dendritic areas, that the rate of solidification was only slightly slower than that at the center of the solid block and that the principal effect of the core was to prevent the free movement and agglomeration of inclusions at the center. The absence of large aggregates of inclusions resulted in considerable improvement in tensile ductility. Since the ductility

aggregates of inclusions resulted in considerable improvement in tensile ductility. Since the ductility was appreciably less than in the outer layers, it is indicated that the segregated inter-dendritic areas as well as the inclusions influence ductility adversely.

The results of this work show the preessity for se-

The results of this work show the necessity for selecting the types of test coupons with care so that the mechanical properties of the coupons will be representative of the castings. The rate of solidification, rate of cooling, and exact heat treatment given each section of the casting should be duplicated in the part of the coupon tested as representative of that section of the casting. It is necessary, in addition, to have sound, well-fed coupons and castings. In practice, the best procedure is to use a test coupon with its maximum thickness equivalent to the heaviest section of the casting and to heat treat this coupon together with the castings. Since the sand core or any such means of expediting the removal of the test piece from the coupon influences the mechanical properties of the test piece, its use is not permissible.

Acknowledgment

The authors are indebted to Dr. E. L. Reed of the Watertown Arsenal Laboratory for the metallographic and fracture photographs.

References

- 1. F. T. Sisco, The Alloys of Iron and Carbon, Vol. II, Properties, McGraw-Hill, 1937, pp. 43-51.
- 2. W. J. Merten, "High Temperature Treatments of Castings and Forgings-from Heavy Sections," *Trans.* ASST, vol. 13, pp. 1-21 (1928).
- 3. R. A. Bull, "Purchase Requirements for Steel Castings with Notes on Physical Properties in Test Bars and in Commercial Castings," *Proc.* ASTM, vol. 32, Part II, pp. 77-105 (1932).
- 4. Report of Steel Division Committee on Test Coupons. Transactions, A.F.A., vol. 45, pp. 97-105 (1937).
- 5. J. F. Harper and H. J. Stein, "The Microscope as a Control Instrument in Annealing Steel Castings," Transactions, A.F.A., vol. 32, pp. 679-689 (1924).
- 6. A. W. Lorens, "Problems and Practices in the Heat Treatment of Steel Castings," *Proc.* ASTM, vol. 32, 1932, Part II, pp.
- 7. C. E. Sims, "Heat Treatment for Grain Size in Cast Steel," Metal Progress, vol. 26, no. 3, 1934, pp. 22-27.
 - 8. J. H. Hall, Discussion of Ref. 6.
- 9. C. W. Briggs and R. A. Gezelius, "The Effect of Mass Upon the Mechanical Properties of Cast Steel," *Trans.*, A.S.M., vol. 26, 1938, pp. 367-385.
- 10. M. F. Hawkes, "Mechanical Properties of Cast Low Alloy Steels," Trans., A.S.M., vol. 39, 1947, pp. 1-41.
- 11. R. C. Wayne, H F. Bishop and H. F. Taylor, "Design of Test Coupons for Cast Steel," Transactions, A.F.A., vol. 52, pp. 1251-1265 (1944).
 - 12. C. H. Lorig, Discussion of Ref. 9.
- 13. J. H. Hollomon, L. D. Jaffe, et al, "The Effects of Microstructure on the Mechanical Properties of Steel," *Trans.*, ASM, vol. 38, pp. 807-844 (1947).
- 14. J. H. Hollomon, "Temper Brittleness," Trans., ASM, vol. 36, pp. 473-540 (1946).
- 15. S. A. Herres and A. F. Jones, "A Method for Evaluating Toughness of Steel," *Metal Progress*, pp. 462-469 (Sept. 1946). 16. H. Jolivet and G. Vidal, "The Value of the Impact Test
- 16. H. Jolivet and G. Vidal, "The Value of the Impact Test for Study of Temper Brittleness," Rev. Metallurgie, vol. 41, pp. 378-388 (1944).
- 17. Steel Founders' Society of America, "Mechanical Properties of Low Alloy Cast Steels," Steel Casting Report No. 1.
- 18. L. D. Jaffe and J. F. Wallace, discussion of Ref. 24.
 19. Steel Founders' Society of America, "Low Temperature
- Impact Properties of Cast Steels-2," Steel Casting Report No. 5.
 20. C. Wells and R. F. Mehl, "Transverse Mechanical Properties in Heat Treated Wrought Steel Products," 1948 Preprint
- of A.S.M.
 21. Climax Molybdenum Company, "Molybdenum," Hudson
- 22. H. F. Hawkes and B. F. Brown, "As-Cast Structures in Cast Steels," Iron Age, Oct. 14, 1948, vol. 162, no. 16, pp. 138-
- 23. C. E. Sims and F. B. Dahle, "Effect of Aluminum on the Properties of Medium-Carbon Cast Steels," Transactions, A.F.A., vol. 47, pp. 65-132 (1938).
- 24. S. A. Herres and C. H. Lorig, "Influence of Metallurgical Factors on Mechanical Properties of Steel," *Trans.* ASM, vol. 40, pp. 775-805 (1948).
- 25. H. F. Bishop and R. E. Fritz, "Segregation in Small Steel Castings," Transactions, A.F.A., vol. 55, pp. 412-426 (1947).

DISCUSSION

- Chairman: F. KIPER, Ohio Steel Foundry Co., Springfield, Ohio.
- Co-Chairman: G. W. JOHNSON, Vanadium Corp. of America, Chicago.
- G. A. LILLIEQVIST³: Is the looseness of the microstructure at the center of the section, as illustrated, due to microshrinkage of segregation?

MR. TAYLOR: I believe the only "looseness" was a coarsened structure at the center of the coupon without the core. We did not find any lack of density as the castings were well fed. Any difference was surely due to segregation. However, this difference was not reflected in the impact specimens.

A. B. STECK³: Did I understand the authors to say that they got a lower carbon content in the interior and a higher carbon content at the surface of the casting?

MR. TAYLOR: The carbon content was 0.275 per cent on the outside and 0.260 per cent in the center.

MR. LILLIEQUIST: What impact value did you obtain on the specimen at the center of the section?

Mr. TAYLOR: For the Cr-Ni-Mo steel at minimum 40 F, the outside coupon, we obtained an impact value of 52 ft-lb (V-notch Charpy) at roughly 253 Brinell hardness.

Mr. LILLIEQVIST: What was the sulphur and phosphorus content?

¹Director of Research, American Steel Foundries, East Chicago, Ind.

²Metallurgist, General Electric Co., Everett, Mass.

Mr. Taylor: Quite low. The sulphur varied between 0.008 and 0.021 per cent.

CHAIRMAN KIPER: I believe the authors stated that the test coupon should represent the casting, or words to that effect. I do not know how that can be done. There is something about the strength of an integrally-cast part which cannot be measured by separately cast coupons.

MR. TAYLOR: I think that is correct: I am committed to the fact that a test coupon measures only the quality of the metal in the ladle. It is impossible for a test coupon to reproduce faithfully the properties of the various parts of the casting.

If you are going to try to interpolate the value of a test coupon you must have in mind the fact that you cannot just take a piece and cast it separately, heat treat it, and expect it to represent what is inside the casting. The idea behind this was to find out what divergence could be expected as a result of segregation and cooling rate. I certainly could not endorse any expectations that a test coupon could reflect conditions in all parts of a casting.

THE RELATIVE EFFECTS OF CHROMIUM AND SILICON CONTENTS ON RATE OF ANNEAL OF BLACK - HEART MALLEABLE IRON

By J. E. Rehder*

PART I: FIRST STAGE ANNEALING

ABSTRACT

Determination of the quantitative effects of chromium content on the annealing rate of black-heart malleable iron in the first, or primary carbide decomposition stage, shows that chromium neutralizes the graphitizing effect of from 4 to 5 times its weight of silicon. The retarding effect of chromium content on annealing rate can be compensated for by increased annealing time, increased annealing temperature, increased silicon content, or a combination of all three since the effects are additive. Charts are presented from which the necessary changes in annealing practice can be determined.

Introduction

ALTHOUGH IT HAS BEEN KNOWN for many years that small amounts of chromium adversely affect the annealing rate of black-heart malleable iron, quantitative determination of the extent of the retarding effect of chromium on annealing rate has not to the writer's knowledge been done. With the extensive use of duplex melting for malleable iron and the accompanying use of larger quantities of scrap in the charge, and with the steady increase of alloy residuals in all steel scrap, the question of handling the chromium residual problem must be faced squarely. It is the object of the present paper to determine quantitatively the retarding effect of chromium in the annealing of malleable iron, as related to the silicon content, and to discuss remedial measures.

Due to experimental difficulties, determination of the effects in second-stage annealing is not complete, and the present paper will discuss in detail first-stage annealing only. The experimental work on secondstage annealing is continuing and the results will be published when complete.

Previous Work

As mentioned above, detailed measurements of the retarding effect of chromium content at different silicon levels has not to the writer's knowledge been re-

ported in the literature. Schwartz1 in a review of the effects of chromium on malleable iron reported in the literature, gives a table showing the amounts of graphite formed during first and second-stage annealing, as affected by chromium content. Since the graphitization was not carried to completion in all cases, and there was apparently only one base composition, the data are of limited use and simply corroborate the fact that chromium is a powerful retardant of graphitization. In the same paper the statement is made that the addition of 0.305 per cent of chromium to an essentially chromium-free iron increased the time necessary to form a given amount of graphite in first-stage annealing by about three times. Details of the composition of the base iron were not given.

In gray cast iron founding a rule of thumb frequently used in balancing silicon and chromium contents of alloy cast irons with respect to chilling tendency, is that they are about equal. This, however, has no necessary relationship with graphitizing rate in the solid state.

Experimental Method

Melting—The plan of experiment was to make up a series of white irons of several silicon contents, with at least three chromium contents at each silicon level, and determine the times necessary for annealing of each iron. It was essential that other conditions and chemical constituents be maintained constant throughout.

A 300-lb heat was made up of pig iron and steel scrap, with a little sulphur, and melted in the induction furnace. The melt was heated to 1580 C (2880 F) and poured into six 50-lb ingots. Before each ingot was poured, 50 per cent ferrosilicon was added to the furnace to increase the silicon content, waiting in each case for 3 min after all silicon was in solution, before pouring the ingot. The ingots were cast in metal molds, the chilling action ensuring that all ingots solidified completely white and free of primary graphite. This provided six ingots of the same composition except with respect to silicon content.

co

ti he ac

fu

th

to

ta

ac

In

at

fe

ro

be

es

or

pr

er

in

da

eff

of

sin

tiv

fo

no

by

Str

ab

eff

ne

sq

sil

Published by permission of the Director-General of Scientific Services, Department of Mines and Technical Surveys, Ottawa.

[•] Foundry Engineer, Mines Branch, Department of Mines and Technical Surveys, Ottawa, Ontario, Canada.

e

11

1

í

Each ingot was then remelted in a 50-lb induction furnace, heated to 1580 C (2880 F), and successive 10-lb taps taken in a preheated hand-ladle. Small additions of ferrochromium were made to the furnace before each tap, the metal being held at temperature for 3 to 4 min after the last chromium had dissolved. Each 10-lb tap was poured into a 5-in. by 10-in. by 1/4-in. thick plate in a green sand mold.

By the above described procedure 24 plates were obtained, six levels of silicon with four chromium contents at each silicon level, with conditions essentially the same for each plate. The reason for holding the molten iron at temperature after each addition of ferrochromium was to obliterate the inoculating effect of the added chromium. It has been the writer's consistent experience that both ferrochromium and ferromanganese, when added to a bath of molten white iron, have a strong inoculating or mottling tendency at first, in spite of their powerful carbide stabilizing effects during subsequent annealing.

In normal malleable iron foundry melting practice this mottling effect of an addition of ferromanganese to the bath is well recognized and iron should not be tapped from the furnace for at least 10 min after an addition to give the mottling effect time to disappear. In gray iron foundry practice it is well known in the automotive field that a small addition of crushed ferrochromium will frequently decrease a chill that resists other inoculants.

These inoculating effects of additions of solid ferrochromium or ferromanganese to a metal bath may be related to a deoxidizing effect, since chromium especially is nearly as powerful a deoxidizer as silicon, or to the inoculating effect of adding any cold solid granulated material to a bath of cast iron. There is a present tendency to regard the effects of adding powerful reducing agents or deoxidizers, such as aluminum, titanium, or silicon, as entirely due to deoxidation. While there is undoubtedly some such effect, it is interesting to note that the addition of molten ferrosilicon to cast iron has little or no inoculating effect.

However, the point is that by the above described procedure a series of plates were obtained of different silicon and chromium contents which are believed to be essentially uniform in other respects. The use of induction melting is not considered of importance, since the results obtained on annealing are of relative interest among themselves, and there is no reason for believing that the relative results obtained will not be applicable directly to malleable irons melted by other means.

Annealing—From each of the plates obtained, a strip 3/4-in. wide was cut off all four edges with an abrasive cut-off wheel and discarded, to avoid edge effects. Dilatometer samples, used for first-stage annealing time determinations, were cut to be 1/4 in. square and 3 in. long.

The dilatometer used for first-stage annealing determinations was a simple silica tube with a hollow silica push-rod operating a dial gage. Scaling was eliminated by passing pure, dry nitrogen through the dilatometer tube via the hollow push-rod. The sample was inserted with the furnace and dilatometer standing at 954 C (1750 F) and all first-stage annealing was done at that constant temperature. Control of temperature was \pm 2 F. All samples were checked metallographically after removal from the dilatometer.

Nodule counts were made on all annealed samples at 50 diameters, and reported as number per square millimeter of the actual sample.

Results

In Table 1 will be found the chemical analyses of the plates used in the present investigation. Complete analyses for all elements were not performed since this was made unnecessary by the melting procedure used. The two lowest silicon series, being at 0.50 and 0.75 per cent silicon respectively, were not used since annealing times would have been prohibitively long. For the same reason the highest chromium level (0.24 per cent) was not used.

TABLE 1-CHEMICAL COMPOSITION OF SAMPLES FOR ANNEALING

Plate	C, %	Si, %	Mn, %	S, %	P. %	Cr, %
3A	2.54	0.99	0.25	0.076	0.128	0.047
3B		0.98				0.089
3C		0.98				0.158
4A	2.53	1.17				0.054
4B		1.16				0.086
4C		1.17				0.158
5A	2.50	1.30				0.047
5B		1.33				0.089
5C		1.34				0.158
6A	2.51	1.50	0.28	0.072	0.121	0.047
6B		1.52				0.089
6C -		1.50				0.165

Nodule counts on annealed specimens, each the average of five determinations, are given in Table 2.

TABLE 2-Nodule Counts

Sample	No. Nodules p Sq Mm Samp	
3A	11	
3C	12	
4A	29	
4B	16	
4C	20	
5Λ	58	
5 B	35	
5C	26	
6A	29	
6 B	32	
6C	29	

The shape of the temper carbon nodules was affected to some extent by the chromium content, especially at lower silicon content, where a filling-in and rounding effect was noticed at the highest chromium content. In Fig. 1 to 4 inclusive are shown examples from the extremes of silicon and chromium content used. It was noted during metallographic examina-

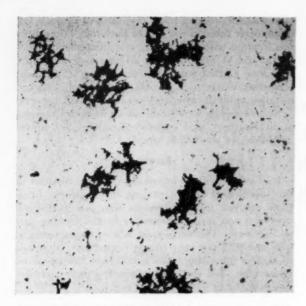


Fig. 1-Sample 3A. 0.99 Si, 0.047 Cr. Unetched. ×100.

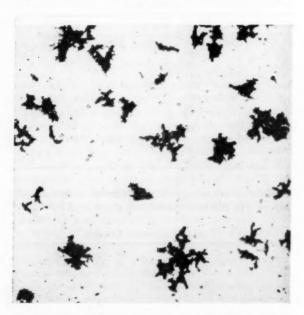
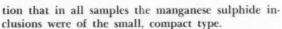


Fig. 3-Sample 6A. 1.50 Si, 0.047 Cr. Unetched. ×100.



The times necessary for completion of first-stage annealing at constant temperature are given in Table 3. The precision of these values is approximately 10 per cent. The results are shown graphically in Fig. 5.

Discussion of Results

From consideration of the chemical analyses shown in Table 1 and the manner in which the melts were prepared, it is believed that a series of plates were obtained which were essentially uniform except with respect to chromium and silicon contents, and that

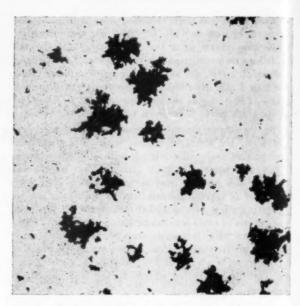


Fig. 2-Sample 3C. 0.98 Si, 0.158 Cr. Unetched. ×100.

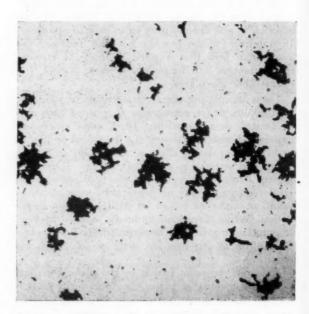


Fig. 4-Sample 6C. 1.50 Si, 0.165 Cr. Unetched. ×100.

TABLE 3-ANNEALING TIMES

aı

re

st

iu ar

Sample	First Stage, Hr at 1750 F		
3A	9.0		
3C	28.0		
4A	6.5		
4B	8.5		
4C	16.0		
5A	4.25		
5B	6.25		
5C	10.0		
6A	2.75		
6B	4.0		
6C	7.0		

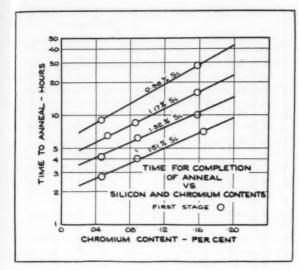


Fig. 5-Annealing time vs Cr and Si contents.

the results of annealing time determinations performed represented real changes due to the effects of variation of chromium and silicon contents,

Nodule Counts—The results of nodule counts given in Table 2 are indefinite, except that broadly speaking the lower silicon contents and higher chromium contents gave lower nodule counts. Certainly the retarding effect on annealing rate of increasing chromium content and decreasing silicon content cannot be accounted for quantitatively by changes in the number of temper carbon nodules formed per unit volume. The shapes of the nodules were also not altered seriously by changes in chromium content, except at low levels of silicon.

Annealing Times—It will be noted that the results in Table 3, when plotted on semi-logarithmic paper, as in Fig. 5, show straight lines, within experimental error. It is well to remember that when a logarithmic plot is used, and straight lines are found to represent the results, the data may be interpreted as percentage changes which are true no matter what the actual or initial time or times may be. For this reason such logarthmic paper is frequently called "percentage paper." This fact is very useful, and makes possible general treatment that is applicable to laboratory as well as to foundry or annealing room data.

Chromium versus Time—It is evident from Table 3 and Fig. 5 that as silicon content decreases, the lines relating chromium content to annealing time are of steeper slope, meaning that a given increase in chromium content causes a greater percentage increase in annealing time as silicon content is lower. If these slopes are measured for each silicon level and the slopes plotted against silicon content, an average line may be drawn that gives the slope at any desired silicon level. These, converted to percentages, are given in Table 4, and in graphical form in Fig. 6. It should be remembered that although the Table gives

results to the first place of decimals, the accuracy is not better than about one in ten, which is, however, sufficient for commercial use.

TABLE 4

Silicon Content, Per Cent	Percentage Increase in Annealing Time made necessary by each Increase of 0.01 per cent Chromium Content
0.90	10.9
1.00	10.4
1.10	9.8
1.20	9.2
1.30	8.7
1.40	8.1
1.50	7.6

For example, if a white iron of about 2.5 per cent carbon content, 1.2 per cent silicon content, and 0.04 per cent chromium content was being satisfactorily annealed in, say, 20 hr at 1700 F first-stage annealing; an increase in chromium content of 0.03 per cent (making a total of 0.07 per cent chromium) would necessitate the first-stage annealing time being extended from 20 hr to a new time of 20 plus 27 per cent (from Fig. 6) or 251/2 hr. Similarly, an increase in chromium content of 0.08 per cent would require an increase of 74 per cent in annealing time (at 1.2 per cent silicon), or a new first-stage annealing time of 35 hr. Commercially the length of first-stage anneal is usually longer than necessary, to take care of normal variations in composition and plant practice, and a small increase in chromium content may cause little trouble. If, however, the safety factor is to be maintained, as it should, the annealing time must be extended as chromium content rises.

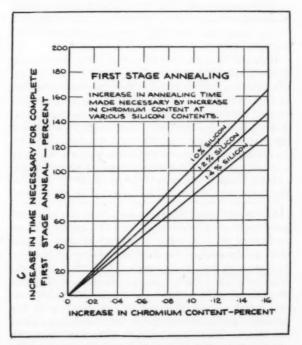


Fig. 6—Increase in first-stage annealing time made necessary by each increase in Cr content of 0.01 per cent.

Chromium versus Temperature—It is well known that increasing the annealing temperature decreases the annealing time, and the relationship has been shown by Rehder² to be essentially constant for various types of iron. It is apparent that if an increase in chromium content has increased the necessary annealing time, the time may be decreased again by raising the temperature of anneal. By comparing the data in Fig. 6 and the temperature—annealing rate data published by Rehder,² it is possible to determine by what amount first-stage annealing temperature must be increased to compensate for a given increase in chromium content. Results of such comparison are given in graphical form in Fig. 7.

It will be noted that the increases in annealing temperature in this case are degrees F, not percentages. Using again our example of a 2.5 per cent carbon, 1.2 per cent silicon, 0.04 per cent chromium white iron which annealed satisfactorily in the first stage in 20 hr at 1700 F; an increase in chromium content of 0.03 per cent (to 0.07 per cent total) could be compensated for by increasing the annealing temperature (see Fig. 7) by about 30 F, making a new first-stage annealing temperature of 1730 F. Similarly, an increase in chromium content of 0.08 per cent (at 1.2 per cent silicon) would require an increase in annealing temperature of 75 F, or a new first-stage annealing temperature of 1775 F if the annealing time were to remain constant at 20 hr.

However, in the latter case it would be unwise to use a 1775 F annealing temperature due to possible trouble from sagging and scaling of castings, 1750 F being considered the safe maximum. In such case, part of the effect of the increased chromium content could be compensated for by increasing the annealing temperature from 1700 to 1750 F, and the remainder by increasing the annealing time, since the effects are additive.

It should be noted that there is a useful coincidence in the graphs of Fig. 6 and 7, in that a given increase in chromium content requires about the same amount, in degrees F increase in annealing temperature or percentage increase in annealing time, for compensation. Taking into consideration commercial tolerances and the accuracy of the present work, the rule could be formulated that for compensation of each 0.01 percent increase in chromium content, either the annealing temperature should be increased by 10 F or the annealing time should be increased by 10 per cent.

Chromium versus Silicon—Since silicon is a powerful accelerator of graphitization, and chromium is a retardant of graphitization, it should be, and is, possible to compensate for increased chromium content by increasing the silicon content. This refers of course to silicon in the charge, and not to the special effects of a late or ladle addition of silicon, which latter in all probability would cause serious mottling of the castings. This relationship has been worked out from the data in Table 3 and Fig. 5, with data previously published by Rehder² and may be summarized by the statement that for a carbon content of approximately 2.5 per cent, and silicon contents between 1.0

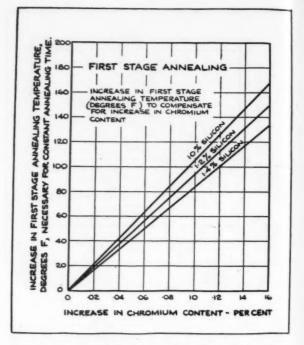


Fig. 7—Increase in first-stage annealing temperature necessary to compensate for an increase in Cr content.

t

te

h

CE

01

an

me

tin

and 1.6 per cent, each 0.01 per cent increase in chromium content requires an increase in silicon content for equal annealing time of from 0.05 per cent silicon at 1.5 silicon to 0.04 per cent silicon at 1.0 silicon.

It should be noted that as carbon content increases, the effectiveness of silicon as an accelerator of graphitization apparently decreases, and so at higher carbon contents more than the above noted amounts of silicon will be necessary to compensate for chromium content.

Using again our example of a 2.5 per cent carbon, 1.2 per cent silicon, 0.04 per cent chromium white iron that anneals satisfactorily in the first stage in 20 hr at 1700 F; an increase in chromium content of 0.03 per cent can be compensated for by increasing the silicon content by about 0.05 times 3, or 0.15 per cent, to a new level of 1.35 per cent. It will be evident immediately that the use of increased silicon content to compensate for increased chromium content is of limited commercial value, since mottle in heavier sections will probably be produced when an increase of more than about 0.02 per cent in chromium content is compensated for by increase in silicon content.

It should be noted well that no determinations of the effect of chromium content on mottling tendency were made in the present study, and to the writer's knowledge little is known on the subject. Until the anti-mottling, or whitening, power of chromium content (if any) is well established, it is considered safer to assume that it has no effect on mottle. If the mottling tendency of higher silicon irons could be decreased by other methods or additions, such as bismuth,³ the use of silicon as a compensator for chromium content in malleable iron annealing would have greater practicability.

Compensation for Increased Chromium Content-By a combination of the methods described above, compensation for the effect of increased chromium content can be made appreciable. Taking again our example of a 2.5 per cent carbon, 1.2 per cent silicon, 0.04 per cent chromium iron that anneals satisfactorily in the first stage in 20 hr at 1700 F; and assuming that the maximum safe annealing temperature is 1750 F and the maximum safe increase in silicon content is 0.10 per cent, what would be the new annealing conditions for an increase of 0.110 per cent chromium content?

From Fig. 7, since the maximum permissible increase in annealing temperature is 50 F, it is seen that 0.055 per cent increase in chromium can be compensated for in this way. Again, the permissible in-0.10

crease in silicon content represents or 0.02 per cent chromium that can be compensated for in this way. Of the total increase in chromium content of 0.110 per cent, 0.055 plus 0.020 or 0.075 per cent can be compensated for by increases in annealing temperature and silicon content. This leaves 0.110 minus 0.075 or 0.035 per cent to be compensated for by increased annealing time, and from Fig. 6 this is found to be about 30 per cent, making a new annealing time of 20 plus 6 or 26 hr. The new conditions for an increase in chromium content of 0.110 per cent are then a 26-hr anneal at 1750 F, with 1.30 per cent silicon content.

Second-Stage Annealing-Some preliminary results have been obtained for second-stage isothermal annealing times as affected by chromium content. These indicate that at higher silicon levels, about 1.5 per cent, each 0.01 per cent increase of chromium content increases second-stage annealing time by about 12 per cent, and at lower silicon levels the retarding effect of chromium increases to still higher levels. In other words, chromium has a greater retarding effect on second-stage annealing, as a percentage, than it does on first-stage annealing. As mentioned above, work is continuing on this part of the investigation.

Conclusions

From the data and discussion presented above, certain conclusions may be drawn. It should be noted that the reference carbon content is approximately 2.5

1. Chromium retards the graphitization rate of white cast iron, each 0.01 per cent increase in chromium content increasing the time necessary for complete first-stage anneal by from 11 per cent at 0.9 per cent silicon content, to 8 per cent at 1.5 per cent silicon content.

2. The increase in silicon content necessary to compensate for a given increase in chromium content, with respect to first-stage annealing time, is from 4 to 5 times the increase in chromium content.

3. The retarding effect on annealing rate of an increase in chromium content can be compensated for by increasing the annealing time, or increasing the annealing temperature, or increasing the silicon content. A combination of these effects may be used, since they are additive, and limitations due to mottling from too high silicon content or warping due to too high annealing temperature thereby avoided.

4. As a practical rule, assuming approximately 2.5 per cent carbon content and 1.1 to 1.5 per cent silicon content, the effect on first-stage annealing rate of each increase of 0.01 per cent chromium content can be compensated for by increasing the annealing time by 10 per cent, or increasing the annealing temperature by 10 F, or by increasing the silicon content by 0.05 per cent.

5. The effect of chromium content, in the range 0.04 to 0.16 per cent, on the shape and number of temper carbon nodules is not large, and is not believed sufficient to account for differences in annealing rates solely on the basis of migration distance of carbon atoms.

6. The retarding effect of a given amount of chromium on the rate of second-stage annealing is larger than is the effect in the first stage of anneal, and probably increases more rapidly as silicon content decreases.

Bibliography

- 1. H. A. Schwartz, "Chemical Composition of Malleable Iron," TRANSACTIONS, A.F.S., vol. 54, pp. 101-115 (1946).
 2. J. E. Rehder, "Effects of Temperature and Silicon Con-
- tent on First-Stage Annealing of Black-Heart Malleable Iron,"
- Transactions, A.F.S., vol. 57, pp. 173-180 (1949).
 3. A. L. Boegehold, U. S. Patent No. 2,370,225 (1945).

PART II - SECOND STAGE ANNEALING

ABSTRACT

Using the same samples of white iron with different silicon and chromium contents as used for determination of first-stage annealing times in Part I of this paper, determinations were made of the isothermal second-stage annealing times, and the maximum rates of cooling through the critical temperature range for complete elimination of pearlite. It is shown that chromium neutralizes the graphitizing power of from 7 to 8 times its weight of silicon in as-cast samples for second-stage annealing. Compensation for the stabilizing effect of chromium

in second-stage annealing consists of using slower cooling rates through the critical temperature range, and, if necessary, then holding at subcritical temperature until pearlite is elimin-

This paper is a direct continuation of the work on first-stage annealing reported in Part 1 using the same samples. A description of the preparation of the samples will not be repeated here, since complete details have been given.

Experimental Method

One-quarter inch thick as cast plates, at four silicon contents and three different chromium contents at each silicon content, formed the base irons as described previously. The dilatometer used then for first-stage annealing was used in the present work to obtain the time necessary at constant temperature of 716 C (1320 F) to completely decompose all pearlitic cementite, after a first-stage anneal corresponding to that previously found necessary.

However, it soon became apparent that the dilatometric method would be suitable for only a few of the more easily annealed compositions, since the curve of dilation versus time became a very shallow asymptote that was difficult to assess accurately. The results obtained were checked metallographically, but even this method became tedious for the samples with longer

annealing times.

Although isothermal data are more easily interpreted quantitatively, in commercial malleable annealing practice second-stage annealing (elimination of pearlite) is nearly always carried out by slow cooling through the critical temperature range. The experimental procedure was therefore changed to that of determining the most rapid rate of cooling through the critical temperature range which would give complete elimination of pearlite. To do this, 1 by ½-in. sections of the ¼-in, thick plates were packed in cast iron chips in short sections of ½-in, steel pipe, and placed in a furnace standing at 954 C (1750 F).

After holding at this temperature for a sufficient time to complete first-stage annealing, the furnace and contents were cooled to 788 C (1450 F) in about 20 min, and a rate-of-cool controller switched on. From 788 C (1450 F) to 716 (1320 F) the cooling was at a controlled rate, after which the sample was removed,

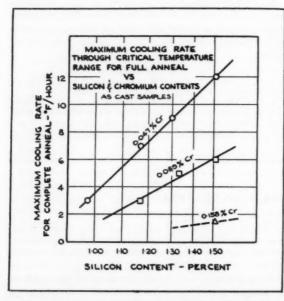


Fig. 8—Maximum Cooling Rates through Critical Temperature Range for Complete Elimination of Pearlite in As-Cast Samples.

cooled, sectioned, polished and etched, and examined under the microscope for traces of pearlite. If pearlite was present, the run was repeated with a fresh sample at slower rates until a sample was free of pearlite (completely ferritic in matrix). This procedure was repeated for all samples.

It was found that for the lower silicon content, higher chromium content irons, cooling rates through the critical temperature range of 1 F per hr or less were required for complete elimination of pearlite. Considering the relatively thin as-cast section, this represented cooling rates far below any commercially practicable, and so experimental cooling rates below about 2 F per hr were not used.

Results

The results of the isothermal annealing work are given in Table 5. Where not specified in this paper, results are accurate to 10 per cent. In Table 1 the longer times are less certain in accuracy.

TABLE 5—TIME NECESSARY AT 1320 F TO ELIMINATE ALL PEARLITE. AS CAST SAMPLES

iı

ta

st

te

te

W

sa

ar da te

slo

co

us

iu

CO

wi

ch

in

wi

Sample	Time, hr
6A	. 18
6B	37
6C	65-70
5A	30
5 B	60-65
4A	37
4B	75-85

The maximum permissible cooling rates through the critical temperature range for elimination of all pearlite are given for as cast samples in Table 6. In each case, use of a cooling rate 10 per cent or less more rapid, will result in the presence of traces of pearlite. The results are given in graphical form in Fig. 8, and it will be noted that the silicon content is plotted on a logarithmic base.

TABLE 6—MAXIMUM COOLING RATES THROUGH CRITIC-AL TEMPERATURE RANGE FOR COMPLETE ELIMINATION OF PEARLITE. As-CAST SAMPLES

Sample	Cooling Rate Deg F per Hr
6A	12
6B	6
6C	13/4
5A	9
5B	5
4A	7
4B	3
3A	3

Since it is known that when isothermal second-stage annealing times are plotted versus silicon content on log-log paper a straight line results, and since the only plot that gives straight lines for cooling rate versus silicon content is on semi-logarithmic paper, it would appear that a plot of maximum cooling rate versus isothermal annealing times at a given temperature should also be a straight line on semi-logarithmic

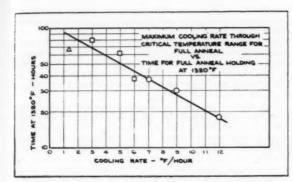


Fig. 9—Isothermal Annealing Times at 1320 F versus Maximum Cooling Rate through Critical Temperature Range.

paper. This was checked with the data available, as shown in Fig. 9.

Discussion of Results

Isothermal Annealing—Comparison of these results with those given previously for first-stage annealing indicates that chromium is quantitatively more retarding in its effect on annealing rate in the second stage than in the first stage. Since the data are not extensive and isothermal second-stage annealing is not of particular importance commercially, these results will not be discussed further directly.

Slow-Cool Annealing—It is evident from the data given in Table 6 and Fig. 8, that chromium neutralizes the graphitizing effect of silicon in marked degree. For example if a cooling rate through the critical temperature range of 5½ F per hr were satisfactory for a silicon content of 1.10 per cent and a chromium content of 0.047 per cent, then if the chromium content increased to 0.089 per cent the silicon content would have to be increased to 1.43 per cent for the same cooling rate to effect a full anneal. That is, an increase in chromium content has neutralized the graphitizing effect of 7 to 8 times its weight of silicon. This is evidently a more powerful effect than obtaining in first-stage annealing, where the ratio is 4 or 5 to 1.

Since it is impractical commercially to increase silicon content sufficiently to neutralize the effects of any serious increase in chromium content, due to the danger of mottle, the only recourse if chromium content increases is to decrease the cooling rate to the slowest practicable and if full anneal is still not being obtained, to hold in the range of 1250 to 1330 F, after slow cooling, until full anneal is achieved. Normally most malleable foundries operate under conditions such that there is some leeway or safety factor in the combination of iron metallurgy and annealing cycle used, so that the effects of small increases in chromium content are automatically absorbed. If chromium content is rising to a dangerous level, the first sign will be a deepening of pearlitic rim, followed, if chromium continues to increase, by traces of pearlite in the center section which rapidly increase in amount with chromium content.

It is of course hardly necessary to mention that the actual values reported in this paper for suitable cooling rates for full anneal cannot be directly used in commercial annealing operations, as it is well known that the values are affected by melt history, section size, and other variables. However, the relationships between values reported are expected to be true for other conditions.

Cooling-Rate vs Isothermal Time—The graph shown in Fig. 9 indicates that the relationship is in fact semi-logarithmic, as was expected from the other relationships mentioned. It is interesting to note that this relationship suggests different mechanisms of graphitization in slow-cooling and in isothermal holding, since at high permissible cooling rates, total annealing time is shorter than isothermally, whereas for very slow permissible cooling rates, slow-cooling takes longer than the isothermal process. This is in agreement with metallographic observation, the isothermal subcritical anneal consisting entirely of decomposition of pearlite, while during slow cooling the austenite appears to decompose directly to ferrite and graphite without intermediate visible formation of pearlite.

Conclusions

From the data presented above, the following conclusions are drawn:

1. The limited amount of data on annealing time at constant subcritical temperature indicates that a given amount of chromium retards second-stage annealing more severely than it does first-stage annealing.

2. Whereas the retarding effect of chromium in first-stage annealing neutralizes the graphitizing effect of from 4 to 5 times its weight of silicon, in second-stage annealing the ratio is one to 7 or 8.

3. Since the effect of chromium as a retardant of anneal is larger in the second stage of anneal than in the first stage, the compensation of the effect of chromium content by increasing the silicon content must follow the 7 or 8 to 1 ratio, and under these conditions it is not of commercial importance to attempt to compensate for chromium by increasing the silicon content, due to danger of mottling.

4. In commercial annealing of normal as-cast malleable (white) iron, the only recourse if chromium content increases to a level where pearlite is retained in the final product, is to use a slower cooling rate through the critical temperature range, and if the slowest obtainable rate is not sufficient, to subsequently hold the charge in the temperature range of 1250 to 1330 F for a sufficient length of time to eliminate the remaining pearlite.

5. It is evident that compensation for the carbide stabilizing effect of chromium must be made separately for first and second stage annealing, following the rules developed in Parts I and II of this paper.

DISCUSSION

Chairman: C. F. LAUENSTEIN, Link-Belt Co., Indianapolis, Ind. Co-Chairman: F. Coghlin, Jr., Albion Malleable Iron Co., Albion, Mich.

J. H. Lansing: 1 Mr. Rehder's paper is most interesting. It would be well if he would point out what differences he found in ultimate strength, yield point, and elongation between the material with no chromium and the same material with 0.06, 0.07 or 0.08 chromium.

Mr. REHDER: Originally it was planned to do this work strictly on the metallurgy involved and then on the basis of results from a commercial malleable iron foundry, where bars of different chromium contents would be made up, and after suitable annealing to determine mechanical properties. However due to time limitations the work on mechanical properties was not done.

R. SCHNEJDEWIND: 2 Mr. Rehder presented quantitative and systematic data most valuable to malleable foundrymen, on the effect of chromium on the annealing rate of white irons. As in all of Mr. Rehder's papers, complete data are given so that his results can be intelligently studied. He has given the industry a clear and quantitative picture of the great retarding influence of chromium on the annealing of white cast iron. There can be no question as to his technique or results. I want to compliment Mr. Rehder on giving us data which we can add to the original data and study further.

I have a few questions regarding things which the paper does not intend to deal with. For example, if the data in Tables 1 and 3 are plotted with the log of time as ordinate and the log of silicon as abscissa, we would get three straight and parallel lines as you would expect. These lines that are obtained represent the relationship of silicon on time of first-stage annealing at the three chromium levels. The slope of these lines is about -3.1 rather than around -2.0 as Mr. Rehder found in his paper in 1949. (See A. F. S. Transactions, vol. 57, 1949, pp. 173 and 549). It is possible in view of the greater accuracy and control in the present work that the value of -3.1 is a better figure than -- 2.0. Does the author have any opinions on this subject? Also, why are there no zero chromium data? I am sorry Mr. Rehder did not run a guinea pig heat with zero chromium.

What are the author's limits of accuracy indetermining annealing time so that further study can be made of the data? It is plus or minus 1/2 hr, 1 hr, or plus or minus some definite per cent?

With regard to the second stage part of the paper, Mr. Rehder has given us cooling rates through a critical range. This range represented 130 degrees, I understand. With such slow cooling rates, this would entail 60 hr, more or less, in the second stage of annealing time with the higher chromiums. However, if we just plot the time he got at constant temperature against the time it took to go through 130 degrees, there is a straight line relationship. The conclusion is that in view of the first stage retardation and the enormous second stage retardation we do not want chromium if we can help it.

MR. REHDER: Thank you very much, Prof. Schneidewind. I enjoy and value your discussions, since you take the trouble to read a paper carefully, check results, and re-plot or re-interpret data if there appears to be an advantage in clarity or understanding. More of this should be done, as it is a valuable addition to the usefulness of a paper.

You mentioned the fact that if the results are plotted on a

log-log basis with time as ordinate and silicon content as abscissa, the slopes of the parallel straight lines obtained are about minus 3.1, as opposed to the slope of about minus 2.0 which would be expected from previously published work. I was aware of this apparent disagreement, but said nothing of it in the paper since at the moment I have no explanation. The previous work was the average of several other investigators' results along with my own data, and the question of whether the more recent, possibly more accurate work is more nearly correct than the older, averaged data is an open one. With respect to present accuracy, it is mentioned in the paper that the results are considered correct within 10 per cent.

MILTON TILLEY: 8 Up to the present time we have not been bothered with the first-stage anneal because we are well protected except in the case of high chromium like 0.15, 0.20 or 0.25 per cent which occurs occasionally. But in the near future we will be confronted with using that data, which I am pleased to obtain at this time.

I am glad to see that Mr. Rehder's information checks our 25 to 30 years experience with chromium contents on the second stage anneal. We anneal with about 6 deg per hr cooling through the critical temperature range. We are safe up to 0.05 chromium as Mr. Rehder's figures indicate. Castings with higher chromium contents must be put re-annealed in a batch-type furnace and given a longer cooling rate through the critical range.

MR. REHDER: I would like to comment on two of the points which you mentioned. The first concerns also Dr. Schneidewind, and is the question of zero, or very low, chromium contents. Although the inclusion of data on an iron with very low chromium content would have been of theoretical interest, it was considered more important to work with raw materials of normal commercial use, such as steel scrap and pig iron, and the chromium contents, used in the paper do cover the range likely to be of practical interest.

The other point of Mr. Tilley's is the question of re-annealing. The effects of successive re-annealing are additive for the first stage, but not for the second stage. If for example some primary or massive carbide is still present after a normal anneal, due to the presence of chromium or insufficient silicon or other cause, this carbide may be eliminated by a second (or in rare cases, third) anneal. If, however, pearlite remains after a normal anneal, it cannot usually be removed by a repetition of the same annealing cycle. The re-anneal in this case must be at a slower rate through the critical temperature range, or be followed by a sub-critical holding period, or both.

W. D. McMillan: 6 Mr. Rehder is to be complimented for another good paper for the benefit of the malleable industry.

There is one thought that comes out of this work which is more far reaching than the mere understanding of the effect of chromium. It is simply that there is no reason to give up, or give in to the little discrepancies that creep into any process that involves the human element.

He has pointed out that a little more chromium than we expected may be compensated for by either time, temperature or silicon, any or all three.

In order to apply a special treatment, it is necessary to know which castings are involved. To know this will require on the spot advance information, a system, a follow up and follow through. Control of annealability is pretty much control of composition. It is a matter of foresight rather than hindsight. There is little satisfaction in learning that this casting missed the anneal because of 7 or 8 points of chromium. Control is a matter of live data regarding the melt while it is melt not only with respect to chromium but to manganese, sulphur, silicon, temperature, etc.

If corrective measures are not made before the metal is poured, there will be the job of segregating the castings, identification and adjustment of cycles. With adequate data you will not be working with unknowns and can establish an orderly procedure to bring the castings through.

The information Mr. Rehder has given is factual and specific. He has provided another tool which is of value only when put to use.

¹ Technical and Research Director, Malleable Founders' Society, Cleveland.

² Prof. of Met. Engr., University of Michigan, Ann Arbor, Mich.

³ Metallurgist, National Malleable & Steel Castings Co., Cleveland.

McCormick Works, International Harvester Co., Chicago.

DETERMINATION OF METAL PENETRATION IN SAND MOLDS*

By

Harold J. Gonya** and David C. Ekey***

Introduction

METAL PENETRATION INTO SAND MOLDS and cores is an old foundry problem. "Metal penetration is a casting surface defect wnich looks as though the metal has filled the voids between the sand grains without displacing them." Another definition is, "Metal penetration and sand adherence results when the metal flows through the interstices between the grains and into the sand to some depth thus producing an adhering mass of sand and metal in the casting."2 There have been many attempts to isolate the cause or causes of metal penetration, but past experimentation has not brought to light any definite cause of metal penetration. One theory postulated by J. B. Caine³ was that voids present in the sand as rammed was a major cause of metal penetration. Another theory, supported by Gertsman and Murton,4 is that ferrostatic pressure and gassy metal are major causes of metal penetration in sands used when casting steel.

This problem of penetration is important, since it affects the machineability of castings. It is difficult to machine a penetrated surface and few cutting tools will hold up under such conditions. Many times the cleaning operation necessary to remove the penetrated sand, is more expensive than the original cost of manufacturing the casting, and can result in scrapping the product.

The object of this experiment was to attempt to determine what variables might be significant in their relation to penetration. It was not hoped to determine the full effect of any particular variable in metal penetration. With this thought in mind it was decided to use a statistical method of experimentation to determine the significance of the variables studied.

The statistical method chosen enables one to study several variables simultaneously with a minimum number of experiments. This was especially desirable in this instance in view of the magnitude of the problem and the limitations of time and funds. Knowing whether or not any particular variable affected penetration would be helpful in setting up a classical type of experiment to further determine the cause of metal penetration. In the classical type of experiment an attempt is made to hold all variables constant except the one being studied.

The number of variables to be studied was limited to five because of the nature of the statistics which involved a four by four Greco-Latin Square as shown

TABLE 1
GRECO-LATIN SQUARE

	1	П	ш	IV
1		5	9	13
	a 1 Ø	ь 2 В	c 3 8	d 4 8
2		6	10	14
	6 8	d 3 ck	a 2 8	b 1
3		7	n	15
	ა გ	a 4 8	d 1 8	2 Q
4		8	12	16
	d 2 8	c 1 8	b 4	a 3 8

Code For Symbols Used

Sample Number

A, B, C, D — Time Variable

A, B, C, D — Ramming Variable

1, 2, 3, 4 — Static Pressure Variable

4, 6, 5, 6 — Moisture Variable

1, II, III, IV — Sand Distribution Variable

N

ect een ro-25 we sed

nd gh

ımı nd

nts

m-

he

elv

ıg.

rst

rv

ne

er

re

ol-

or

of

at

or

184

he

oi it.

ed

ly

n.

d.

n

re

^{*}Taken from theses presented to the faculty of Ohio State University in partial fulfilment of the requirements for the Master of Science Degree.

^{**}Project Engineer, Harris-Seybold Co., Dayton, Ohio.

^{***}Graduate Student, Ohio State University.

in Table 1. This square required five variables, each variable having four levels of variation.

The variables used were sand grain distribution, per cent moisture, static metal pressure or head, setting time of the test specimens in the mold, and number of compacting rams. These variables, shown in Table 2, were chosen after considerable investigation as to the possible causes of penetration and consultation with Dr. D. C. Williams of the Industrial Engineering Department.

TABLE 2

VARIABLES USED - Sand Distribution

Sand	_	Mesh Size										
	20	30	40	50	70	100	140	200	270	Pan		
G-D				20	20	20	20	20				
A-7	0.3	4.1	32.7	55.5	7.1	-		1.2		-		
A-85	0.15	0.90	1.3	7.0	8.0	10.3	15.1	27.8	19.2	10.3		
S-B-S*	0.7	40.0	54.0	4.3								

Moisture, %	Rams		Time in Mold Before Pouring, min				Static Pressure, psi		
α — 0	A - 2		-	5	1	_	4.48		
B - 3	B - 9	ь	_	30	2	_	3.86		
8 - 5	C -15	c.	_	120	3	-	2.87		
6-10	D -50	d	_	300	4	_	1.94		

Sand Black Sand

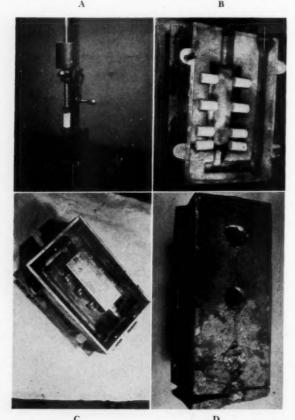


Fig. 1-(A) 11/8-in. Rammer; (B) Mold with Cores in place; (C) Assembled Flask and Pattern; (D) Brass Casting.

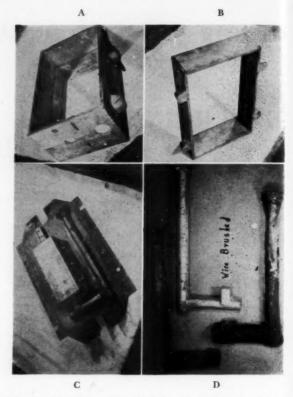


Fig. 2-(A) Flask; (B) Flask Pattern; (C) Aluminum Pattern; (D) Sprues, Aluminum and Brass.

Brass and aluminum alloys were chosen for study of the problem of penetration with the above-mentioned variables. These two alloys were selected because of the limitations of melting facilities in the foundry and also because they are two alloys with which metal penetration is frequently associated. For the test on aluminum, and aluminum alloy of 95 Al, 5 Si was used, and for the test on brass, an alloy of 85-5-5-5 was used. There were no facilities available for melting steel, and the maximum metal requirements of the experiment was too small to justify the operation of the cupola for gray iron.

C

si

th

ti

sta

ta

th

siz

to

of

th

pe

un

Summary

The five factors studied were all highly significant in causing brass penetration in the sand. It cannot be determined by the statistical method which levels of each variable were significant nor can it be determined whether the interactions between any one or all of the variables were significant.

The five factors studied were not significant in causing penetration of aluminum in sand molds.

From a critical analysis of the brass samples, the static pressure and distribution variables were more important in causing penetration than were the other three variables of ramming, moisture content, and core setting time.

Equipment

Much of the equipment necessary to carry out this experiment was not available and had to be designed

and fabricated by the investigators prior to conducting the experiment.

A specially designed pattern as shown in Fig. 2-C was made to meet certain requirements of the experiment. To provide for a differential in static pressures, the pattern was designed to be poured on end. Figure 1-B shows the 11/8-in. coreprint for the sand test specimens, with the sand specimens set in place. The aluminum pattern was mounted on an aluminum match plate to facilitate the making of molds. The lighter weight and machineability of the aluminum alloy was the reason for this choice.

Since the mold had to be poured on end, conventional flasks were not suitable. The plywood flask pattern shown in Fig. 2-B was constructed to make the special flasks. The four lugs on the sides of the pattern were designed to provide a means of aligning the pattern and flask.

To facilitate the making of sand molds, the flasks were constructed from an aluminum alloy rather than steel. The lighter weight and resulting ease of handling was the main advantage of the aluminum alloy flasks. It was necessary to mill the faces of the flasks so that the pattern plate and flask would be matched for a close fit. Then the lugs on the side of the flask were drilled and reamed in alignment with the pattern plate to accommodate 5%-in. flask pins. A total of five flasks were made. All of the machining operations in fabricating the equipment were performed in the machine shop of the Industrial Engineering Department.

Preparation of Test Specimens

In this experiment, four levels of each variable were used in accordance with the requirements of the Greco-Latin Square. The four levels of silica sand grain distribution are shown in Table 2. Two of the distributions were obtained from foundry suppliers, and one was obtained from a local foundry. The fourth distribution (G-D) was prepared from the sand supply in the foundry laboratory. To get the proper distribution of G-D sand, it was necessary to sieve this sand in a Ro-Tap shaking machine. This work required two days to get 4000 grams of the desired sand separated. The sieve analysis of the other three sands was determined by the investigators in the laboratory. These four levels of grain distribution were chosen to get a range that would include fine, medium, and coarse-grained sands. The primary reason for using the G-D uniform sand mixture was to test the theory postulated by Sanders⁵ in which he states that better physical sand properties can be obtained from "grain sizing" or using a normal distribution of the sand grain sizes. Sanders further theorized that the lack of normal distribution of the sand grain sizes might cause metal penetration. The investigators found no evidence to substantiate Sanders' theory of metal penetration into the sand test core. The measurements of metal penetration in Table 7 show that the G-D sand mixture specimens had very little penetration and that the actual penetration occurred under conditions of greater alloy static pressure. This is significant since the G-D sand had a uniform plateau distribution as compared to the normal distribution curve of sand mesh sizes as advocated by Sanders.

Each of the four types of sand were mixed with four per cent Western bentonite as the bonding agent. All mixing was done by hand because of the small amounts of sand used. After the sand was thoroughly mixed, the correct amount of water was added. The per cent of moisture was also varied over four levels as shown in Table 2. The resulting sand mixtures were then stored in sealed jars to prevent the loss of moisture from the various sands. Moisture checks were made with a moisture teller prior to making the test specimens to determine whether or not the moisture content had changed, and no variation greater than one-tenth of one per cent was observed. The zero per cent moisture samples were obtained by ramming the various sands with the required number of rams and drying for 20 hr at 225 F in a constant temperature electric oven.

The mechanical properties of each of the sand mixtures was determined in the laboratory. To determine the permeability and green compressive strength, standard 2-in. sand test specimens were molded since permeability and green compressive strength machines were not available for the 1½-in. specimens. The specimen weights and mechanical test results are shown in Tables 3, 4, 5, and 6. The advantage of obtaining the specimen weights ahead of time was to facilitate the preparation of the sand test specimens prior to setting them in the molds. This saved considerable time and facilitated more accuracy when molding the test cores for use in the penetration experiments. Analytical balances were used to determine all measurements of weight throughout the

TABLE 3—PHYSICAL CHARACTERISTICS OF SAND TEST

	Per cent	No. of		2-in.	2-in.
Sand	Moisture	Rams	Weight, Grams	Specimen Perme- ability	Specimen Green Strength
G-D	3	2	48.0		
G-D	3	9	51.0		
G-D	3	15	51.7		
G-D	3	50	55.0		
G-D	3	2	146.0	76.0	2.4
G-D	3	9	155.0	48.0	4.6
G-D	3	15	160.0	39.5	5.8
G-D	3	50	168.0	27.6	7.2
G-D	5	2	49.0		
G-D	5	9	52.0		
G-D	5	15	53.0		
G-D	5	50	55.0		
G-D	5	2	151.0	70.0	2.4
G-D	5.	9	160.0	42.0	4.5
G-D	5	15	164.0	34.8	5.8
G-D	5	50	171.0	23.0	7.4
G-D	10	2	51.5		
G-D	10	9	54.5		
G-D	10	15	56.0		
G-D	10	50	59.0		
G-D	10	2	163.0	45.0	2.9
G-D	10	9	169.0	24.7	4.9
G-D	10	15	174.0	18.6	5.3
G-D	10	50	181	10.8	8.0

TABLE 4-PHYSICAL CHARACTERISTICS OF SAND TEST

	SPECIMENS										
Sand	Per cent Moisture	No. of Rams	Weight, Grams	2-in. ccimen Perme- ability	2-in. Specimen Green Strength						
A-7	3	2	47.5								
A-7	3	9	51.0								
A-7	3	15	51.5								
A-7	3	50	53.5								
A-7	3	2	151.0	500 plus	3.0						
A-7	3	9	155.0	450.0	4.7						
A-7	3	15	159.0	400.0	5.5						
A-7	3	50	165.0	280.0	7.3						
A-7	5	2	49.5								
A-7	5	9	51.5								
A-7	5	15	53.5								
A-7	5	50	55.0								
A-7	5	2	153.0	500.0	2.1						
A-7	5	9	156.0	375.0	4.3						
A-7	5	15	164.0	320.0	4.8						
A-7	5	50	171.0	245.0	5.8						
A-7	10	2	52.5								
A-7	10	9	54.5								
A-7	10	15	56.0								
A-7	10	50	58.0								
A-7	10	2	162.0	300.0	2.5						
A-7	10	9	169.0	210.0	3.8						
A-7	10	15	175.0	172.0	4.3						
A-7	10	50	179.0	118.0	6.3						

TABLE 5-PHYSICAL CHARACTERISTICS OF SAND TEST

		SPI	ECIMENS		
Sand	Per cent Moisture	No. of Rams	Weight, Grams	2-in. Specimen Perme- ability	2-in. Specimen Green Strength
A-85	3	2	49.0		
A-85	3	9	52.0		
A-85	3	15	53.0		
A-85	3	50	55.0		
A-85	3	2 .	150.0	34.0	2.7
A-85	9	9	159.0	22.5	4.5
A-85	3	15	163.0	19.4	5.9
A-85	3	50	170.0	13.9	7.6
A-85	5	2	54.0		
A-85	5	9	56.5		
A-85	5	15	57.5		
A-85	5	50	59.5		
A-85	5	2	153.0	40.0	3.0
A-85	5	9	161.0	23.3	5.5
A-85	5	15	165.0	19.0	6.4
A-85	5	50	173.0	13.4	8.3
A-85	10	2	53.5		
A-85	10	9	56.0		
A-85	10	15	57.0		
A-85	10	50	59.0		
A-85	10	2	160.0	27.0	3.9
A-85	10	9	170.0	15.0	6.3
A-85	10	15	174.0	11.7	7.6
A-85	10	50	182.0	7.1	10.0

experiment.

The penetration experiment was designed to utilize 1½-in. x 2-in. test specimens so that the mold cavity would not be too large because of limited melting

TABLE 6-PHYSICAL CHARACTERISTICS OF SAND TEST

	Per cent	No. of		2-in. *	2-in.
Sand	Moisture	Rams	Weight, Grams	Specimen Perme- ability	Specimer Green Strength
S-B-S	3	2	52.5		
S-B-S	3	9	54.5		
S-B-S	3	15	55.5		
S-B-S	3	50	57.5		
S-B-S	3	. 2	158.5	500 plus	4.2
S-B-S	3	9	167.0	500 plus	5.9
S-B-S	3	15	173.0	500 plus	6.8
S-B-S	3	50	180.0	375	8.5
S-B-S	5	2	54.0		
S-B-S	5	9	56.0		
S-B-S	5	15	57.0		
S-B-S	5	50	58.6		
S-B-S	5	2	164.5	500 plus	3.0
S-B-S	5	9	172.0	450	3.3
S-B-S	5	15	177.0	400	4.1
S-B-S	5	50	184.0	360	5.4
S-B-S	10	2	57.0		
S-B-S	10	9	58.6		
S-B-S	10	15	59.5		
S-B-S	10	50	61.0		
S-B-S	10	2	178.0	300	2.6
S-B-S	10	9	186.0	198	3.0
S-B-S	10	15	189.0	148	3.4
S-B-S	10	50	194.0	98	3.7

capacity. The smaller specimen size was also used because the 11/8-in. rammer has a double end ram whereas the 2-in. rammer produces a single end rammed specimen. It can be shown that double end ramming gives a more uniformly compacted test specimen.

tie

ba

Fi

hi

m

in

The Experiment

Heap sand, used for making molds for aluminum alloy and brass castings in the laboratory was used in preparing molds. After the mold was prepared, the test specimens were set in the core prints and the mold was closed. The time variable was measured from the time the mold was closed until the alloy was poured. It was found necessary to seal the parting surface of the mold with paste to prevent runouts. Because of the high static pressure encountered with brass, it was necessary to clamp cast iron plates on each side of the mold. When aluminum plates were used the pressure broke the plates and resulted in one mold bursting out through the side.

A pot-type gas-fired furnace was used to melt both the aluminum alloy and the brass. The largest crucible which the furnace would hold was number 40. This size crucible was sufficient to melt enough alloy for one mold. Pigs of aluminum and brass were used exclusively. It was felt that the use of scrap might vitiate the melt quality. The aluminum alloy was fluxed with zinc chloride to remove dross and other impurities. The brass was fluxed with phosphorcopper to improve its fluidity.

An attempt was made to hold the metal pouring

TABLE 7-AMOUNT OF BRASS PENETRATION IN SAND

No.	Time, Min	Moist- ure, %	Distri- bution	No. of Rams	Static Pressure, psi	Amount of Penetration In.
1	5	0	A-7	2	4.48	0.56
10	5	10	S-B-S	9	3.86	0.56
7	5	5	A-85	15	1.94	0.0
16	5	3	G-D	50	2.87	0.0
5	30	3	A-85	2	3.86	0.0
14	30	5	G-D	9	4.48	0.02
3	30	10	A-7	15	2.87	0.44
12	30	0	S-B-S	50	1.94	0.05
9	120	5	S-B-S	2	2.87	0.56
2	120	3	A-7	9	1.94	0.17
15	120	0	G-D	15	3.86	0.05
8	120	10	A-85	50	4.48	0.05
13	300	10	G-D	2	1.94	0.0
6	300	0	A-85	9	2.87	0.0
11	300	3	S-B-S	15	4.48	0.56
4	300	5	A-7	50	3.86	0.56

temperature constant. To do this, the metal temperature was raised slightly above the desired pouring temperature and allowed to cool to the desired temperature as measured by a calibrated thermocouple.

One of the more difficult parts of the experiment was to control the metal temperature so that it attained the desired level at the time the mold was ready to be poured in accordance with the time variable. The pouring temperature for the aluminum alloy was 1300 F and the brass was poured at 2150 F.

d

n

d

i-

n

n

e

d y tid es e

eis

r

To facilitate measurement of the metal penetration, the castings were sectioned in a horizontal plane passing through the center of each specimen cavity. The aluminum castings were sectioned on a 96-in. band saw. No penetration of aluminum alloy was observed in any of the test cores, however, several of the test cores were completely penetrated by brass. Figures 3 and 4 show the penetration of various brass samples and Fig. 5 and 6 show the aluminum samples. Severe penetration of the test cores by the brass hindered the sectioning of the brass castings. Various methods were tried to cut the brass castings including power saw, abrasive wheel and band saw. Prior to

sectioning it was necessary to grind off the surface penetration of the casting. Then the 48-in. band saw with a 6-pitch blade at 150 rpm was used to partly section the casting. To complete the sectioning it was necessary to break the castings at the penetrated specimens.

In measuring the amount of brass penetration it was necessary to get an average of many readings to determine the amount of penetration in the samples. The amount of penetration was determined from two samples, positioned opposite each other in the mold, and having identical variables. Three observers made 20 independent measurements of each test specimen to arrive at the final average penetration as shown in Table 7. The measurements were made with a steel machinist scale calibrated in sixty-fourths of an inch. Table 8 gives the average penetration for each sample measured, which is the average of four measurements taken at random on the sectioned surface of the sample.

The statistical devices used in this experiment were the Greco-Latin Square and the "F" test. These devices were used as tools in determining the significance of the variables in their effect on penetration. It must be remembered that the test was not designed to determine which level of each variable is important in its cause of penetration.

The Greco-Latin Square is an arrangement for the simultaneous study of several factors. The variations of each factor may be studied and their effects upon the experimental error may be excluded. A restriction of the Greco-Latin Square is that every factor must be used at least once in combination with every other factor studied in the experiment. An examination of Table 1 will show that the same combination of factors does not occur twice in any of the sixteen squares. This arrangement allows each level of each factor to be affected by all of the variables thus cancelling out the effects of the variables other than the one being studied. An example of this principle may clarify the meaning. The first row of Table 1 is concerned with the variable ramming, specifically two rams. Across that row it will be noted that every

TABLE 8

Sample No.	Measurement	Measu	Penetra prements n In.	, 0	ht Side) Average Penetration	Measurem	ent of B Measure in Ir		ration (Left	Side) Average Penetration
1	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56
2	0.12	0.16	0.11	0.11	0.12	0.25	0.18	0.20	0.20	0.21
3	0.56	0.56	0.56	0.56	0.56	0.35	0.38	0.32	0.26	0.33
4	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56
5	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00
6	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00
7	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00
8	0.05	0.06	0.05	0.05	0.05	0.04	0.05	0.03	0.03	0.04
9	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56
10	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56
11	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56	0.56
12	0.08	0.06	0.04	0.04	0.06	0.03	0.02	0.06	0.04	0.04
13	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00
14	0.01	0.01	0.01	0.01	0.01	0.03	0.03	0.03	0.03	0.03
15	0.08	0.02	0.06	0.07	0.06	0.07	0.01	0.03	0.07	0.04
16	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00	0.00

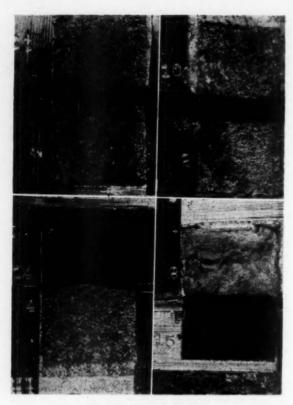


Fig. 3-Brass Samples.

other variable and each level of every other variable across row "A." Thus the effect of all the other factors are cancelled out and the effect of two rams is the level studied. The effect for sand distribution can be studied using the arrangement of Table 1 and considering the columns instead of rows. To study the other three variables, the Greco-Latin Square must be arranged so that the variables studied are either in columns or rows.

Calculation of the error variance (σ^2) in the analysis of the Greco-Latin Square experiment requires at least a pair of observations for each set of conditions. The mold was designed to have a left and right side enabling the experimenter to get a pair of samples in every mold.

The Greco-Latin Square arrangement does not allow a study of interactions among the variables. Interactions is the variation resulting from the combined effect of two or more factors. The effect of each factor operating independent of each other may be different than the total effect of the two factors operating together.

In experimental tests where two or more factors are being studied, the total variability of result may be placed in two broad classifications; the normal small sampling variation or unexplained variation, and the variation due to the effects of the controlled variables. The Greco-Latin Square can be used to determine how much each controlled factor contributed to the total variation.

The objective of the statistical method of analysis

TABLE 9

"F" Test $F = \frac{\sigma \, \hat{x}^2}{\sigma \, e^2} \qquad \sigma \, \hat{x}^2 = \frac{8 \, \Sigma \, (\hat{x} - \hat{x})^2}{\Sigma} \qquad = \frac{\Sigma \, \mathbb{B}}{3}$ $\frac{\sum \, \mathbb{B}}{z} \, \frac{\sum_{1}^{1V} \, (\frac{8}{\Sigma} \, x)^2}{\frac{1}{8}} \, - \frac{(\frac{32}{\Sigma} \, x)^2}{\frac{1}{32}}$ $\sigma \, e^2 \, = \frac{\frac{16}{\Sigma} \, \frac{2}{\Sigma} \, (x - \hat{x})^2}{\frac{1}{16}} \quad = \frac{\sum_{1}^{2} \, \mathbb{E}}{16}$ $\sum \, \mathbb{E} \, = \frac{\frac{332}{\Sigma} \, x^2 - \frac{16}{\Sigma} \, (\frac{2}{\Sigma})^2}{\frac{1}{2}}$ Check: $\sum \, x^2 \, - \, (\frac{\Sigma \, x})^2 \, \cdot \, \sum_{1}^{5} \, \Sigma \, \mathbb{E} \, + \Sigma \, \mathbb{E}$

Table of Factors

Factor	Z(Zx)2	$\frac{\sum (\sum x)^2}{8}$	ΣF	D. F	$\sigma_{\rm F}^2$	F
Sand	23, 974	2, 9967	1, 3947	3	0.4649	240
Moisture	13,480	1,6849	0.0829	3	0.0276	14. 25
Ramming	13.448	1.6810	0.079	3	0.0263	13.58
Time	13.819	1.7273	0.1253	3	0.0418	21.55
Position	15.364	1.9206	0.3185	3	0.1062	54. 79
Error			0.0310	16	0.00194	

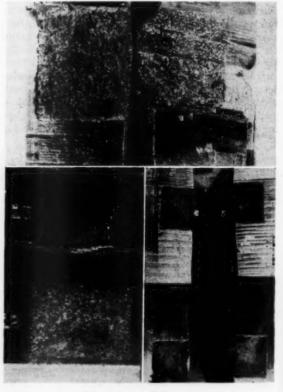


Fig. 4-Brass Samples.

is to determine if the means of the various levels of each factor differ enough to refute the hypothesis that they could come from the same population. Limits may be established on the expected or chance variation by an interpretation of the error variance (σ e²). If the variation exceeds these limits, the excess amount is due to identifiable causes.

The "F" test is designed to ascertain the significance of the differences among a set of means. A lack of significant differences as indicated by the "F" test intimates that the differences which do exist among means is due largely to chance. The hypothesis of homogeneity among means would then be accepted.

The value of "F" is defined as the ratio of the variance (σe^2) of a set of means to the estimate of the error variance. Since the distribution of "F" is known, the probability that the value of "F" could have occured by chance may be determined. If this probability is sufficiently small, the conclusion may be drawn that the hypothesis of homogeneity has been refuted.

In summing up these statements, it can be said that the value of the "F" test applied to the Greco-Latin Square is to tell if any of the variables were significant in causing penetration.

The value of "F" in this experiment at the 1 per cent significance level, i.e. the probability of 99 per cent that the variable caused penetration is 5.29.

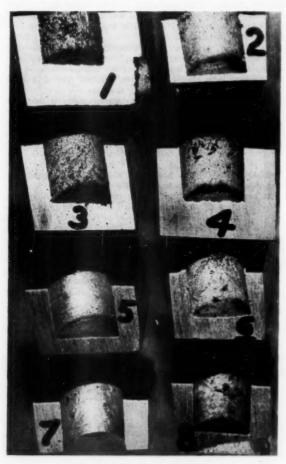


Fig. 5-Aluminum Samples.

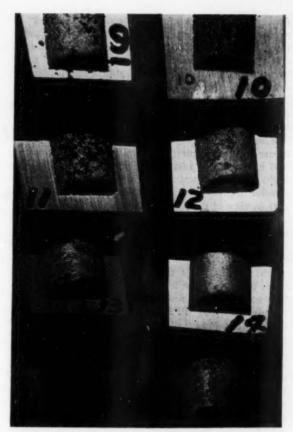


Fig. 6-Aluminum Samples.

Table 9 gives the values of "F" for each variable. In each case the significance is very strong that the variables caused penetration. It is so strong that not much faith can be put into the importance of relating the values one to the other to determine which is more significant. For example, sand distribution had an "F" value of 240 whereas the ramming variable had a value of only 13.58. But an "F" value of 13.58 is highly significant with the degrees of freedom encountered in this problem. Therefore it is impractical to say that sand distribution is much more important than ramming.

The conclusions that can be made from this statistical analysis is that each variable studied was highly significant in causing penetration. Since interactions between the variables cannot be analyzed in this particular statistical test, it can be assumed that either the variable itself is important or the individual variables reacting with each other are significant.

Discussion of Experiment

Disregarding the statistical significance of the data for a moment, some assumed conclusions can be drawn. It will be noted that the A-7 and 5-B-S sand test cores were completely penetrated in every position except the top position (number four position) where the static pressure was the least (1.96 psi). Neither of the other two types of sand had metal penetration greater than 0.06 in. regardless of the

static pressure. Two general assumptions may be made; (1) sand as coarse as 5-B-S sand is easily penetrated by brass, and (2) coarse sand subjected to a static pressure greater than 2.87 psi will be penetrated, by brass. Because of the difference in penetration in both the coarse sands between positions three and four (2.87 and 1.94 psi respectively), there is a possibility that a critical static pressure lies between these two points.

These general propositions would indicate an increase in metal penetration when the sand is coarse, and both fluidity and static pressure of the alloy are high. Further experimentation will be needed to determine the practical limitations of alloy static pressure in view of their affect on metal penetration.

The investigators failed to get any penetration whatsoever in the tests when using an aluminum alloy (95 Al-5 Si). This lack of aluminum alloy penetration was not expected especially in view of the varied conditions of the experiment. A clue which helps to explain the lack of aluminum alloy penetration was found in a research bulletin published by the National Bureau of Standards.6 It was found that the running quality of commercially pure aluminum, was a straight line function with the pouring temperature (independent variable) and the position of the curve translatable by the maximum temperature of the melt. Although fluidity and surface tension are but two of the factors involved in determining the running quality of an alloy, the running quality is significant in that a melt with a low value of fluidity is not conducive to penetration. The temperature at which the penetration experiments were carried out was 1250 to 1300 F which is representative of a very low running quality and not highly subjected to penetrating voids in the sand. Samples 10 and 15 in Fig. 6 show the extent of aluminum penetration into the very coarse 5-B-S sand. The metal did not penetrate beyond the voids in the first layer of sand grains and vigorous wire brushing or sand blasting could have removed these grains from the surface of the casting.

Acknowledgments

The writers wish to acknowledge the counsel and guidance of Dr. Douglas C. Williams during the period involved for the investigation. Appreciation is expressed to Mr. Loring G. Mitten for his assistance in setting up the statistical method for analyzing the investigation. Thanks is given to Mr. Joseph Anderson, laboratory technician, for his aid in carrying out the molding and pouring operations encountered in completing the investigation.

Bibliography

 Analysis of Castings Defects, American Foundrymen's Society, p. 89 (1947).

 W. A. Spindler, Foundry Technology, Ulrich's Book Store, Ann Arbor, Mich., Chapt. 5, p. 4 (1947).

3. J. B. Caine, "Sintering Test Sub-Committee Report, Transactions, American Foundrymen's Society, vol. 50, pp. 796-7

4. S. L. Gertsman and A. W. Murton, "An Investigation of Metal Penetration in Steel Sand Cores," Transactions, American Foundrymen's Society, vol. 58, pp. 595-603 (1950).

5. C. A. Sanders, "Controlling Sand Grain Distribution Reduces Steel Foundry Costs," AMERICAN FOUNDRYMAN, vol. 13, no. 4, p. 143, April 1948.

 A. I. Krynitsky and C. M. Saeger, Jr., "Effect of Melting Conditions on the Running Quality of Aluminum Cast in Sand Molds," *Journal of Research*, National Bureau of Standards, vol. 13, October 1934.

7. H. A. Freeman, Industrial Statistics, John Wiley & Sons, Inc., New York (1942).

8. P. G. Hoel, Introduction to Mathematical Statistics, John Wiley & Sons, Inc., London (1947).

DISCUSSION

Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

O. J. MYERS: Does this research tool obviate the necessity for reproducing these results by actually reproducing the experiments? Does that mean that sand is four times as important as the metal head in creating penetration?

MR. EKEY: It definitely would not obviate the necessity for further research. This particular method would lead to further research. We found that sand was significant, so the next step would be to set up a classical type of experiment with sand as the one variable.

We did not have partial intercorrelation. Take the time element. We got indications that time is a very important factor. We had a finer integration of our moisture variable, but there might be a partial correlation between time and static

pressure or moisture, probably the moisture.

S. L. Gertsman: I would like to amplify some of this work insofar as checking any single variable is concerned. Both in our work and that of others we have been trying to check one variable at a time. The Greco-Latin Square, however, appears to be a useful tool in carrying out any research in the foundry, since there are from 30 to 50 variables in foundry work. You wish to determine which variables have any bearing on a problem. Therefore, I think it is a good idea to bring this technique to the attention of foundrymen.

In regard to the author's statement that Dr. Williams insisted that one of the variables which should be attacked is the effect of leaving the core in the mold anywhere up to 300 min before pouring. There might be a loss of moisture by leaving the core in the mold up to 300 min before pouring and that might have some effect on penetration.

MR. EKEY: I would merely like to repeat that we would have a finer integration of our moisture variable due to this sitting time, which goes along with what Mr. Gertsman mentioned.

¹Director of Foundry Res.. Archer-Daniels-Midland Co., Minneapolis.
²Head, Steel Section, Bureau of Mines & Tech. Surveys, Ottawa, Ont., Canada.

EFFECT OF VANADIUM ON THE PROPERTIES OF CAST CHROMIUM-MOLYBDENUM STEELS

By

N. A. Ziegler, W. L. Meinhart and J. R. Goldsmith

ABSTRACT

An investigation was made on (a) 1.25 per cent chromium — 0.5 molybdenum, (b) 2.5 per cent chromium — 0.5 molybdenum, and (c) 5.0 per cent chromium — 0.5 per cent molybdenum cast steels, in which vanadium ranged from 0 to 0.3 per cent and carbon from 0.1 per cent max to 0.3 per cent in each group. Results obtained in the various tests on characteristics and properties are presented.

The presence of vanadium has a slight tendency to increase the thermal sluggishness and air hardenability.

Increasing vanadium reduces the ferrite and pearlite grain size, produces some extremely fine spheroidal carbides, but makes the dendritic pattern more pronounced, coarse and persistent.

In general, vanadium moderately increases the tensile strength and yield point, slightly reduces the ductility and to a greater extent reduces the impact resistance.

Metal arc welding of these steels presented no difficulties with regard to cracks and voids. Ductility of the resultant weld assemblies was somewhat decreased by the larger percentages of vanadium.

Introduction

THE PRESENT ARTICLE is a continuation of the authors' study of the properties of cast low-alloy steels, the previous steps of which have been presented in a series of publications.³⁻⁷ One of these papers⁶ describes the effect of vanadium on the properties of carbon and carbon-molybdenum steels, while another⁷ deals with the characteristics of low chromium-molybdenum steels. Compositions of the latter article fall within the range of steels recommended for superheated steam service.⁸

It is common knowledge that vanadium in amoun's up to about 0.25 per cent improves the high-temperature creep resistance and refines the grain structure of carbon-molybdenum and chromium-molybdenum steels. On the other hand, its effect on room temperature physical properties, thermal behavior and weldability has not been too well known. This is particularly true about castings. The authors' earlier publication, clarifies some of these points regarding

the carbon and carbon-molybdenum types. The present one is an attempt to further the knowledge with respect to chromium-molybdenum-vanadium steels.

A fairly complete bibliography of the work done on vanadium as an alloying element in irons and steels was presented in a previous article. One appended to this paper 2 gives a few additional references, thus bringing it up to date.

High temperature properties and graphitization susceptibility of chromium-molybdenum-vanadium steels are being studied.

Experimental Procedure

The technique of preparation, heat treating and processing of the experimental compositions was the same as previously described.¹⁻⁷ Each heat of about 200 lb was cast into (a) keel coupon blocks and (b) welding plates. Composition-wise they are divided into three groups: (a) 1.25% Cr - 0.5% Mo, (b) 2.5% Cr - 0.5% Mo and (c) 5.0% Cr - 0.5% Mo.

Table 1 is a compilation of nominal compositions, chemical analyses, dilatometer data and physical properties of all of these steels. The first two groups are divided into four subgroups, containing 0.1 max, 0.15, 0.20 and 0.30 per cent carbon, respectively. The third group is composed of only three subgroups, containing 0.1 max, 0.15 and 0.30 per cent carbon. Each subgroup consists of four or more steels with vanadium ranging between 0 and about 0.30 per cent.

Coupon blocks were used for making tensile* and impact** test bars. Test bars of the 1.25% Cr - 0.5% Mo group were heat treated (before machining) in two different ways: (a) triple, i.e. normalizing from 1750 F (950 C), air quenching from 1550 F (840 C) and drawing at 1250 F (680 C) and (b) double, i.e. normalizing from 1750 F (950 C) and drawing at 1250 F (680 C). The 2.5% Cr - 0.5% Mo and 5.0% Cr - 0.5% Mo groups were subjected to the triple heat treatment only. It has been the authors' practice to use the double heat treatment for carbon and carbon-molybdenum steel modifications (1.4,5,8 and 7) and

Of the authors, N. A. Ziegler is Research Metallurgist, W. L. Meinhart is Assistant Research Metallurgist, and J. R. Goldsmith is Metallurgical Engineer in charge of research foundry, Crane Co., Chicago.

^{*} Standard 0.505-in. diam, 2-in. gage test bar.

^{**} Standard Charpy test bar with keyhole notch.

triple—for chromium-molybdenum types (2.3 and 7) with chromium above 1.25 per cent. The low-chromium group of the present paper serves as a "connecting link," between carbon-molybdenum and chromium-molybdenum groups of steels. Hence, it was given both heat treatments, so that the properties of its constituent steels could be directly compared with those of any composition presented in previous publications.

All test bars thus heat treated were subjected to tensile, Vickers hardness and Charpy impact testing. The latter was conducted (a) at room temperature and (b) at 1000 F (570 C). Dilatometric thermal analyses was made of all of these steels except one.

As can be seen from Table 1, the minor elements (Si, Mn, S and P) as well as molybdenum, are reasonably uniform. Carbon, chromium and vanadium (within experimental error) vary as previously specified, thus permitting certain studies to be made.

Thermal Analysis and Microexamination

The results of thermal analysis are presented in Table 1. The thermal cycle in each case was to heat the sample*** in vacuum to 1830 F (1000 C) in 2 hr, and then cool to room temperature (a) in 5hr† and (b) "in air."††

Selected dilatometer curves are compiled in Fig. 1 and 2 for 1.25% Cr - 0.5% Mo, in Fig. 3 and 4 for 2.5% Cr - 0.5% Mo and in Fig. 5 and 6 for 5.0% Cr - 0.5% Mo steels. In each one of these compilations, dilatometer curves are arranged in either four (Fig. 1, 2, 3 and 4) or three (Fig. 5 and 6) subgroups with

increasingly higher carbon.

It was found that vanadium in amounts up to about 0.3 per cent exerts a relatively small effect on the appearance of most of the thermal curves. Hence, within each subgroup in the above six graphs, only two curves are recorded: one representing a steel with no vanadium, and the other with the highest (for this particular subgroup) vanadium content. Curves representing intermediate vanadium compositions are omitted to conserve space.

Figures 1, 3 and 5 are compilations of dilatometer curves obtained on slow (5 hr) and Fig. 2, 4 and 6 are those obtained on more rapid cooling ("in air").

After completion of the dilatometer cycle, each sample was subjected to hardness testing (Table 1) and micro-examination.

1.25% Chromium - 0.5% Molybdenum Group

Within each of the four pairs of thermal curves (Fig. 1 and 2), transformation-on-heating of the vanadium-bearing steel occur (as should be expected) at temperatures a little higher than those for the corresponding vanadium-free variety. On slow cooling (Fig. 1) no suppressed transformations can be observed in any one of the eight curves which is also substantiated by micro-examination. However, the hardness of vanadium-bearing compositions, containing carbon in the amounts of about 0.15 per cent and

higher (Serial No. 8, 12 and 16) is greater than that of similar ones but free from vanadium (Serial No. 5, 9 and 13). The two very low carbon compositions (Serial No. 1 and 4) have practically identical hardness, regardless of vanadium.

It is also interesting to note that the low carbon 0.29 per cent vanadium steel (Serial No. 4), after slow cooling, does not contain any normal pearlite. All carbon is present in the form of extremely fine spheroidal carbides, some of which are arranged in a faint lameller-like pattern, as can be seen in Fig. 7. This type of structure was previously observed and reported by the authors in low-carbon-molybdenum-vanadium steels.6

The latter two compositions (Serial No. 1 and 4) maintain relatively low and almost identical hardness values even after the rapid cooling (Fig. 2). Nevertheless, both of them, with this cooling rate, display initial stages of suppressed transformations which is also substantiated by the presence of small amounts of Widmanstätten type of structural constituent (structure not shown).

With carbon gradually increasing, the thermal sluggishness, revealed by the more rapid cooling rate (Fig. 2), as well as the resultant hardness, also increase. This is somewhat more pronounced in the vanadium-bearing (Serial No. 8, 12 and 16, Fig. 2) than in vanadium-free steels (Serial No. 5, 9 and 13, Fig. 2)

Progressive increase of the product of suppressed transformation, caused by the increase in carbon content, and relative microhardness of structural constituents can be observed in Fig. 8 and 9. In the former (0.15% C, 0.25% V) the ferritic matrix measures about 260 and the suppressed transformation product (bainite) about 300 micro-Vickers (VPN). In the latter (0.23% C, 0.33% V) bainite, with the hardness of over 300 micro-Vickers (VPN), is the predominating constituent. Areas rich in ferrite are softer.

Another interesting observation (as shown in Fig. 10 and 11) was made regarding the grain size of vanadium-free versus vanadium-bearing steels. As one might expect, the latter has finer ferrite and pearlite grains. On the other hand, the dendritic pattern in the latter is coarser and more pronounced.

2.5% Chromium - 0.5% Molybdenum Group

In this group the effect of vanadium on the transformations-on-heating, as well as on the thermal sluggishness and resultant hardness, is less pronounced, both on slow (Fig. 8) and on more rapid (Fig. 4) cooling. In the former case, fully suppressed transformations are observed only in the higher carbon steels (Serial No. 31 and 36). In the latter, all steels, with the exception of the two low-carbon ones (Serial No. 17 and 20), display fully-suppressed transformations.

Figure 12 is another demonstration of spheroidization of carbides and elimination of pearlite, caused by the presence of vanadium, which is more pronounced in the low-carbon steels (Serial 20, Fig. 3).

Figures 13, 14 and 15 present a microhardness study of different structural constituents in vanadium-free

^{*** 50-}mm long and 4 mm in diameter.

[†] Average cooling rate of 3.2 C/min or 5.8 F/min.

^{††} Average cooling rate of 50 C/Min or 90 F/min.

d -). ye

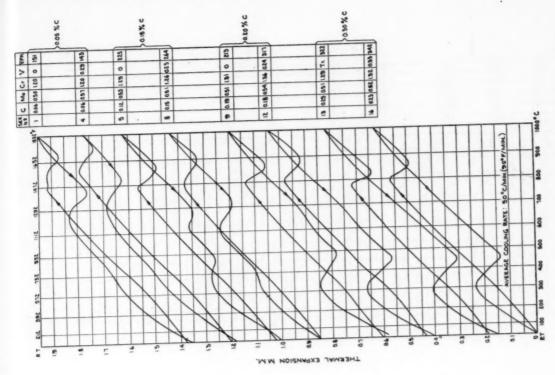


Fig. 2—Thermal curves of 1.25 Cr-0.5 Mo steels containing various amounts of carbon and vanadium. (Rapid cooling rate).

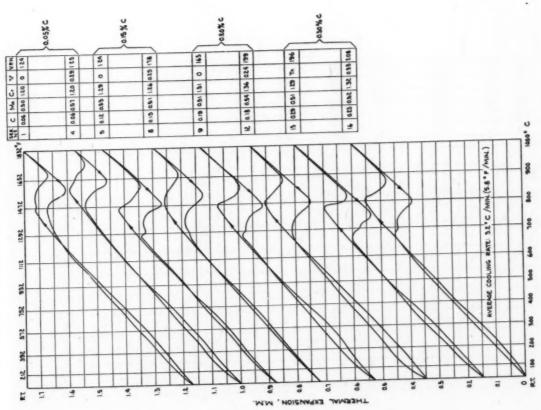
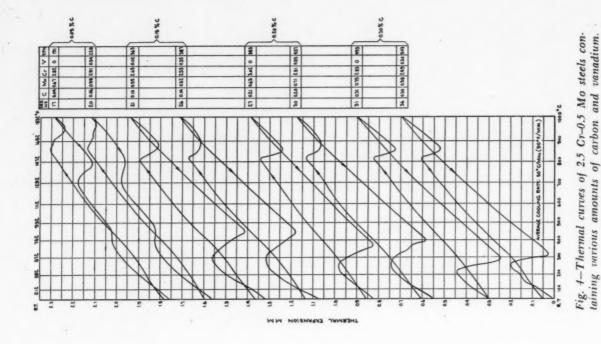


Fig. 1—Thermal curves of 1.25 Cr-0.5 Mo steels containing various amounts of carbon and vanadium. (Slow cooling rate).

.

00 4

*



> 0.30% C 0.05 % C > 0.20 % C 015%C 30 020 011 281 025 05 26 0.14 0.52 2.53 0.35 2:0 173 O 288 840 13.0 TS TEN 0 583 870 180 18 34 030 058 248 030 44 SER C NO C. V VIN 80 006 059 251 034 128 E1 0.16 0.55 2.65 0.02 191 Fig. 3-Thermal curves of 2.5 Cr-0.5 Mo steels containing various amounts of carbon and vanadium. 17 0.04 0.67 2.82 O (Slow cooling rate). AVERAGE COOLING RATE: S.E.C/MIN. (S.B. F/MIN.) 908 909 300 400 500

9

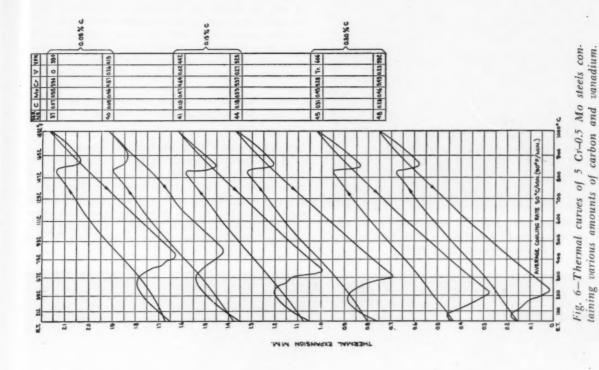
THERMAL EXPANSION MM.

5

(Rapid Cooling rate).

LS

(Rapid cooling rate).



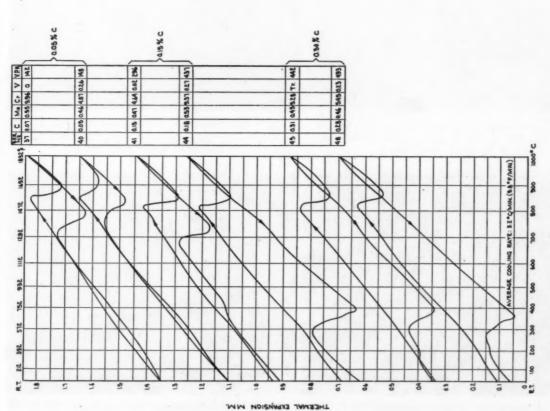


Fig. 5-Thermal curves of 5 Cr-0.5 Mo steels containing various amounts of carbon and vanadium. (Slow cooling rate).

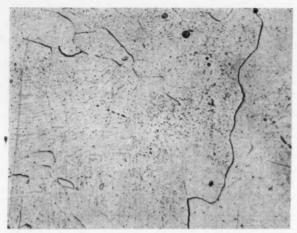


Fig. 7-Microstructure of steel containing 0.06 C, 1.20 Cr, 0.51 Mo, 0.29 V. Serial No. 4, slow cooling rate. Etched in nital. Mag. 500×.

(Fig. 13, Serial No. 31) and vanadium-bearing (Fig. 14 and 15, Serial No. 36) steels. The former (Fig. 13), after slow cooling displays some ferrite and pearlite-like constituent, both of which are relatively soft, in a martensite-bainite matrix. In the latter (Fig. 14), the same cooling rate produces a martensite-bainite structure throughout. With a more rapid cooling rate (Fig. 15), the ratio of martensite to bainite in the vanadium-bearing steel increases, which explains the increase in microhardness of the martensitic portions.

5.0% Chromium - 0.5% Molybdenum Group

In this group of steels the effect of vanadium on transformations-on-heating and on the resultant hardnesses shown in Fig. 5 and 6 is even less pronounced than in the 2.5% Cr series. The vanadium-bearing steel (Serial No. 44) is definitely harder than that without vanadium (Serial No. 41), only in the intermediate-carbon pair, both after the slow (Fig. 5) and

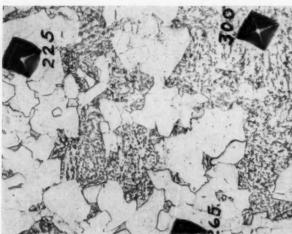


Fig. 8-Microstructure of steel containing 0.15 C, 1.26 Cr, 0.51 Mo, 0.25 V. Serial No. 8, rapid cooling rate. Etched in nital. Mag. 500×.

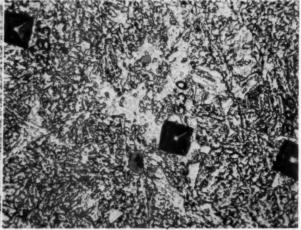


Fig. 9-Microstructure of steel containing 0.23 C, 1.32 Cr, 0.52 Mo, 0.33 V. Serial No. 16, rapid cooling rate. Etched in nital. Mag. 500×.

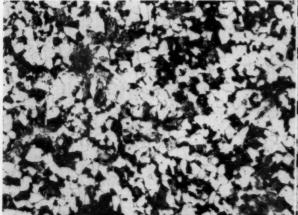


Fig. 10-Microstructure of steel containing 0.29 C, 1.29 Cr, 0.51 Mo, no V. Serial No. 13, slow cooling rate. Etched in nital. Mag. 100×.

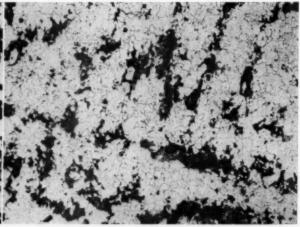


Fig. 11-Microstructure of steel containing 0.23 C, 1.32 Cr, 0.52 Mo, 0.33 V. Serial No. 16, slow cooling rate. Etched in nital. Mag. 100×.

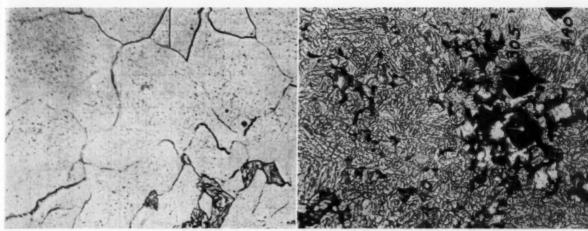


Fig. 12-Microstructure of steel containing 0.06 C, 2.91 Cr, 0.59 Mo, 0.34 V. Serial No. 20, slow cooling rate. Etched in nital. Mag. 750×.

Fig. 13—Microstructure of steel containing 0.31 C, 2.83 Cr, 0.75 Mo, no V. Serial No. 31, slow cooling rate. Etched in nital. Mag. 500×.

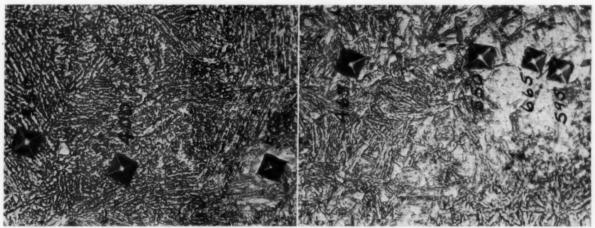


Fig. 14—Microstructure of steel containing 0.30 C, 2.49 Cr, 0.58 Mo, 0.30 V. Serial No. 36, slow cooling rate. Etched in nital. Mag. 500×.

Fig. 15—Microstructure of steel containing 0.30 C, 2.49 Cr, 0.58 Mo, 0.30 V. Serial No. 36, rapid cooling rate. Etched in nital. Mag. 500×.

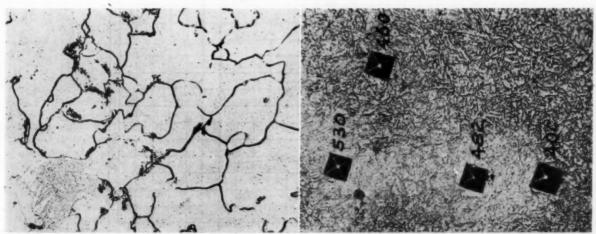


Fig. 16—Microstructure of steel containing 0.05 C, 4.87 Cr, 0.46 Mo, 0.25 V. Serial No. 40, slow cooling rate. Etched in nital. Mag. 500×

Fig. 17—Microstructure of steel containing 0.18 C, 5.37 Cr, 0.53 Mo, 0.27 V. Serial No. 44, rapid cooling rate. Etched in nital. Mag. 500×.

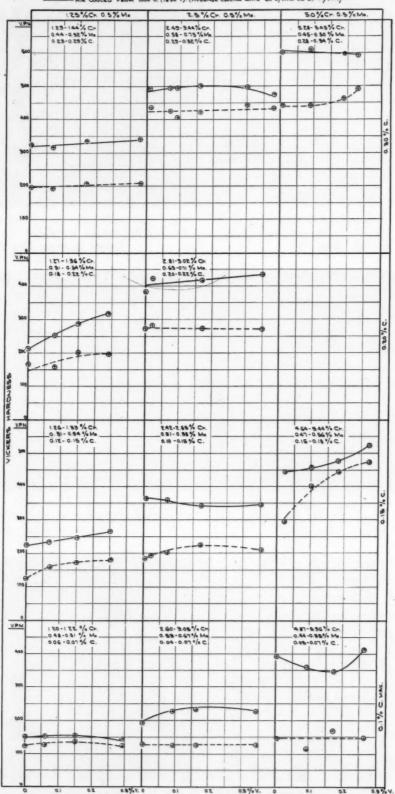


Fig. 18-Effect of vanadium on hardness of dilatometer samples of Cr-Mo steels.

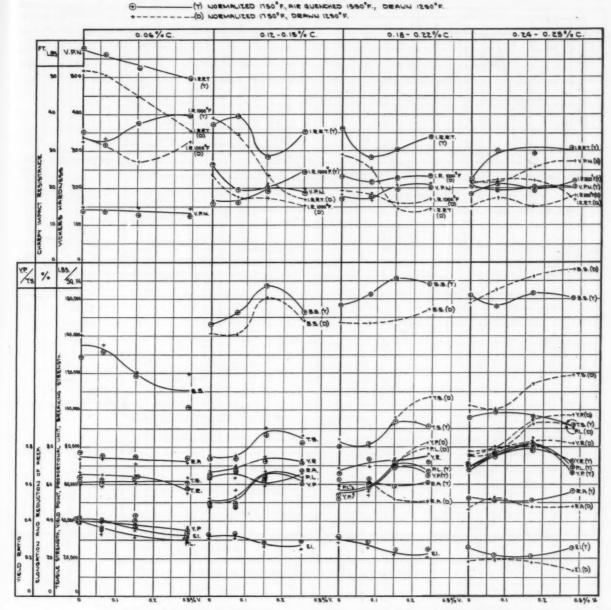


Fig. 19-Effect of vanadium on physical properties of 1.25 Cr-0.5 Mo steels.

after the more rapid (Fig. 6) cooling. This might be explained by the fact that in this particular composition the addition of vanadium and slightly higher carbon has a greater relative effect on the hardenability. For this reason, it is hard even after slow cooling. On the other hand, both low-carbon steels (Serial No. 37 and 40) are soft, and the high-carbon ones hard (Serial No. 45 and 48) regardless of the vanadium content or the cooling rate (Fig. 5 and 6).

Figure 16 illustrates the structure of the low-carbon vanadium-bearing composition (Serial No. 40), after slow cooling (Fig. 5). Similarly to the structures shown in Fig. 7 and 12, pearlite is absent and carbides appear as fine spheroidal particles, although in

the present case the grain size is somewhat finer than in the two preceding ones.

The structure of Fig. 17 (Serial No. 44, rapidly cooled) is typical of all of those steels in which transformations are completely suppressed. It is composed of somewhat softer bainite and harder martensite. In higher carbon compositions, the ratio of martensite to bainite increases.

Hardness of Dilatometer Samples

Figure 18 graphically shows the hardness values of dilatometer samples (a) after slow, and (b) after more rapid cooling. Effect of vanadium is recorded separately for each subgroup (i.e. carbon-chromium

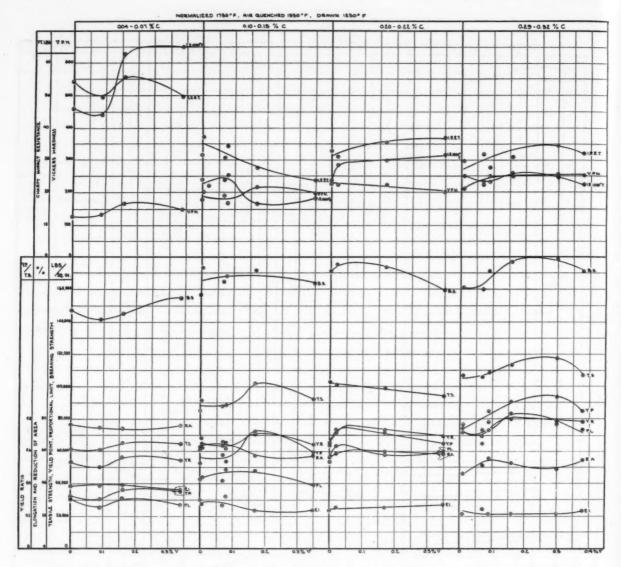


Fig. 20-Effect of vanadium on physical properties of 2.5 Cr-0.5 Mo steels normalized at 1750 F, air quenched at 1550 F and tempered at 1250 F.

combinations). Briefly, the effect of vanadium is slight, except for two (1.25% Cr - 0.20% C and 5% Cr - 0.15% C) subgroups. Even then, however, it more likely might be due to some interfering factors or experimental error, rather than to vanadium.

Physical Properties

Figures 19, 20 and 21 are graphic representations of the effect of vanadium on various physical properties of 1.25% Cr - 0.5% Mo, 2.5% Cr - 0.5% Mo and 5.0% Cr - 0.5% Mo groups of steels, respectively. Each figure is divided into four (Fig. 19 and 20) or three (Fig. 21) graphs, each of which represents a different carbon level within the given group. Each point is an average of two or more tests.

As previously pointed out, the 1.25% Cr - 0.5% Mo group of steels was subjected to two different heat

treatments (Table 1). Hence, in Fig. 19 two sets of test data have been plotted. It might be noted that the effect of the heat treatments on tensile properties and hardness was relatively small. Only in the high carbon subgroup was this difference sufficiently large to warrant separate plotting. On the other hand, the impact resistance was strongly affected in all cases. In general, the double heat treatment tends to make these steels stronger but less ductile and impact resistant, while the triple heat treatment improves ductility and impact resistance with a sacrifice in tensile strength and yield point.

Considering Fig. 19, 20 and 21 together, it might be noted that the effect of vanadium on all properties, regardless of chromium and carbon contents, is relatively small. The general trend is to increase the strength and yield point and to reduce the ductility

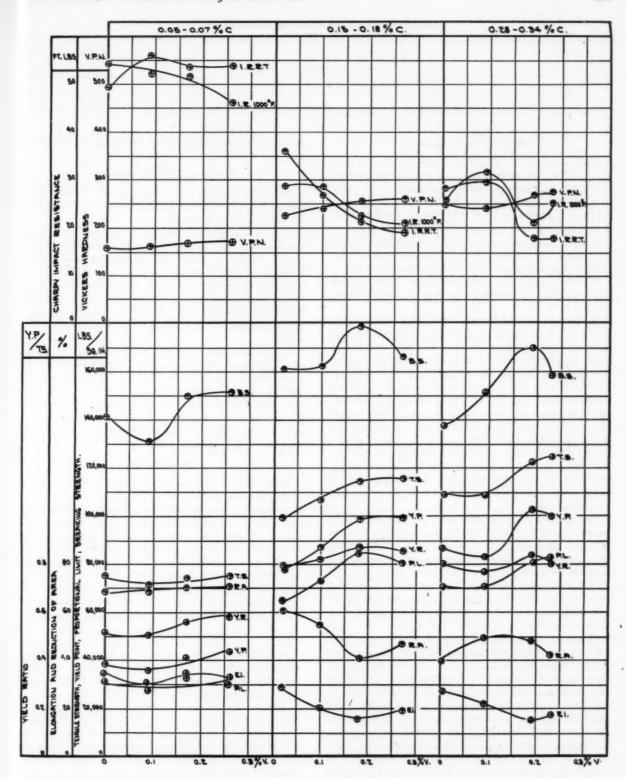


Fig. 21-Effect of vanadium on physical properties of 5 Cr-0.5 Mo steels normalized at 1750 F, air quenched at 1550 and tempered at 1250 F.

and room temperature impact resistance. The effect on the impact resistance is particularly noticeable in

the 1.25% Cr · 0.5% Mo group (Fig. 19) after double heat treatment.

TABLE 2-CHEMICAL ANALYSIS OF WELD DEPOSITS

GROUP	Si	Mn	S	P	C	Cr	* Mo	V
1.25% Cr - 0.5% Mo	0.33	0.65	0.029	0.017	0.10	1.10	0.55	None
	(-	1-8	_	-	0.07	2.28	0.52	0.20
2.5% Cr - 0.5% Mo	0.44	0.87	0.015	0.021	0.08	2.35	0.45	None
5.0% Cr - 0.5% Mo	_	-	-	-	0.06	5.24	0.50	0.22

Welding Experiments

Each composition listed in Table 1 was cast into 7/8 in. thick pairs of welding plates, 4 all of which were subjected to the triple heat treatment. They were prepared for welding by machining one edge of each to form a standard A.S.A. "U" bevel (20-deg bevel and a 3/16 in. root radius). Each pair of plates of the same composition was then welded together by the metal arc process, using lime-coated steel electrodes (EXX15 Class, ASTM-A316 Specification).

Table 2 gives the compositions of representative weld deposits of the three groups of steels, respectively.

It should be noted that the electrodes were selected so that their composition (particularly chromium and molybdenum) matched that of the parent metal as close as possible. There was no vanadium-bearing electrode available for the 1.25% Cr - 0.5% Mo group. Likewise, only a limited amount could be found to accommodate the 2.5% Cr - 0.5% Mo group. Hence, all of the steels of the former group and about half of those of the latter were welded with vanadium-free electrodes. In connection with the present experimentation, the interest is in the behavior of the parent metal rather than in that of the weld deposit. For this reason, exact matching of vanadium in the weld to that of the parent metal was not considered too important.

All plates were preheated with oxy-acetylene flame to 400 to 500 F (200 to 260 C) and welded with 7 to 8 passes on assemblies 7/8 in. thick with a 7-in. running length of welding. Soundness of the welds was verified by radiographing them all.

Affected zones in most of these weld assemblies were so hard that it was impractical to cut and machine them into test pieces. For this reason, all of them were stress-relieved at 1350 F (732 C).

Standard tensile* and Charpy impact** bars (two of each per weld) were machined from the stress relieved welds. In the former the weld and the affected zone were left between the gage marks. In the latter, the keyhole notch was cut through the weld metal to the fusion line in an attempt to force the break in the heat-affected zone adjacent to the weld, as illustrated in a previous publication.⁴ All of these test bars were subjected to testing, but the resultant data were omitted to conserve space, as similar properties of parent metals were given in Table 1 and Fig. 19, 20 and 21.

A side-bend-test (3/8 by 7/8 by 10 in.) was made in a jig for each weld assembly, and each was also sub-

jected to hardness distribution studies across the weld and the affected zone.

Tensile and Impact Testing

In tensile testing, all samples of the 1.25% Cr-0.5% Mo group (Table 1, Serial No. 1 to 16) broke through the parent metal.

The two subgroups, containing 0.1 per cent max and 0.15 per cent carbon of the 2.5% Cr - 0.5% Mo group (Table 1, Serial No. 17 to 26), also broke through the parent metal. Some of the samples of the 0.20 per cent carbon subgroups (Serial No. 27 to 30), broke through the parent metal, and others through the weld. Samples of the 0.30 per cent carbon subgroup (Serial No. 31 to 36) all fractured through the weld.

In the 5.0% Cr - 0.5% Mo group, only the samples of the 0.1 per cent max carbon subgroup (Table 1, Serial No. 37 to 40) broke through the parent metal, and all the rest (Serial No. 41 to 48) through the weld. In general, all measured properties of the weld assemblies compare quite favorably with those of the parent metal, even though their ductility and impact resistance are somewhat lower.

Bend Tests

Longitudinal sections for the side bend tests were prepared in accordance with the American Standards Association Code for Pressure Piping requirements and bent at the weld zone around a 1½-in. mandrel. Of the 48 pieces thus tested, only five (namely Serial No. 30, 35, 44, 47 and 48, Table 1) broke before 180 degrees was reached. The others all passed the Code requirements even though some of them had very short fine cracks in the weld and the affected zone. The frequency of these cracks tends to mount with the increasing chromium, vanadium and carbon.

Hardness Distribution Studies

A cross-section from a weld of each composition was subjected to a hardness distribution survey across the weld and the affected zone, using a Vickers pyramid hardness tester with a 30-kilogram load.

The results of these tests are graphically represented in Fig. 22, separately for each chromium-carbon subgroup. In general, hardnesses of the welds in the low-carbon subgroups are higher than those of the parent metal. This difference becomes less pronounced with the increase in carbon and/or chromium. Likewise, the overall hardness tends to increase with the increasing carbon and/or chromium. No definite correlation can be observed within each subgroup between the vanadium content and the position of the corresponding hardness distribution curve.

^{* 2-}in. gage, 0.505-in. diameter.

^{**} Keyhole notch.

0

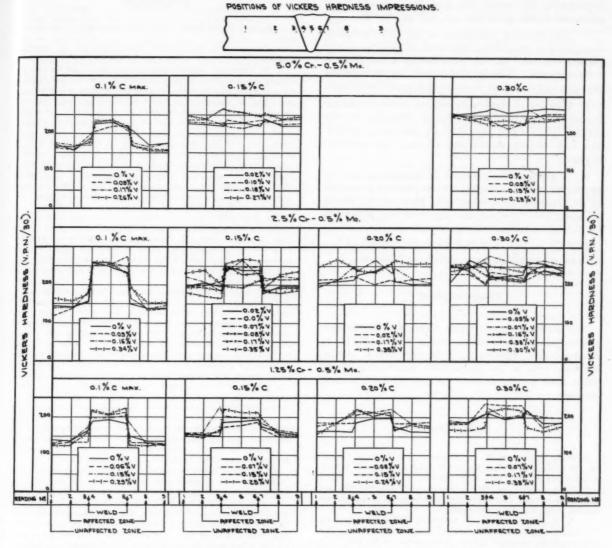


Fig. 22—Hardness distribution across the weld and the affected zone in Cr-Mo-V steels, stress relieved at 1350 F.

As previously pointed out, all of these tests were performed on stress-relieved samples, which does not reveal the variations present in "as welded" properties. However, some idea as to what might be expected in the affected zones of these steels immediately after welding can be obtained from the study of the hardness values of the dilatometer samples (Fig. 18), and particularly of those cooled at a rapid rate. Study of the dilatometer curves also contributes to the understanding of the thermal properties (and consequently weldability) of these steels.

Summary

1. The effect of increasing vanadium content between 0 and 0.30 per cent on the properties of (a) 1.25% Cr - 0.5% Mo (b) 2.5% Cr - 0.5% Mo and (c) 5.0% Cr - 0.5% Mo cast steels has been investigated. Carbon content within each of these groups

ranged between about 0.1 per cent max and 0.30 per cent.

2. Dilatometric analysis shows that vanadium tends to raise transformation temperatures on heating, particularly for low-carbon and lower-chromium steels.

3. Vanadium has a slight tendency to increase thermal sluggishness and air-hardenability, but this effect is quite small.

4. Vanadium tends to spheroidize carbides. In low carbon-high vanadium compositions pearlite frequently is completely absent. This is more pronounced after slow cooling from about 1830 F (1000 C), comparable to annealing, and irrespective of the chromium content.

5. Spheroidization of carbides is closely associated with the increased transformation temperature and closing of the gamma loop, which might result in the presence of residual delta ferrite.

6. Vanadium reduces the size of ferritic and pear-

litic grains, but makes the dendritic structure more pronounced and persistent, as well as coarser.

- 7. The effect of vanadium on all room temperature physical properties, regardless of chromium and carbon contents (in the presence of 0.5% Mo), is relatively small. The general trend is to increase the strength and to reduce ductility and impact resistance.
- 8. Welding of these compositions by the metal arc process, employing heavy coated electrodes of the AWS EXX15 type, presented no particular difficulties with respect to cracking or voids in the weld deposits.
- 9. In general, the presence of vanadium does not appear to have much effect on the physical properties of the resultant weld assemblies. However, all five of 48 welds which did not pass the bend test, prescribed by the Code, contained high vanadium.
- 10. No attempt is being made to recommend any definite composition for a specific application.

Acknowledgment

The authors wish to thank Crane Co. for permission to publish these results and to A. F. Lahr, H. A. Peterson, B. E. Jenkins, R. A. Mueller and P. Hjulin for their help in the preparation and testing of the steels, and to the Control Laboratory for chemical analyses.

Bibliography

1. N. A. Ziegler and W. L. Meinhart, "The Effect of Copper on the Properties of Cast Carbon-Molybdenum Steels," Trans-ACTIONS, A.F.A., vol. 52, p 1151 (1944).

2. N. A. Ziegler and W. L. Meinhart, "Characteristics and Properties of Some Cast Chromium-Molybdenum Steels," Trans-

actions, ASM, vol. 34, p 589 (1945).
3. N. A. Ziegler and W. L. Meinhart, "Effect of Nickel on Physical Properties and Thermal Characteristics of Some Cast Chromium-Molybdenum Steels," Transactions, ASM, vol. 37, p

4. N. A. Ziegler, W. L. Meinhart and J. R. Goldsmith, "The Effect of Managanese on the Properties of Cast Carbon and Carbon-Molybdenum Steels," Transactions, ASM, vol. 38, p 398

(1947).

5. N. A. Ziegler, W. L. Meinhart and J. R. Goldsmith, "The Effect of Silicon on the Properties of Cast Carbon and Carbon-Molybdenum Steels," Transactions, ASM, vol. 40, p. 617 (1948).

- 6. N. A. Ziegler, W. L. Meinhart and J. R. Goldsmith, "Effect of Vanadium on the Properties of Cast Carbon and Carbon-
- Molybdenum Steels," *Transactions*, ASM, vol. 41, p. 565 (1949).
 7. N. A. Ziegler, W. L. Meinhart and J. R. Goldsmith, "Characteristics and Properties of Cast Low Chromium-Molybdenum Steels, Transactions, ASM, vol. 42, p. 175 (1950).

8. 1947 Supplement to Book of ASTM Standards, Including

Tentatives, Part 1-A, Ferrcus Metals, p 268-273.

9. T. W. Merrill, "Vanadium Data Sheets," The Vancoram Review, vol. 4, no. 4, 1945, p 14; vol. 5, no. 1, 1946, p 10; vol. 6, no. 2, 1950, p 7.

10. F. F. Franklin, "Service Properties of Manganese-Vanadium Plate Steels," The Vancoram Review, vol. 5, no. 3, 1948,

11. J. F. Wallace, "Effect of Aluminum and Vanadium on Toughness of High Hardenability Cast Steels," Transactions, A.F.S., vol. 57, p 269 (1949). 12. S. L. Case, "Vanadium as a Deoxidizer," *Iron Age*, vol.

164, no. 20, Nov. 17, 1949, p 97.

DISCUSSION

Chairman:-FRANK KIPER, Ohio Steel Foundry Co., Springfield.

Co-Chairman: G. W. JOHNSON, Vanadium Corp. of America. Chicago

T. W. MERRILL: 1 I would like to compliment the authors on adding another block to this monumental work of investigating the properties of cast steels. We enjoyed reading this paper on the effect of vanadium in chromium-molybdenum steels, although we were somewhat disturbed by the summary. In analyzing this feeling we decided that we were prejudiced by years of thinking in terms of the effect of vanadium in lower alloy steels. Many of the steels investigated by the authors have a relatively high alloy content with respect to carbon content, which certainly influences the effects of small additions of vanadium. We feel that it is essential to keep this in mind when reading the conclusions and not to make the mistake we subconsciously did in assuming that vanadium has such effects on alloy cast steels generally.

The dilatometer curves constitute a valuable addition to the knowledge of vanadium's effects. It has been observed by others that vanadium raises the transformation temperature slightly, with a small increase in the effect for steels of higher carbon content. For rough calculation we use the approximation that 0.01 per cent vanadium raises the Ac, 2 to 4 F depending on

the carbon content.

With regard to the authors' statement that vanadium has a slight tendency to increase thermal sluggishness and air-hardenability, we feel that some confusion may exist. It has been shown by Grossmann and others that small amounts of vanadium, below approximately 0.07 per cent, have a very strong effect on the hardenability of carbon and low alloy steels. As the vanadium content is increased, vanadium-containing carbide particles are dissolved in the austenite at ordinary heat-treating temperatures with increasing difficulty. The presence of such undissolved carbides appears to hasten transformation, thus counteracting the effect of the vanadium in solution. With appreciable amounts of chromium and particularly molybdenum present to form carbide particles, and with relatively low austenitizing temperature, we feel that the effects of vanadium on transformation are relatively minor, but that the larger quantities would be found to lower hardenability if other elements were sufficiently alike to permit a careful comparison.

We got the wrong impression from the authors' statement that vanadium tends to spheroidize carbides, but the disagreement may be in the meaning of the word spheroidize. We know that vanadium is a strong carbide former and that vanadium-containing carbides are difficultly soluble, but that vanadium does not hasten the growth of small carbide particles into longer spheroids, which, to us at least, is implied by the authors' statement. On the other hand, we believe that the resistance to growth or coalescence of vanadium-containing carbides is the key to vanadium's effect in resisting softening at elevated temperatures.

We were quite surprised to hear vanadium credited with causing a coarser dendrite structure. Just the opposite has been observed in cast steels for many years. A beneficial effect has been observed in many types of alloy steels; tougher as-cast structures have resulted from the addition of 0.08 to 0.20 per cent vanadium. An analogous condition exists in the use of vanadium in ingots for large forgings where forgeability and transverse properties are definitely improved. We would like to see more evidence than is given in this paper to substantiate the authors' statement.

As to the authors' findings on vanadium's effect on tensile properties, we believe that the situation is somewhat similar to that regarding hardenability. In these relatively high alloy, low carbon steels the effect of vanadium on room temperature tensile properties is overshadowed by other factors.

We were impressed by the observation that as-cast and stressrelieved plates of these steels had tensile properties similar to those with normalized and tempered heat treatments, which illustrates their thermal insensitivity. We believe that this helps to explain the observed lack of response to vanadium additions with respect to room temperature properties.

¹ Metallurgical Engr., Vanadium Corp. of America, Bridgeville, Pa.

A. J. Kiesler: 9 Were these steels aluminum killed?

LS

ld.

ca.

on

ng

al-

12

of

Oy

ıt.

on

he

TS

y.

n

m

a

n-

n

a.

le gh h h s h i ts

s-nisster of deer

MR. GOLDSMITH: Yes, with 2 lb of aluminum addition.

Mr. Meinhart (Authors' Reply): The authors wish to thank Mr. Merrill for his interesting discussion and contribution which he has made to this paper.

Thermal sluggishness as revealed by the dilatometer curves definitely is an indication of hardenability. However, the rapid cooling rate of 50 C or 90 F per minute used in the dilatometer test is much slower than that of the water quench used in the Jominy hardenability tests. This may account for some of the differences between our results and those mentioned by Mr. Merrill. The data presented in this paper and that given in a previous paper on similar steels without chromium (Ref. 6) show that raising the vanadium content from 0 to about 0.3 per cent results in a small increase in thermal sluggishness. On the other hand, vanadium appears to have a more pronounced effect on increasing the hardness in the higher carbon groups. Carbon, chromium and molybdenum are all more effective in increasing the thermal sluggishness than vanadium.

The chromium and molybdenum levels of the steels used in these tests are becoming important in the casting industry, because of the good high temperature properties they confer. It is likewise well known that vanadium increases the creep resist-

ance of these steels at elevated temperatures.

The statement that vanadium tends to spheroidize carbides may be a little misleading. More fine spheroidal carbides have been observed in these steels as the vanadium content was increased. This was most evident in the low carbon group and at least partially resulted from closing of the gamma loop by

² Chief Foundry Met., General Electric Corp., Schenectady, N. Y.

the vanadium. The carbides of vanadium appear more stable than those of iron, chromium and molybdenum. Hence they resist going into solution during heat treatment and the remains of the original pearlite structure partially spheroidizes. The stability and sluggishness of these carbide results in a fine irregular dispersion of carbides in the pearlite during transformation instead of the true lamellar form usually observed in carbon or low-alloy steels.

The more pronounced dendritic pattern has been found in a number of samples containing between 0.2 and 0.3 per cent vanadium, however, no detrimental effect on toughness accompanied the appearance of this pattern in the heat treated steels (See Serial Nos. 13 and 16 of Table 1 in the paper). The grain size of the high vanadium steel is much finer than that of the vanadium free steel. Therefore, the dendrite pattern appears more pronounced in the former when comparing the two steels at the same magnification.

It is agreed that vanadium additions to these steels usually do not exceed 0.10 to 0.15 per cent. Nevertheless, the authors considered it worthwhile to cover the field thoroughly and obtain data both in this range as well as up to about 0.30 per cent vanadium, all of which is presented in Table 1.

With regard to Mr. Merrill's remarks about tensile properties of plates used for welding, it should be noted that all plates were subjected to a triple heat treatment before welding and stress relieved after welding. Therefore, properties of the parent metal away from the heat affected zone would be similar to those of the coupons subjected to the triple heat treatment. The properties of these steels are definitely improved by the triple heat treatment as compared with those obtained by stress relieving only.

CORE OIL EVALUATION METHOD

By

A. E. Murton,* H. H. Fairfield* and B. Richardson*

During a comparative investigation of 19 commercial core oils, and several prospective binder materials carried out in the sand laboratories of the Mines Branch at the request of the Steel Castings Institute of Canada, certain findings appeared to be of general interest. These features include:

(1) The addition of heating elements at the bottom of the laboratory core oven, and the installation of a positive adjustable draft system (late model core ovens have, as standard equipment, reflector plates which reflect heat toward bottom of cores).

(2) A method of assessing core oils by means of oaktree contour curves¹ for comparing the baking characteristics of two or more binder materials.

(3) The effect of humid storage on the strength of baked cores.

(4) The effects of high oven humidities on the baked strength of core oils.

Core Oven Alterations

In the testing of core oils certain inconsistencies inherent in such naturally occurring substances as sand and vegetable oils impose themselves upon a system of testing which already has room for variation in results. The strength of cores made using the same core oil with AFS 50/70 test sand and distilled water, mixed, rammed up and baked under closely controlled duplicate conditions² will vary from day to day for reasons which are not yet apparent.

This day-to-day variation was reduced to some extent by alterations to the core baking set-up at the Mines Branch. Three 1,000-watt domestic stove heating elements of the "calrod" type were coiled in a ½-in. deep well in the transite board which forms the base of the core oven. Two of these elements are manually controlled; the third is thermostatically controlled by a Dekhotinsky-type thermostat similar to the one which originally came with the oven. The new thermostat was mounted where it would be below the cores during the time of bake (the thermostat originally supplied is mounted just above the core baking region).

The lower thermostat and thermometer can be drawn out to allow passage of the core trays. This new arrangement ensures a close temperature control both above and below the cores throughout the baking cycle, and the total variation amounts to less than $\pm~2~\mathrm{F}$ as read on the oven thermometers.

In the original installation the 4-in. chimney pipe from the core oven was led out through a window, with the result that on certain windy days the core gases were blown back in the pipe and out the oven door. An exhaust fan consisting of a size "O" utility blower directly connected to a 1/20 hp, 1800 rpm, 60 cycle motor was placed in the smoke pipe. This fan provided a positive exhaust, but the wind still had some effect on the speed of bake, so a draft gage was installed, the leads from which go to either side of an orifice plate with a 1-in. diameter hole³ installed in the 4-in. chimney pipe. A damper in the pipe is regulated to give a pressure difference across the orifice of 0.25 in. of water, and this pressure is closely maintained at all times.

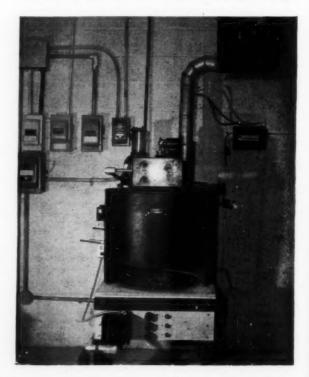
These changes may be seen in Fig. 1, which shows the core oven with the door removed so that the new heating coils may be seen. The thermostats and thermometers mounted above and below the baking zone in the oven may be seen on the left-hand side. The exhaust fan and draft gage are above and to the right of the oven.

Improvements in operation resulting from the oven modifications are:

(1) The oven comes to temperature much more quickly.

(2) Baking temperatures are maintained within ± 2F both above and below the baking cores.

Fig. 1—Core oven with door removed to show electric heating elements in bottom. The control thermostats are on the left-hand side of the oven; draft gage is on wall at right, and exhaust fan above draft gage.



^{*}Metallurgist, Physical Metallurgy Div., Mines Branch, Dept. of Mines and Technical Surveys, Ottawa, Ont., Canada; chief metallurgist, Wm. Kennedy & Sons Co. Ltd., Owen Sound, Ont., Canada; and research engineer, Steel Castings Institute of Canada, Ottawa, Ont., Canada, respectively.

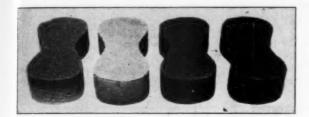


Fig. 2—Cores on left were baked with lower heating element off. Core on extreme left is top-side up, next core bottom-side up. Cores on right were baked same length of time at same temperature (baked 40 min at 450F) with upper and lower heaters on.

as

n

n

u

of

n.

e

18

ıt

n

(3) When a full load of cores is placed in the oven the temperature drops, but the baking temperature is recovered more quickly since the new heaters were installed in the bottom of the oven.

(4) The "band" of results for baking the same core mix at 400F on different days has been narrowed by about one half for this particular core oven.

(5) The cores are baked to the same hardness on the bottom and on the top.

A possible disadvantage in all-electric baking is the lack of combustion gases from the burning of fuel. The pressure of these gaseous products does have an effect on the baking process in oil- or gas-fired ovens so that the results of cores baked in the all-electric oven are comparative among themselves but cannot be compared directly with test cores baked in a fuel-fired oven.

Cores containing a bake-out dye were baked with the new bottom heating coils on and off. The two cores on the left in Fig. 2 were baked with the bottom elements turned off. The core on the extreme left is top side up and the one next to it is bottom side up. Shading from dark to light may be seen on the ends of these cores. The two cores on the right in Fig. 2 were rammed from the same mix but baked with both top and bottom elements turned on. All cores in this test were baked 40 min at 450F. The cores on the left crumbled on the bottom although they were hard on top. Those on the right were hard top and bottom.

The band of results for one core oil on the conventional core baking curves at the maximum strength for a 400F bake was approximately 42 psi wide previous to the core oven alterations, and is now about 18 psi wide at the same location for the same core oil.

Assessing Core Oils by Baked Strength Testing

In the evaluation of core oils on the basis of baked strength, the requirements to be considered for steel foundry cores are:

(1) Rapid strength attainment so that large cores will not require excessive baking time.

(2) Hold strength for a long time in order that small cores will not be burned out while the large cores are being baked through, and so that the large cores will not be overbaked on the outside before the inside is thoroughly cured.

(3) Not be too sensitive to ordinary variations in oven temperature.

(4) Bake up at reasonably low temperatures for fuel economy in the core oven.

(5) Économical in satisfying these requirements. In testing core oils a standard mixture is prepared. The mixture used in this investigation consisted of A.F.S. 50/70 test sand, 3000 grams; test core oil, 30 grams; distilled water, 90 grams.

The A.F.S. tensile test cores were rammed and baked. The relationship of baking time, temperature, and tensile strength is shown in Fig. 3. This curve shows that with core oil (A) a strength of 125 psi was reached in 280 min when baked at 300F; 67 min at 350F; 45 min at 400F; and 35 min at 450F.

Compilation of a graph such as shown in Fig. 3 requires a great deal of test work. Fortunately, the patterns of such curves for vegetable-base core oils have common characteristics, and after completing similar graphs for 20 commercial core oils certain trends became apparent for all of the oils tested:

 Oil-sand mixtures attain maximum strength more quickly when baked at high temperatures than at lower temperatures.

(2) Oil-sand mixtures retain maximum strength over a longer bake time interval at lower temperatures.

(3) The curve shown in Fig. 4 was taken from the

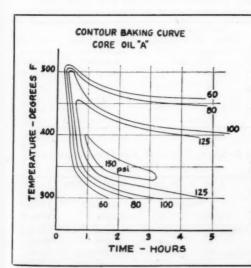
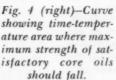
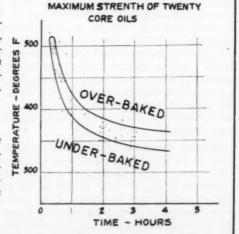


Fig. 3 (left)—Relationship of baking time, temperature, and tensile strength is shown in contour-type baking curve (core oil A).





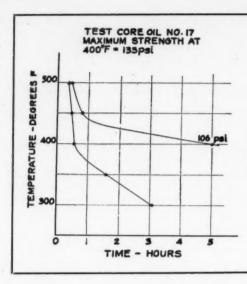
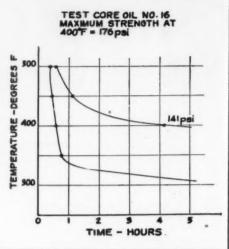


Fig. 5—Strength curves for two core oils based on 80 per cent of maximum strength reached during bake at 400 F.



results of complete baking tests for 20 core oils. Within the loop are plotted the time-temperature points for the attainment of maximum strength for the satisfactory core oils. Where the maximum strength of any oil at any temperature fell below the loop, those oils tended to underbake. When the maximum strengths fell above or to the right of the loop, those oils tended to overbake.

It is suggested that the evaluation of commercial core oils be carried out as follows:

(1) Bake cores at 400F for 30, 45, 60, 90 and 100 min using a standard mix.

(2) Record maximum baked strength.

(3) Plot the baked strength and record: (a) time to reach 80 per cent of maximum strength; (b) time when overbaked core drops below 80 per cent of maximum tensile strength.

(4) Those oils which have suitable characteristics can be rated in terms of cost of oil per ton of core sand mixed to give 200 psi maximum strength at 400F.

Figure 5 shows curves for two core oils based on 80 per cent of the maximum strength reached during a bake at 400F. It is immediately apparent that oil No. 16 bakes to a strength of 141 psi more rapidly than oil No. 17 bakes to a strength of 106 psi. Economics permitting, the strength of oil No. 17 could be increased by using more oil, with the possible consequent disadvantage of a larger gas evolution in the mold, but there is no guarantee that it would bake up any faster. Table 1 gives suggested rated values for core oils according to this scheme based on 1-in. A.F.S. tensile specimens standard-rammed from core mixes containing 1 per cent test oil and 3 per cent distilled water in A.F.S. 50/70 test sand, and baked in the oven previously described.

Arbitrary Standard: An alternative method of assessment is suggested where complete baking cycles are carried out at 300, 350, 400, 450 and 500F and results plotted as in Fig. 3. Core oil can be assessed on the basis of an arbitrary standard and for this method a fixed standard of 125 psi was selected. Figure 6 shows the 125 psi curves for the same two oils represented in Fig. 5. Oil No. 16 comes up to 125 psi tensile strength

in about 39 min, holds that strength for 4 hr with an allowable variation in baking temperature of minus 35F plus 30F. Oil No. 17 requires just about 1 hr 54 min to reach a strength of 125 psi, holds that strength for only 4 hr with, at times, no allowable variation in oven temperature. Oil No. 16 would bake up just about as well at 350F, whereas oil No. 17 would not.

Based on the 125-psi criterion and on test data taken from baked strength results on 19 commercial core oils, assessment standards are given in Table 2.

Effect of Humid Storage on the Strength of Baked Cores

One of the most significant tests in this series was the one concerned with the drop in strength of baked cores stored for 24 hr at various humidities from zero to 100 per cent. The results appear to be of immediate practical importance to foundries located in those parts of the country where the humidity hovers between 70 and 80 per cent or higher for many months of the year. While humidity has a temperature co-

TABLE 1—BAKED STRENGTHS FOR CORE OILS RATED ON 80% OF MAX. STRENGTH AT 400F.

Oil	Time at 80 Max. 400 F Strength for Bake, m	Baked 375 F	Lowest Temp. of Bake for Standard	Allowable Variation in Oven	
Rating	From	To	Strength, F	Temp., F	
Above Avg.	45	300	350	60	
Avg.	60	300	350	40-60	
Below Avg.	. 75	300	365	20-40	
Poor	75 plus	270	365 plus	20 or less	

Table 2—Baked Strength for Core Oils Rated on Basis of 125 psi at 375F.

Oil	Time at 1 Baked at min	375 F,	Lowest Baking Temp., for Standard	Variation in Oven
Rating	From	To	Strength, F	Temp., F
(A) Above Avg.	45	300	350	60 plus
(B) Avg.	60	300	350	40-50
(C) Below Avg.	60	270	365 plus	20-40
(D) Poor	60 plus	270	365 plus	20 or less

efficient, the same 70 grains of moisture per pound of dry air which give about 90 per cent relative humidity at 60F will still give the air at 70F a humidity of 65 per cent. Fairly high indoor humidity conditions usually persist in Eastern Canada from the time of the spring breakup right through summer and up until the first cool days of fall.

The annual percentage relative humidity taken at 8:00 a.m. for many years in some United States cities having climates somewhat similar to cities and towns in Canada are: Bismarck, N. D., 81; Burlington, Vt., 77; Buffalo, N. Y., 78; Chicago, Ill., 76; Milwaukee, Wis., 78; Port Huron, Mich., 81.

It can be assumed that the 75 per cent humidity test will be closest to that encountered in foundry core storage, and some of the test cores were seriously weakened when stored at that humidity. One hundred per cent humidity lasting 48 hr will be encountered occasionally in Eastern Canada during a warm rainy spell

Method of Testing: The cores were standard rammed A.F.S. tensile specimens baked for 90 min at 400F. The core mix was made up as follows: 1500 grams A.F.S. 50/70 test sand; 15 grams test oil; 45 grams distilled water.

an

us

54

th

on

ist

en

re

as

ed

ro

ite

se

oe-

hs

0-

ble

Sixteen cores were rammed up for each test oil, baked together and stored at room temperature for 24 hr at different humidities. Four of the oil cores were stored in dry air while the remaining 12 were stored—one lot of four cores at 50 per cent relative humidity, one lot of four at 75 per cent relative humidity, and one lot at 100 per cent humidity. Table 3 gives the results of these tests.

Further work should be directed toward finding an ingredient which, while not harmful to the desirable qualities, might lessen the tendency of the cores to weaken in storage. In the meantime dry storage of cores could be considered. In this connection 25 per cent relative humidity would probably be a safe level. Also, most of the weakening of the cores seems to take place within the first 24 hr in storage.

Cores which have been weakened by storage at high humidities do not recover their strength by further

TABLE 3—BAKED TENSILE STRENGTH FOR CORES STORED 24 HOURS AT ROOM TEMPERATURE AT RELATIVE HUMIDITIES AS SHOWN.

		Baked Strength, psi								
Oil No.	"Dry" Storage	50% R.H. Storage	75% R.H. Storage	R.H. Storage						
1	166	150	135	82						
2	136	134	129	91						
3	200	175	146	84						
4	151	138	124	88						
5	146	146	118	82						
6	190	184	174	131						
7	194	174	159	119						
8	165	156	144	106						
9	196	180	160	137						
10	155	149	126	97						
11	118	. 109	103	75						
12	124	105	74	45						
13	136	122	91	55						
14	157	151	133	86						
15	15.5	12.7	11	-						
16	185	174	159	110						
17	140	122	88	62						
18	207	188	156	111						
19	137	122	91	51						
Avg.	154	142	122	85						

storage at low humidities. The strength can, however, be recovered by redrying the cores in an oven at about 220F.4

Variations in humidity have long been blamed for variations in the baked strengths of test cores. The effect of this factor was studied by baking 20 core oils with high oven humidities at 400F and comparing the results with those obtained on the same oils at normal humidities.

To make this test the core oven previously described was fitted with an evaporating pan placed over the bottom heating coils, as shown in Fig. 7. Sufficient water was dripped into the pan at a constant rate through the needle valve-funnel arrangement to give a dew point in the oven of about 120F. The full line in Fig. 8 shows the average of 29 test bakes of 20 core oils, 19 of which were baked once each, while the twentieth was baked 10 times at usual humidities.

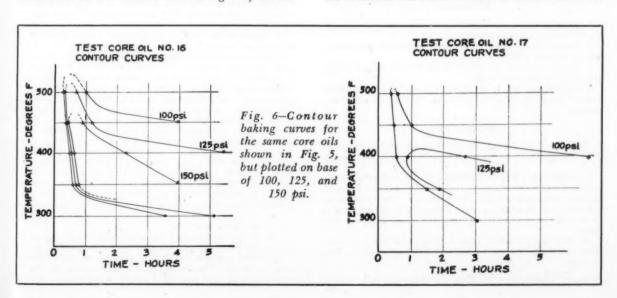


Fig. 7-Below-The core oven was fitted with an evaporating pan placed over the bottom heating coils for humid baking of test cores. Right-Needle valve and funnel arranged to control flow of water to the evaporating pan.





The dotted line is the average for 29 exactly similar bakes except at the high humidity. All of the oils were found to be affected when results for the individual core oils were compared. The two sets of averages from this large number of samples leave little doubt as to the general effect of oven humidity. It will be seen that the moisture level in the core oven does not affect the strength of oil in sand cores but does affect the time of bake.

Theoretically, such a result is to be expected when it is realized that a large part of the baking time for cores is consumed in evaporating off the water which the cores contain, and that once this water has been removed, the oil in the cores bakes up very rapidly. It is, then, only to be expected that a high partial pressure of water vapor in the oven atmosphere would extend that portion of the baking time devoted to the removal of water from the cores. The results of this study tend to confirm the theory.

Further work is being directed toward finding the relationship which would seem to exist between oven humidity and the displacement of the humid bake curve shown in Fig. 8.

References

1. Harry W. Dietert, FOUNDRY CORE PRACTICE, American Foundrymen's Society, Chicago (1950).

2. FOUNDRY SAND TESTING HANDBOOK, American Foundrymen's Society, Chicago (1944).

3. Ovid W. Eshbach, Handbook of Engineering Fundamentals, John Wiley & Sons, New York.

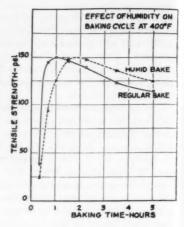
4. Unpublished results of continuing studies at the Mines Branch, Dept. of Mines & Technical Surveys, Ottawa, Ont., Canada.

DISCUSSION

Chairman: E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland. Co-Chairman: H. K. SALZBERG, The Borden Co., Bainbridge, N. Y.

F. S. Brewster (Written Discussion): ¹ The authors are to be congratulated for attacking this very pressing problem of standardizing baked core strength of various oils. The variables are numerous and sometimes elusive. Every step in the preparation of the test core and its final baking has a bearing on the final result.

We have today standard sand, a reference oil and a standard laboratory oven. These are a start but still leave many variables at large. Fig. 8—Curves showing effect of high oven humidity on baking speed at 400F. The curves are averages of 29 bakes. Maximum strengths are not affected.



The authors' work shows clearly how variations in the installation of the core oven can change results. In a process so dependent on controlled air circulation, exposure to changing atmospheric conditions is bound to upset the process.

The exhaust stack of the oven should contain an atmospheric vent or draft breaker and should extend vertically 5 ft above any nearby roof tops. Where this is impossible, a small exhaust fan can be used—always including the atmospheric vent.

It is necessary that the oven be equipped with a bright reflector plate below the core plate level to equalize heating rate. The omission of this plate would cause the differences noted by the authors between top and bottom of the specimens.

The following table shows the range of hardness one can expect from the oven unless properly installed.

Table 1-'Hardness' of Cores Baked at 425 F for 75 Min.

Core No.	Rotary Dry Hardness			
	Тор	Botton		
1	69	69		
2	67	70		
3	66	66		
4	68	68		
5	68 66	67		

We would expect to find that any slight variation in preparation of the cores to show up prominently in the early stages of the bake. Figure 8 shows how fast strength is changing in the 30 to 60-min range. Any slight difference in the rate of moisture removal or rate of temperature rise is magnified in this stage of the bake.

In regard to the effect of humidity on tensile strength, we must be very careful in the interpretation of results. The A.F.S. Subcommittee on Core Strength is now engaged in studying this point and finds it somewhat complicated. In the first place, it is apparent that we should use absolute humidity rather than relative figures. It is the weight of water per cubic foot of air entering the oven that determines the humidity of the oven air. Secondly, the relationship does not seem to be a continuous curve. There are high and low points and flat stretches. Thus, in one range of humidity, tensile strength may increase with humidity, in another it may decrease and in another there may be little or no change. The hope of the Committee is to collect reliable data to the point where tensile results can be corrected to some chosen point on the humidity curve, providing the absolute humidity is known at the time of baking. In one location in Michigan it was observed that during a year's time, absolute humidity ranged from 2 to 12 grains per cubic foot of dry air. The average for the year cannot be applied to any one test made during that time.

The authors have made a fine contribution to the evaluation of core oils by their contour curve method of plotting results. It is a clear and practical way of showing the useful baking range of an oil. The contour curve should repiace the family type of curves in the interests of simplicity and clarity.

¹ General Manager, Harry W. Dietert Co., Detroit.

CHAIRMAN ZIRZOW: I want to call your attention to the fact that these tests were all run on straight oil-sand cores. There was no cereal in these cores. If we introduce cereal we introduce other variables and that is why we left that out.

I agree with Mr. Brewster that there are peaks and valleys in the humidity. I do not blame that on humidity but on the failure to determine humidity in the oven.

MEMBER: I was impressed by the method of setting up these contour curves. Chairman Zirzow mentioned that these tests were run without cereal in the sand mixture. In 90 per cent of the sand mixtures used in the foundry, corn flour is used and it has an effect on the oil. Therefore I believe it is necessary to run the test with the amount of cereal that is going to be used in production in the foundry.

For many years I have recognized that moisture or humidity is an important factor in storage of core oils. You can determine the loss in strength of a core oil bond between two different oils due to storage. When cereal is added to the sand mixture that condition may be reversed. An indication of that is the green strength you will get from your mixes.

In determining the baking range of core oils the most important range is the range past the peak strength. Oil that will reach its peak strength early and hold its peak strength on a long bake is more important than one which will come close to its peak strength early but not reach its peak strength.

MR. FAIRFIELD: When checking the contour curves (Fig. 4) of some 24 different oils, we found that the time they reached their maximum strength was practically identical for all of them.

At 500 F, for example, all of the available oils come to maximum strength within a matter of a few minutes of each other. At 400 F all of the oils came to their maximum strength within ± 10-min period. Of course, the important difference is that some of them retain their strength for longer periods than others.

MR. RICHARDSON (Authors' Closure): The authors are particularly pleased to find such a large area of agreement between these comments and our own findings, all the more so when the wealth of practical experience of Mr. Brewster and others is considered. We especially appreciate this discussion where it suggests a future line of investigation such as adding cereal binders to the oil cores.

With regard to the oven installation; what we were after, fan or no fan, gauge or no gauge, vent or no vent, was a more consistent bake day-to-day and season-to-season of the standard ingredients mixed and molded according to standard A.F.S. procedures. We achieved our best results with the arrangement described. Others might do better with a different arrangement. That depends on variables which have not yet been isolated or even recognized. As to the reflector plate, we will have to try one out and see how it compares with our bottom heaters.

In connection with the humid bake we agree that fundamental units are useful. We measured the dew point and that is a fundamental unit. Relative humidities are equally fundamental when the temperature is specified since grains of water per pound of dry air may be read directly off of the psychrometric chart. Our results were given for one high oven humidity and we would like to see something similar for varying oven humidities related to time and temperature of bake.

IMPROVEMENT OF MACHINABILITY IN HIGH - PHOSPHORUS GRAY CAST IRON - PART II

By W. W. Austin, Jr.*

ABSTRACT

This paper reviews additional results of an investigation carried out at Southern Research Institute under the sponsorship of The Woodward Iron Company of Birmingham, Alabama. Foundry scale applications of the previously described desulphurization-zirconium addition treatment for machinability improvement in high-phosphorus gray iron are described, and performance data including cost estimates and results of tool

life machinability tests are presented.

The results of this work indicate that significant improvement in machinability may be accomplished through application of the proposed treatment under foundry operating condi-tions. While best results were obtained by desulphurization with calcium carbide in the basic-lined cupola followed by ladle addition of zirconium ferrosilicon, published results indicate that satisfactory sulphur elimination may also be obtained with caustic soda employing a special reladling technique. The total cost of the proposed zirconium treatment utilizing caustic soda desulphurization is estimated from current prices at \$2.75 per ton of treated iron.

Introduction and Summary

IN A PREVIOUS PAPER1 the results of laboratory scale experiments on improvement of machinability in high-phosphorus gray iron were set forth. It was found in this preliminary investigation that significant improvements in machinability could be effected through a combined desulphurization-zirconium addition treatment. More specifically, this treatment, found to be most effective on high-phosphorus irons of eutectic or slightly hypereutectic composition, consisted essentially of desulphurization to 0.02 per cent S maximum, followed by the ladle addition of approximately 0.12 to 0.15 per cent zirconium in the form of 13 per cent zirconium-ferrosilicon. Increased machinability ratings on the order of 40 to 45 per cent were obtained employing a simple drill penetration test.

Since the publication of this initial paper,1 the scope of the investigation has been expanded to include a number of foundry scale tests. Moreover, studies of additional experimental methods and more

comprehensive procedures for evaluating machinability have been undertaken. It will be the purpose of the present paper to summarize the results of this expanded investigation and to indicate how these findings may be most effectively applied to commercial foundry operations.

In general, it has been found that foundry scale desulphurization to the extent required for optimum machinability improvement (i.e., 0.015 per cent to 0.020 per cent S) is not easy. There are, nevertheless, commercially feasible methods of achieving sulphur contents in this range. A proprietary reladling technique employing caustic soda,2 the introduction of powdered calcium carbide into the molten iron,3 and the basic-lined cupola with calcium carbide in the charge4 are cited as examples of suitable commercial desulphurizing procedures.

Following desulphurization the ladle addition of 0.12 to 0.15 per cent of zirconium in the form of 13 per cent zirconium-ferrosilicon has been confirmed as a satisfactory method of controlling microstructure to give the desired improvement in machinability. Since this addition requires metal temperatures in the neighborhood of 2600 to 2800 F for effective alloy dissolution, it is important that the highest practicable spout temperature be maintained. In this connection the basic-lined cupola employing lump calcium carbide in the charge has been shown capable of producing low-sulphur iron with spot temperatures in excess of 2900 F.4 This naturally makes an ideal melting practice for desulphurizing and zirconium alloy addition.

Finally, it has been shown that where thorough desulphurization is impractical, the ladle addition of 13 per cent zirconium-ferrosilicon alone is sufficient to produce appreciable machinability improvement in normal gray iron analyses. This application is now in commercial operation in at least one production foundry in the Birmingham district.

Experimental Methods and Discussion of Results

Experimental work has been carried out on three separate but related phases of the investigation. These may be classified as additional laboratory studies,

beer

atio

CO

fu

lb

th

(1

ve

21

^{*} Senior Metallurgist, Southern Research Institute, Birmingham, Alabama.

² Small superior numerals refer to bibliography at end of paper.

foundry scale operations, and machinability evalua-

More than 150 additional laboratory heats have been prepared for the purpose of studying alloying agents and desulphurization methods. This work consisted of induction furnace melting of selected base iron charges employing standardized melting and casting procedures in order to eliminate uncontrolled variables insofar as possible. As previously indicated, the base iron employed conformed except for its phosphorus content (0.60 to 0.65 per cent P) to specification No. 110 "Automotive Soft Iron" of the Society of Automotive Engineers (Table 1).

TABLE 1-AUTOMOTIVE SOFT IRON S.A.E. Specification No. 110⁶

	% Min	% Max
Total Carbon	3.40	3.70
Silicon	2.30	2.80
Manganese	0.50	0.80
Sulphur		0.12
Phosphorus		0.25

For melting of laboratory heats a 20-kw frequency converter equipped with 12-lb and 30-lb induction furnaces, and a 50-kw converter with 100-lb and 250-lb furnaces were used. Test specimens consisted of the keyhole-type chill specimen described previously (Fig. 1), the standard 1.20-in. dia. x 22-in. long transverse bar, and cylindrical specimens (3.75-in. o.p. x 2.5-in. i.d. x 13-in. long) for tool life machinability tests. Keyhole chill specimens and the 1.20-in. transverse bars were cast in dry sand while the cylindrical specimens were cast in green sand.

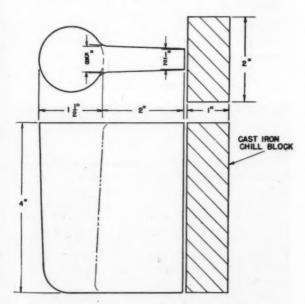


Fig. 1-Keyhole-type chill test specimen.

The drill penetration test described previously¹ has been used throughout this investigation for the evaluation of machinability. In the later phases of the

work this test has been supplemented by tool life tests as described by Boston.¹¹

Desulphurization Methods

In view of the low sulphur content required for optimum effectiveness of zirconium additions, an extensive study was made of desulphurization methods. Preliminary work indicated that adequate desulphurization could be achieved with large (2 to 3 per cent) additions of soda ash, or through a two-stage soda-ash treatment. However, due to excessive refractory costs and handling difficulties involved with foundry application of such treatments, it was desirable to devise better desulphurizing techniques. With this objective in mind, laboratory studies were made of metallic magnesium additions and certain synthetic basic slags. Foundry scale experiments on desulphurizing with calcium carbide and basic refractories were also conducted, with some interesting results.

Through current developments in the production of nodular graphite cast iron⁶ it is well known that small additions of magnesium to gray iron will exert a pronounced desulphurizing action. Consequently, magnesium desulphurization was investigated in connection with zirconium additions for machinability improvement.

While it was found that the addition of 0.20 to 0.30 per cent magnesium would reduce sulphur contents from 0.08 per cent S to as low as 0.035 per cent S, it was also observed that the magnesium addition was accompanied by a decided chilling effect. This resulted in substantially increased hardness and adversely affected the machinability of the treated specimens.

The addition of 1.0 per cent zirconium-ferrosilicon overcame the chilling effect produced by magnesium and produced a structure substantially equivalent to that of the untreated base iron. While these findings confirm the desulphurizing power of magnesium and the softening effect of zirconium-ferrosilicon additions to low-sulphur irons, they unfortunately also reveal the highly detrimental chilling effect frequently associated with magnesium additions. In view of these facts, the use of magnesium as a desulphurizing agent in connection with treatments for improving machinability does not appear to be promising.

Experimental results of a more promising nature were obtained through desulphurization of laboratory scale heats with a highly basic synthetic slag. For this purpose a mixture containing 48 per cent quicklime, 47 per cent fluorspar, and 5 per cent cryolite was added to a high-sulphur iron and held 15 min at 2500 to 2700 F in the induction furnace. It was found that a single addition of 3.0 per cent of this slag would reduce the sulphur content of a 0.11 per cent S iron by approximately 50 per cent, while a two stage treatment using 5.0 per cent additions would reduce the sulphur content of a 0.08 per cent S iron to 0.010 per cent S.

Although additional work on this subject was suspended when highly favorable foundry scale results of desulphurization with basic refractories and calcium carbide were made available,4 it is believed that a desulphurization treatment comparing favorably with the conventional soda-ash treatment may be devised using basic raw materials of this type.

Alloy Additions for Improvement of Machinability

Aside from the investigation of desulphurization methods, a parallel study was made of a number of alloy addition agents for use in improvement of machinability by direct addition without resorting to desulphurization. In this phase of the work, a selected group of alloy materials was examined systematically for evidence of beneficial effects upon the machinability of high-phosphorus gray iron. Since the production of optimum machinability in as-cast gray iron depends largely on the decomposition of pearlitic iron carbide to free ferrite and graphite, a majority of the addition agents were selected because of their recognized abilities to favor carbide decomposition. Among the materials studied were: calcium-silicon alloy, calcium-magnesium-silicon alloy, low-carbon and highcarbon ferrotitanium, copper-titanium alloy, ferroboron, cerium, bismuth, and copper-bismuth alloys.

Of all the addition agents investigated, the most potent in influencing the structure and machinability of the high-phosphorus base iron was bismuth. Although insufficient data are available at the present time to explain all of the somewhat anomalous effects of bismuth additions, several specific trends have been observed which are of definite interest in the study of

machinability improvement.

The use of bismuth as an addition agent to cast iron was studied by Smith and Aufderhaar⁷ who reported that it promoted graphitization of combined carbon in heavy sections, thus lowering the strength and hardness. While no specific effects upon machinability were described, graphitization and softening indicated likelihood of improved machinability.

In a series of laboratory heats it was found that the

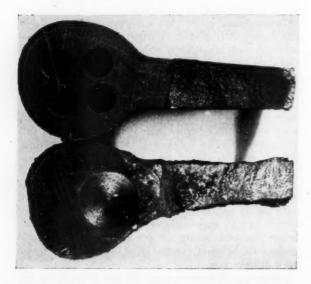


Fig. 2—Chilling effect resulting from a 0.10% Bi addition. Top: Untreated base iron; Bottom: Same iron plus 0.10% Bi.

addition of as little as 0.01 per cent bismuth caused a marked change in the structure and properties of the iron, while the addition of 0.10 per cent bismuth resulted in improvement of the drill penetration rating on the order of 85 per cent over that of the as-cast base iron. Indeed, except for fully-annealed specimens, drill penetration ratings on bismuth-treated heats were the highest observed in the entire investigation. Unfortunately, a pronounced increase in chilling tendency was also observed (Fig. 2). Moreover, the degree of chilling was shown to be closely related to pouring temperatures (Fig. 3), a factor which makes accurate control of results even more difficult.

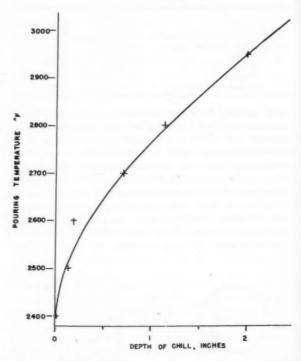


Fig. 3-Chill depth vs pouring temperature in bismuth-treated (0.01% Bi) iron.

Another adverse effect accompanying bismuth additions was the occurrence of a banded structure at the surface of the casting. While the microstructure of the interior of a $1\frac{1}{2}$ -in. section was largely ferritic, a condition characterized by the formation of a pearlitic band or zone approximately $\frac{1}{4}$ in. in depth occurred at the surface of the casting (Fig. 4, 5, 6). This pearlitic zone is substantially harder than the interior structure and actually possesses poorer machinability than the untreated base iron.

Since simultaneous additions of bismuth with numerous other alloys, inoculants, and prior desulphurization were all of negligible effect in overcoming these adverse conditions, bismuth must be regarded at this time as a rather unpredictable agent whose harmful influences more than offset the improvements associated with its use.

loc

g

i

d

i

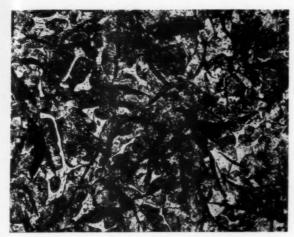


Fig. 4-Microstructure of untreated high-phosphorus base iron showing normal Type A graphite in pearlitic matrix. Mag. 200×.



Fig. 5-Microstructure at center of 11/2 in, section of high-phosphorus base iron treated with 0.10% Bi showing predominantly ferritic matrix. Mag. 200×.

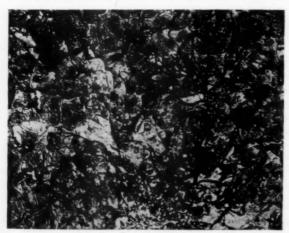


Fig. 6—High-phosphorus base iron treated with 0.10% Bi. Microstructure 1/8 in. below surface of a 11/2-in. section showing almost entirely pearlitic matrix with localized areas of fine Type B graphite. Mag. 200×.

Foundry Tests

In addition to the laboratory scale experiments, 22 cupola heats have been made for the purpose of evaluating and improving the proposed desulphurization-zirconium addition treatment described previously. Most of this work was done in 20 to 24-in. diam cupolas, although several tests were conducted employing 1000-lb ladles of iron from a 42-in. diam production cupola. In the initial phases of the foundry work, a study was made of the possibilities of improving the effectiveness of the conventional soda-ash treatment for desulphurization of cupola iron. One approach to this was a study of the possibilities of devising a continuous desulphurization process employing the counter current flow principle to soda-ash desulphurization.

Counter Current Desulphurization

Through the generous cooperation of Jackson Industries, Inc., Birmingham, Alabama, a refractory-lined trough or forehearth was installed on a 22-in. diam experimental cupola, as indicated in Fig. 7.

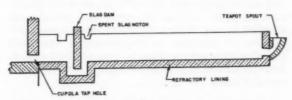


Fig. 7-Modified forehearth for counter-current desulphurization.

This trough was lined with a proprietary refractory known to give satisfactory results with soda-ash desulphurization. After lining, the inside dimensions of the trough were 6 in. by 6 in. by 60 in. A slag dam prevented cupola slag from entering the desulphurizing section, and at the lower, or discharge end, a teapot spout was provided for continuous tapping. Just below the slag dam a spent soda slag notch was provided. In order to minimize heat losses, the entire length of the trough was covered with insulating fire brick.

In operation, the preheated trough was allowed to fill with hot metal while simultaneous additions of soda ash briquettes were made, creating a layer of molten soda slag on the surface of the metal. As iron flowed through the trough at the normal melting rate of the cupola, predetermined additions of fresh soda ash were made at the discharge end, thereby raising the soda slag level until molten soda ash overflowed at the spent slag notch. Thus the purest soda ash was exposed to iron of lowest sulphur content in accord with the counter current principle. Using 30 to 35 lb soda ash per ton of molten metal, maximum desulphurization occurred after 20 to 30 minutes operation.

Although the elimination of approximately 50 per cent (0.062 per cent S reduced to 0.030 per cent S, or 0.046 per cent S reduced to 0.022 per cent S) of the sulphur content of cupola iron was achieved with this device, it is not considered sufficiently efficient for use

in connection with zirconium additions for improved machinability. Chief among the difficulties encountered was excessive heat loss. In addition to normal radiation losses from the relatively large surface area of the molten metal, the introduction of cold soda ash caused a substantial drop in temperature. Consequently, temperatures inside the trough were too low for optimum fluidity of the soda slag and much too low for effective alloy dissolution in the ladle. The process is, nevertheless, sound in principle, and with a reasonable amount of additional effort, it is believed that greatly improved operation could be achieved.

Desulphurization with Calcium Carbide

In the evaluation of various desulphurization techniques, attention was given to the use of powdered calcium carbide as a desulphurizer. For this purpose application was made, through the cooperation of the American Cast Iron Pipe Company and the Linde Air Products Company, of a novel addition technique. A dispensing unit was provided which, utilizing the principle of air-operated sand-blasting equipment, would inject a stream of powdered calcium carbide and dry nitrogen into the ladle of molten iron. Intimate contact between the metal and carbide was afforded by the violent agitation accompanying the injection of nitrogen; in this manner it was found that desulphurization could be accomplished very effectively. Table 2 summarizes typical data from this series of heats.

TABLE 2-LADLE DESULPHURIZATION WITH CALCIUM CARBIDE

Specimen No.			% Mn	% % S P		Remarks		
C-17-0	3.64	2.48	0.50	0.060	0.69	Untreated base iron		
1	3.66	2.52	0.50	0.018	0.69	Desulphurized*		
C-16-1	3.34	2.23	0.61	0.016		Desulphurized*		
C-18-0	3.63	2.20	0.71	0.080	0.64	Untreated base iron		
1				0.014		Desulphurized ^e		
2	3.64	2.72	0.69	0.010	0.64	Desulphurized plus 1.25% Zr-Fe-S		

52 lb calcium carbide per ton metal injected during 6 min. (b) 33 lb calcium carbide per ton metal injected during 8 min.

(c) 46 lb calcium carbide per ton metal injected during 6 min. (d) 1.25% addition of 13% zirconium ferrosilicon gave 0.097%

Zr residual.

When applied to 2000-lb ladles of molten cupola iron, a temperature drop of approximately 100 F resulted from the 6 to 8-min treatment. This is not considered a serious drawback in view of the almost phenomenal sulphur removal obtained. A distinct advantage of the treatment over conventional soda ash desulphurization is the complete freedom from erosion of ladle refractories. While this is a major problem with soda ash which forms a fluid, highly corrosive slag, the slag produced by calcium carbide is a solid granular dross-like material that shows no tendency to wet or attack ladle refractories. After the injection of carbide this slag rises to the surface where it may be easily skimmed off.

Although material costs for calcium carbide desulphurization are appreciably higher than for conventional soda ash desulphurization (Table 5), the ad-

vantages of more effective sulphur removal and freedom from refractory erosion tend to offset this difference in costs.

This series of experiments on calcium carbide desulphurization was arranged for the purpose of making a production scale check of the proposed desulphurization-zirconium addition treatment for machinability improvement. Accordingly, immediately after desulphurization, zirconium additions were made and specimens were cast for machinability evaluation. These specimens were identical to those previously used for laboratory heats so that no unknown variables of section thickness or cooling rates would be interposed to affect the results. Results of drill penetration machinability tests compared favorably with previous results from laboratory heats as indicated in Table 3.

TABLE 3-COMPARISON OF DRILL PENETRATION MACHINABILITY BETWEEN FOUNDRY SCALE AND LABORATORY SCALE HEATS

				Analy	sis, 9	6			
Specim		-		-	_	-	_		
No.	T.C.	Si	Mn	S	P	Zr	Remar	KS	
91	3.38	2.14	0.60	0.057	0.61		Untreated	base	iron
99	3.61	2.09	0.57	0.011	0.63	0.105	Treated*		
C-18-0	3.63	2.20	0.71	0.080	0.64		Untreated	base	iron
C-18-2	3.64	2.72	0.69	0.010	0.64	0.097	Treated*		
				Prop	erties	1			
Speci	men								
No.		Hardness ⁴		Depth Chill, in.		hill, in. D	T.M.	•	
9)1		B-89		0.19)	51.2	
9	99 B-84		0.09		9	77.0			
C-18-0		B-90		0.09)	55.2		
(C-18-2			B-80	0.015		15	71.3	

- (a) Specimen No. 91, 99 from 30-lb lab scale heats. Specimen No. C-18-0, -2 from foundry scale heat, i.e. 2000lb ladle from 42-in. cupola.
- (b) Desulphurized (two stages, 3% soda ash) plus 1.0% Zr-Fe-Si addition in induction furnace.
- (c) Desulphurized (calcium carbide injection, 46 lb per ton) plus 1.25% Zr-Fe-Si in 1000-lb ladle.

 (d) Rockwell "B" scale, ½6-in. diam steel ball 100-kg load.
- (e) Drill test machinability, mm penetration per minute.

It is evident from the results shown that the proposed desulphurization-zirconium addition treatment will effect substantial improvements in the machinability of high-phosphorus gray iron whether it is applied to small laboratory induction furnace heats, or to commercial foundry operations.

Desulphurization with Basic-Lined Cupola

During, and just after, World War II British investigators8, 9 published results indicating that satisfactory desulphurization of cupola iron could be accomplished through the use of basic refractory linings in the cupola. More recently Carter has confirmed these findings giving important additional information on the simultaneous use of calcium carbide as a part of the charge in the basic-lined cupola.

Carter's results on basic cupola operation show that sulphur contents as low as 0.031 per cent may be obtained from a 0.119 per cent S all-scrap charge using 3 per cent calcium carbide, or that sulphur contents

m

be

y e

28

0

1)

ıt

n-

38

d

a-

a

ıg

TABLE 4-BASIC-LINED CUPOLA HEATS

Specimen No.	T.C.	% Si	% Mn	% S	% P	Zr2	Hardness°	D.T.M.4	Remarks
C-19-7	3.54	3.25	0.58	0.017	0.71	****	B-82	52.7	Untreated base iron
C-19-8						0.022	B-75	60.5	0.5% addn. Zr-Fe-Si ^a
C-19-9	3.51	3.56	0.58	0.015	0.68	0.107	B-73	75.0	1.0% addn. Zr-Fe-Si
C-19-10			0.58			0.119	B-70	84.0	1.5% addn. Zr-Fe-Si
C-22-5	3.87	1.56	0.61	0.011	0.55		B-88	53.9	Untreated base iron
C-22-6	3.81	1.78		0.011		0.051	B-84	65.2	0.5% addn. Zr-Fe-Si
C-22-7	3.82	1.83	0.59	0.011	0.54	0.048	B-82	68.3	1.0% addn. Zr-Fe-Si
C-22-8	3.78	2.15		0.010		0.122	B-82	92.6	1.5% addn. Zr-Fe-Si
(a) Residua			zirconium	ferrosilicon.					n steel ball, 100-kg load. benetration per minute.

as low as 0.009 per cent S may be obtained from an all-pig charge using 6 per cent calcium carbide in the charge. Moreover, substantial reductions in coke ratios and increases in melting temperatures are made possible through the use of calcium carbide. Spout temperatures as high as 2935 to 2950 F were observed by Carter with 3 per cent carbide in the charge and a normal coke ratio (6:1), while temperatures ranging between 2770 F and 2870 F were obtained with 7 per cent carbide in the charge and a coke ratio of 25:1.

In view of these findings, it was felt that this process would be ideal for carrying out the desulphurization-zirconium addition treatment of high-phosphorus gray iron. Consequently, arrangements were made, through the generous cooperation of the American Cast Iron Pipe Co. to conduct two experimental foundry heats for desulphurizing in the basic-lined cupola with calcium carbide in the charge, to which suitable ladle additions of zirconium-ferrosilicon would be made. In these heats, high-phosphorus (0.70 per cent F) pig charges were melted with 5 to 6 per cent calcium carbide at a normal (7:1) coke ratio. The usual specimens were cast for evaluation of machinability and other properties. Table 4 summarizes the essential data from these heats.

In the first heat (Specimen No. C-19-) unusually high silicon contents were encountered. This was corrected in the second heat (Specimen No. C-22-) by reducing the silicon charged and using 10 per cent steel in the charge. The high carbons in this heat are the result of substantial carbon pickup which accompanies the use of calcium carbide, provided the silicon content is not excessive.

Excellent desulphurization was attained in both heats and the zirconium additions produced satisfactory improvements in the drill penetration rating as indicated. In fact the degree of improvement in machinability (60 to 75 per cent) is actually better than that obtained (45 to 55 per cent) in laboratory scale heats described previously. A more elaborate evaluation of the machinability of these and other specimens was conducted at the University of Michigan employing a tool life machinability test. Results of these tests are given in a subsequent section.

Cost Data on Desulphurization-Zirconium Addition Treatment

In the original paper¹ it was pointed out that maximum machinability of high-phosphorus gray iron may be achieved through annealing. Although annealing

is frequently employed to assure uniform machining qualities in certain types of difficult machinable castings, investment and operating costs for production annealing furnaces are comparatively high. Consequently, it is not feasible for the average gray iron foundry to resort to this means of improving machinability except in extreme cases.

Actual annealing costs range from an estimated \$10.00 to \$15.00 per ton of castings under ideal conditions to as much as \$50.00 to \$75.00 per ton when annealing is contracted in small job lots. It should be pointed out that annealing costs of \$10.00 to \$15.00 per ton of finished castings actually amounts to \$7.00 to \$10.50 per ton of molten iron, based on a casting yield of 70 per cent. It has been the goal of this investigation to develop a satisfactory treatment for machinability improvement that would compare favorably with the \$10.00 to \$15.00 per ton figure for annealing under ideal conditions. The degree of attainment of this goal may be judged from the cost estimates presented in Table 5.

At first glance it would appear that soda ash desulphurization is considerably better than either of the calcium carbide treatments. Unfortunately, attempts to verify this conclusion in foundry scale operations have not been successful. In the first place, laboratory scale experiments indicate that a two-stage soda-ash treatment with 3.0 per cent soda ash additions and 10-min holding periods in each stage are required for the necessary degree of desulphurization. Although such a treatment may be conveniently car-

Table 5—Cost Data on Desulphurization-Zirconium Addition Treatment

Desulphurization	Costs Per Ton Treated Iron						
Method	Desulphurizer	Refractories	Zr Alloy	Total			
Soda ash*	\$2.72	\$0.10	\$1.65	\$4.47			
Calcium carbide ^b	3.26		1.65	4.91			
Basic cupola ^e	4.80 3.00		1.65	9.45			
Caustic soda reladling	d 1.00	0.10	1.65	2.75			
(a) Two stams domain	buniantian mi	sh and anda	b /60	11			

(a) Two stage desulphurization with 3% soda ash (60 lb per ton). Soda ash briquettes at \$52.75 per ton.

(b) Powdered calcium carbide injection. Desulphurizer costs based on 2.3% calcium carbide (46 lb per ton) plus 25 cu ft dry nitrogen. Calcium carbide at \$128.00 per ton, dry nitrogen at \$1.42 per 100 cu ft.

(c) Basic cupola refractory costs as given by Carter. Desulphurizer 4.0% calcium carbide (80 lb per ton) added to cupola charge.

(d) Proprietary reladling process employing 20-30 lb caustic soda per ton on treated iron. Cost figures by Best.²

(e) Based on 1.25% addition of 13% zirconium ferrosilicon at \$0.066 per lb. ried out in the induction furnace, its translation into foundry practice is fraught with great difficulties due to the lack of adequate means of maintaining the hot metal at optimum temperatures. While it is possible that superheating in the cupola (through oxygen enrichment of the blast, or other suitable means) might be resorted to as a means of maintaining temperature during desulphurization, no commercial heats of this type have been produced in this investigation.

While the two desulphurization methods utilizing calcium carbide are substantially more expensive, they nevertheless represent commercially feasible procedures that have actually been carried out in foundry scale operations. Of the two, best results were obtained with the basic-lined cupola. The advantages of substantial carbon pickup, permitting increased quantities of steel scrap in the charge, very high spout temperatures, conducive to rapid dissolution of the zirconium addition, and the exceptionally good desulphurization, assure the effectiveness of the treatment and at least partially offset the somewhat higher cost.

With regard to reduced costs, reference is made to a comparatively new proprietary treatment in which caustic soda is employed to desulphurize steel through a reladling technique. According to Best,2 it is possible to reduce the sulphur content of molten iron from 0.20 per cent S to as low as 0.010 per cent S, using 20 to 30-lb of caustic per ton of treated iron. In large ladles a temperature drop of 150 to 200 F is experienced. The cost of this treatment is reported to be approximately \$1.00 per ton. If this method of desulphurization can be employed with the proposed zirconium addition treatment, a total cost of \$2.75 per ton of treated iron would be attainable. It is believed that experimental work on the application of this method to the proposed zirconium treatment would be of definite value in proving its economic and commercial feasibility.

Comparison of Machinability Tests

Throughout this investigation machinability measurements were based on the simple drill penetration test introduced by Keep and Loren¹⁰ and described by the author in the original paper.¹ While this test

cannot be regarded as a highly accurate quantitative method for evaluating machinability, it nevertheless provided a simple, economical means for comparing semi-quantitatively the large number of specimens studied. The use of the drill penetration test in this manner served as a satisfactory screening procedure to indicate the more promising specimens for subsequent evaluation by more elaborate machinability tests. Since the chief objective of the investigation was to overcome the harmful abrasive effects of phosphorus on the machinability of gray iron, the tool life test was a logical choice for determining the effectiveness of the proposed treatment.

For this purpose the test devised by Boston¹¹ was selected. In this test the specimen is turned on a lathe under standard machining conditions, and the time required for failure of the cutting tool is measured at various cutting speeds. Curves are then plotted for each specimen relating tool life to cutting speed.

Through the cooperation of Prof. O. W. Boston of the University of Michigan, several groups of specimens, including untreated high and low-phosphorus irons, annealed high-phosphorus iron, and specimens of desulphurized and zirconium-treated gray iron, were submitted to this test. Suitable specimens consisting of cylindrical castings 3.75 in. o.d. x 2.5 in. i.d. x 13 in. long were produced in several of the larger laboratory heats and in the foundry tests. While some difficulty was encountered due to the presence of sand inclusions and lack of uniformity in certain specimens from the foundry heats, sufficient data were obtained to indicate the value of the treatment. Table 6 enumerates and compares the results of these tool life tests with drill penetration ratings on the same specimens.

0

ir

ol

si

m

bi

iu

is the

tic to a c Ap iza ou \$3.

It will be noted that both the drill test and the tool life test rated the first four specimens in the same order; and although the degree of improvement varied somewhat between the two tests the correlation is good. In this connection it is of interest to note that through the use of the proposed treatment, tool life ratings for high-phosphorus gray iron may be made equivalent to those of low-phosphorus irons.

At the present writing at least one production foundry in the Birmingham district (The Production

TABLE 6-COMPARISON OF TOOL LIFE AND DRILL PENETRATION MACHINABILITY TESTS

C			Analysi	s, %						
Specimen No.	T.C.	Si	Mn	s	P	Zr		Rema	rks	
91	3.38	2.14	0.60	0.057	0.61		High-phosph	orus iron untreat	ed.	
92A	3.34	2.12	0.56	0.057	0.65		High-phosph	orus iron anneale	d 1500 F 1 hr.	
94	3.46	2.08	0.55	0.059	0.26		Low-phospho	rus iron untreate	d.	
99	3.61	2.09	0.57	0.011	0.63	0.105	High-phospho	orus iron desulphu	rized with ziro	conium addition.
C-22-7	3.82	1.83	0.59	0.011	0.55	0.048		eateda specimen fr		
						Prop	erties			
Specimen				Depth				Test	Tool	Life Test ⁴
No.		Hardne	SS ^a	Chill, in.	0%	Ferrite*	mm/min.	per cent	V ₂₀ °	per cent
91		B-89		0.19	,,	0	51.2	100	133	100
92A		B-58				100	82.0	160	324	244
94 99		B-84		0.09		10-20	58.7	115	180	135
		B-84		0.09		20-30	77.0	150	182	137
C-22-7		B-82		0.06		30-40	68.3	133	187	141
				(a) 1.0% ad	dition	13% zirconi	um ferrosilicon.			
				(b) Rockwe	ll "B" s	cale 1/16-in.	diam steel ball wi	th 100-kg load.		
				(c) Percent	ferrite	in matrix s	tructure.			
				(d) Tool lif	e test 1	by Boston.11				
							(fpm) for 20-min	tool life.		

8

Foundries Division of Jackson Industries, Inc.) is regularly employing desulphurization and zirconium-ferrosilicon additions to achieve improved structure and machinability in its product. While the degree of desulphurization and zirconium additions employed are somewhat less than those recommended in the proposed treatment, tests have indicated sufficient improvements in the properties and machinability of their treated castings that this foundry has conducted additional studies of the effects and methods of addition of zirconium alloys to gray iron. It is understood that the results of these investigations will be published at a later date.

Conclusions

The following specific conclusions may be drawn from the results of this phase of the investigation:

1. While magnesium additions are effective in reducing the sulphur content of high-phosphorus gray iron, the increased chilling tendency accompanying the use of magnesium renders its use unsatisfactory for machinability improvement.

2. Synthetic basic slags consisting of lime, fluorspar, and cryolite have shown definite promise as effective desulphurizers.

3. Bismuth additions were found to be highly potent in influencing the structure and machinability of high-phosphorus gray iron. However, adverse chilling and carbide stabilizing effects more than overcame the beneficial effects observed.

4. Soda ash desulphurization, applying the countercurrent principle to experimental cupola operation, was not altogether successful due to excessive heat losses. However, it is believed that a highly efficient treatment may be devised utilizing this principle if optimum operating conditions are maintained.

5. Desulphurization by calcium carbide injection was shown to be quite effective. While more expensive than soda ash, calcium carbide has the advantages of freedom from attack on ladle refractories, and much greater desulphurizing efficiency.

6. The basic-lined cupola employing calcium carbide in the charge has been shown to be a highly satisfactory means of producing low-sulphur iron for application of the proposed desulphurization-zirconium addition treatment.

7. A comparison of costs for various desulphurization methods indicates that soda ash desulphurization is the least expensive of the methods studied, while the basic cupola process is the most expensive. Unfortunately, soda ash desulphurization to the required low sulphur content (0.02 per cent S max) has not yet been accomplished in foundry scale tests.

8. A proprietary reladling process employing caustic soda as the desulphurizing agent has been reported to give results suitable for the proposed treatment at a cost of approximately \$1.00 per ton of treated iron.² Application of this process to the proposed desulphurization-zirconium addition treatment could be carried out at a total cost of \$2.75 per ton of treated metal, or \$3.91 per net ton of finished castings, assuming 70 per cent yield.

9. Tool life machinability tests have essentially

confirmed the findings of earlier drill penetration tests for evaluating the machinability of specimens in various conditions.

10. Annealing gives the greatest improvement in machinability of all the treatments studied both in terms of drill penetration and tool life tests.

11. Tool life machinability tests indicate that the proposed desulphurization-zirconium addition treatment will make tool life ratings of a high-phosphorus (0.65 per cent P) gray iron equivalent to an untreated low-phosphorus (0.25 per cent P) iron of normal S content.

12. At least one production foundry in the Birmingham district is presently employing a modified form of the proposed zirconium treatment in the production of light gray iron castings.

Acknowledgment

The author wishes to acknowledge the invaluable assistance and support of the following individuals and their respective companies without whose cooperation the work would have been impossible: Mr. B. C. Colcord, President, Woodward Iron Company, Woodward, Alabama, sponsor of the research project; Dr. James T. Mackenzie, Mr. C. K. Donoho, and Mr. S. F. Carter of the American Cast Iron Pipe Company, Birmingham, Alabama, for sound technical counsel and assistance in the execution of numerous experimental cupola and foundry tests; Messrs. J. O. Jackson, R. L. Jackson, F. S. Middleton, and W. C. Jeffrey of Jackson Industries, Inc., Birmingham, Alabama for calling attention to the problem and for additional technical assistance in the preparation of experimental cupola heats and production specimens; Mr. Thurston W. Harwell of The Linde Air Products Company, Birmingham, Alabama, for technical assistance and for furnishing calcium carbide; and to Prof. O. W. Boston of the University of Michigan, for conducting tool life machinability tests.

Bibliography

- 1. Wm. W. Austin, Jr., "Improvement of Machinability in High Phosphorus Gray Cast Iron," Transactions, American Foundrymen's Society, vol. 56, pp. 431-444 (1948).
- Foundrymen's Society, vol. 56, pp. 431-444 (1948). 2. E. P. Best, "Special Ladle Technique Eliminates 90% of Sulphur in Iron," Steel, vol. 126, June 19, 1950, pp. 108-111.
- 3. Propzietary Process of The Linde Air Products Company, New York, N. Y.
- S. F. Carter, "Basic-Lined Cupola for Iron Melting," Transactions, American Foundrymen's Society, vol. 58, pp. 376-392 (1950).
- Society of Automotive Engineers, S.A.E. Handbook, 1947
 Edition, p. 388, Society of Automotive Engineers, Inc., New York, N. Y.
- F. R. Morral, "Nodular Iron, A Bibliography," American Foundryman, vol. 18, July 1950, pp. 61-63.
- E. K. Smith and H. C. Aufderhaar, "Effect of Adding Bismuth to Cast Iron," Metals and Alloys, vol. 2, Nov. 1931, pp. 282-286.
- 8. E. S. Renshaw, "Basic Cupola Process for Desulphurization," Foundry Trade Journal, vol. 70, June 24, 1943, p. 149.
- Technical Council, Institute of British Foundrymen, "Desulphurization and Dephosphorization of Molten Cupola Iron in Basic Lined Ladles," Foundry Trade Journal, vol. 81, April 3, 1947, pp. 259-267.
- 10. G. Schlesinger, "The Determination of Machinability" Machinery (London) vol. 69, Oct. 3 (1946).
- 11. O. W. Boston, Metal Processing, J. Wiley and Sons (1941)

DISCUSSION

Chairman: H. H. WILDER, Vanadium Corp. of America, Detroit.

Co-Chairman: J. L. Brooks, Muskegon Piston Ring Co., Sparta, Mich.

F. W. BOULGER (Written Discussion): 1 It is always reassuring to have additional evidence that simple laboratory machinability tests can give useful and reliable information. The correlation between ratings in drill tests and tool lives on lathes is very good, considering the difference in machining conditions. Subsequent experience in the commercial foundry showed that the author's laboratory ratings were reliable.

The author found that bismuth has a very potent effect on the machinability of cast iron. In addition to changing the microstructure, bismuth may also act as a lubricant during cutting. It seems likely that much of the bismuth is insoluble in cast iron, especially when the additions are as large as 0.10 per cent. Peoples and Pray found that bismuth markedly improves the machining properties of austenitic steels. It does not affect the microstructure of such alloys.

The poor machining properties of high-phosphorus gray irons have barred them from many applications. Therefore, Austin's work is of considerable importance to Southern industry. He has demonstrated that the machinability of such gray irons can be improved by comparatively simple desulphurization treatments and zirconium additions. The cost of the treatments do not seem excessive in view of the benefits obtained. For many applications,

lower machining costs will more than offset the higher price of treated gray iron. This has been the experience with wrought steels where special additions are commonly made to improve machinability.

The author is to be complimented on his careful, thorough,

and successful study on high-phosphorus gray iron.

J. T. Mackenzie: I would like to congratulate Dr. Austin for a fine piece of work. However, he does not bring out in that set of costs the very important item of getting low-phosphorus iron in the South. If he would credit the process a \$15 differential on pig iron you would have to use, you can see what it really does mean to the Southern district.

In regard to the carbide desulphurization in the ladle, I understand there is a more efficient unit now. Instead of blowing in a blast of nitrogen, you just keep the nitrogen flowing gently out of the tube. There is a sort of pepper grinder arrangement for dropping the carbide into the ladle. As Dr. Austin pointed out, there is no refractory attack at all in this method of desulphurization. I think it has a great future.

DR. Austin: I would like to thank Dr. Boulger and Dr. Mac-Kenzie for their kind comments on this paper. The idea of bismuth as a lubricant certainly has interesting possibilities. I am grateful to Dr. MacKenzie for pointing out the important price differential between the high-phosphorus Southern pig iron and the low-phosphorus Northern pig iron. This is especially important when one considers that the low-phosphorus pig iron would have to be imported into the South.

iı

ir

ba

51-

¹ Supervising Metallurgist, Battelle Memorial Institute, Columbus, Ohio.

² Technical Director, American Cast Iron Pipe Co., Birmingham, Ala.

COMPACTION STUDIES OF MOLDING SANDS

By

R. E. Grim* and Wm. D. Johns, Jr.**

ABSTRACT

The compaction characteristics of Illinois fireclay, Illinois illite, Wyoming bentonite, and Mississippi bentonite-bonded sands were studied by means of an apparatus especially designed for the purpose, using varying amounts of sand-clay-water components.

Illite and fireclay exhibited compaction properties distinctly different from those of the bentonite-bonded sands. Under like conditions, compaction decreased in the order fireclay, illite, Wyoming bentonite, and Mississippi bentonite.

Compaction characteristics were correlated with differences in gross bulk densities. The relationship among compactability, gross bulk density, maximum green strength, and dry strength were brought out.

Introduction

COMPACTION AND FLOWABILITY of a molding sand are functions of the plastic properties of the claywater system used as a bond. The bond clays ordinarily employed are known to exhibit diverse properties, and it is therefore not surprising that molding sands show different compaction properties depending upon the type of clay bond used. Also the importance of minor variations in amount of tempering water in the sand-clay-water system is not fully appreciated. The properties of certain clay-water systems are extremely sensitive to minor change in water content, necessitating close control in practice.

It is known that practically any desirable green strength can be attained with any clay bond, depending on the proportions of sand, clay, and tempering water used. Therefore, it is suggested that compaction characteristics be given priority in choosing the type of clay bond to be employed.

Experimental Procedure

A simple testing device was constructed to study the compaction characteristics of several synthetic molding sands. It consisted of a standard 2-in. pipe-T, the central opening of which was fitted with a 10-in.

length of 2-in. pipe (Fig. 1). These were fastened securely to a heavy steel base to insure a constant vertical position. The additional openings of the "T" were fitted with 2-in. plugs that could be readily unscrewed. A piston, machined to fit the inside of the vertical pipe section, pushed and compressed the sand into the apparatus.

The testing procedure adopted was as follows: Both plugs were inserted in the "T" arms, and 900 grams of molding sand (prepared according to standard A.F.S. procedure) was introduced through the top. The piston was inserted in place, the entire apparatus placed in a laboratory-size hydraulic press, and the sand compressed under a pressure of 640 psi. Other pressures tried gave substantially the same conclusions. The piston displacement was recorded in millimeters. One plug was then removed, and pres-

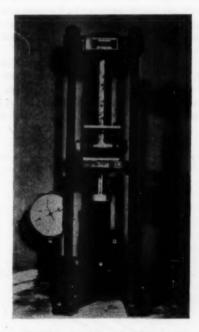


Fig. 1-Photo of testing device used in these studies.

^{*} Research Professor of Geology, Univ. of Illinois, and Consulting Clay Mineralogist, Illinois State Geological Survey, Urbana, Ill.

^{**} Research Assistant, Engineering Experiment Station, Univ. of Illinois, and Illinois State Geological Survey, Urbana, Ill.

sure was applied again to the piston until the sand extruded 1 centimeter from the opening. The movement of the piston during this operation was again recorded in millimeters.

Clay-sand mixtures involving 4, 6, 8, and 10 per cent Illinois fireclay, Illinois illite clay, Wyoming bentonite, and Mississippi bentonite bonds were tested at varying moisture contents. Families of curves were constructed, plotting piston displacements as a function of tempering water (Fig. 2-5). Sands were mulled and tempered before testing according to standard procedure.

Ottawa silica sand showing the following sieve analysis was used in these tests:

	Sieve size*	Per cent
Retained on	30	0.1
Retained on	40	16.63
Retained on	50	56.41
Retained on	70	17.40
Retained on	100	7.98
Passed	100	1.46
* Sieve numbers	are those of U.	S. series.

Graphical Representation of Results

All experimental data have been plotted graphically as shown in Fig. 2 to 5. Families of curves are shown for each type of bond used, individual curves representing a given amount of clay bond, with piston displacements plotted as a function of tempering water. The constancy of certain trends of these curves suggests significant variations which can be interpreted in terms of differences in compaction properties of the sands employed. It might appear from analysis of Fig. 2 to 5, and from consideration of the experimental procedure used, that the bulk densities of these sands are being measured. Examination of the graphical data shows that the curves are similar in appearance to bulk density curves published earlier by Grim and Cuthbert.¹

These curves show several significant differences, however, which distinguish them from bulk density curves. In the case of bulk density curves derived from determinations made with standard A.F.S. test pieces, the points of minimum bulk density trend to-

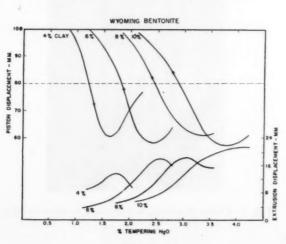


Fig. 2

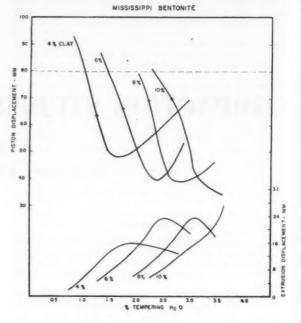


Fig. 3

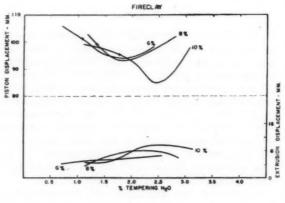


Fig. 4

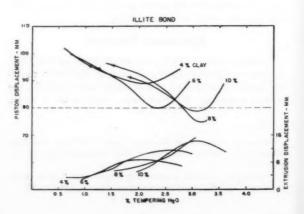


Fig. 5

ON DISPLACEMENT - MM

ACEMENT - MIN.

EXTRUSION

ward higher values with increasing clay content. Figures 2 to 5 show the trend in the opposite direction. Especially in the case of Wyoming and Mississippi bentonites, the minimum densities shown in the present tests decrease with increasing clay contents, whereas the maximum compaction increases with increasing clay contents.

It would appear that in addition to normal variations in bulk density, significant variations in piston displacements are affected by conditions of non-uniform compaction in the testing apparatus. The latter condition is predominant in producing variations in the gross bulk density, that property being determined in the present study. It is not surprising that variations in bulk density as determined by the standard procedure should be at variance with those produced by the present method of study, because in the formation of a standard test piece a very small amount of sand is compacted into a small cylinder, necessitating no horizontal movement of the sand with the vertical application of pressure. In the case of the apparatus used for the present discussion, the amount of sand is increased almost ten-fold and irregularities are introduced in the compaction chamber (the horizontal "T") necessitating appreciable horizontal movement of the sand for complete compaction upon vertical application of pressure. It would seem that conditions produced here are more in accordance with those of actual working conditions than those encountered in the formation of a standard A.F.S. test piece.

It should also be noted that in the determination of bulk densities by the standard procedure, variations in minimum bulk densities, even between differing types of clays, say bentonite and fireclay, are of the order of 1 to 2 per cent. In the case of the method employed in the present study, differences in bulk density between bentonite and other bond clays are of the order of 20 to 30 per cent. Again it would appear that the development of large voids and poor compaction are in this case the dominant factors in affecting gross bulk density variations. If we assume that this procedure somewhat nearly approaches actual working conditions, it would appear that the standard test methods are inadequate.

Discussion of Graphical Data

On the basis of the foregoing discussion, a decrease in piston displacement as given in Fig. 2 to 5 is interpreted as a decrease in gross bulk density, and, therefore, compactibility of the sand. Observation of the sands in the compaction chamber of the apparatus after compression, but before extrusion, showed that whenever piston displacements dropped below 75 to 80 millimeters, there was obvious visible evidence of incomplete compaction and distribution of the sand, the decrease in uniformity of compaction varying directly with the piston displacement. It was also noted that those mixtures showing piston displacements below 75 to 80 millimeters had a tendency to form aggregates or clumps when the sand was introduced. The sticky consistency of the bentonite bonds permits little mutual adjustment of grain aggregates of

the sand, which is to say, the sand exhibited little flowability. When the values rose above 75 to 80 millimeters, the sands lost this characteristic and tended to be free-flowing.

Bentonite-bonded Sands

Figures 2 and 3 show that the Wyoming and Mississippi bentonite bonded sands are similar. Over a considerable range of moisture contents piston displacements dropped far below 75 to 80 millimeters, the apparent critical point for the development of poor compaction. Here the development of large voids and non-uniform compaction is dominant in influencing the gross bulk densities. The bentonite-bonded sands are those which exhibited the tendency to develop large aggregates or clumps when the sand was introduced into the compaction chamber. It would appear that each of these clumps has some inherent mechanical strength, and that the cumulative strengths of all the aggregates were sufficient to resist mutual adjustment or flowability, and, consequently, compaction when pressure was applied, and to prevent complete filling of the compaction chamber.

For an individual mixture, illustrated by a single curve, this mechanical strength of clumps increases with increasing tempering water to the point of minimum piston displacement (minimum gross bulk density), at which point the sand offers maximum resistance to compaction. Further increases in tempering water render the sand plastic and effectively weaken individual clumps and allow compaction to proceed. This results in increased piston displacement and compaction.

It was observed that the tendency for Mississippi bentonite mixtures to form such aggregates was greater than for Wyoming bentonite. This would account for its poorer compaction characteristics. Figures 2 and 3 also show that compaction characteristics of Wyoming and Mississippi bentonite-bonded sands are extremely sensitive to variations in amount of tempering water. A small change in the amount of tempering water in these sands will produce significant differences in compactibility. Thus in the case of bentonite-bonded sands, careful control must be exercised over the amounts of tempering water added and maintained.

In addition to gross bulk density relationships, Fig. 2 and 3 illustrate curves derived by plotting piston displacement during extrusion as a function of tempering water. These curves conform quite well with the gross bulk density curves. As the gross bulk density decreases, the amount of displacement during extrusion increases to the point of minimum gross bulk density. It appears that during extrusion of the sand, a great deal of the movement of the piston, in addition to moving the sand forward, is taken up in compacting the sand by filling appreciable void space that must have existed in the sand column. This gives additional evidence that the bentonites exhibit poorer compactibility during compression.

Fireclay- and Illite-bonded Sands

Figures 4 and 5 show that the fireclay- and illitebonded sands are similar to each other, but that they differ appreciably from those bonded with the bentonites. Here the order of magnitude of piston displacements is much lower. In practically all mixtures, regardless of water content, piston displacements exceeded the critical 75 to 80 mark. These sands were also free-flowing and showed no tendency to form clumps or aggregates. Visual observation showed that compaction and uniformity of sand distribution in the compaction chamber was good following com-

pression.

Differences in compaction appear to be small even among sands bonded with varying amounts of clay. In general fireclay appears to show slightly superior compaction properties over illite clay, yet both exhibit properties markedly superior to the bentonites. In the case of the fireclay- and illite-bonded sands careful control of tempering water is not as important in influencing variations in compaction as in the case of the bentonites. Comparatively wide variation in water content will cause only slight or moderate differences in compaction. This should be an important factor in general use.

The extrusion displacement curves in Fig. 4 and 5 conform to the gross bulk density curves in that they exhibit little variation in piston displacements with increasing amounts of tempering water and clay. Most of the piston movement is effective in pushing the sand forward during extrusion, showing that little void space and, therefore, good compaction existed

after compression.

It was also noted during extrusion that the pressure required for the poorly compactible mixtures of bentonite sands was roughly twice that for the fireclayand illite-bonded sands. It is possible to visualize that in the shaping of a mold the greater portion of the sand attains high density on compression, but other portions of the mold, because of irregularities, will contain sand of low density or even be incompletely filled. Only additional pressure will force sand into such irregularities. The experimental data suggest that in order to accomplish this the bentonites would require much more force. Stated another way, under given compression the fireclay- or illite-bonded sands would pack much more uniformly.

In order to determine the relative merits of compaction of the four clay-bonded sands studied, it is necessary to compare them under like conditions. Comparison of those curves representing equal additions of clay, say 6 per cent, suggests that at varying moisture contents the sands exhibit decreasing compactibility in the order fireclay, illite clay, Wyoming

bentonite, and Mississippi bentonite.

Compaction vs Compression Strength

Bentonite bonded sands — The points of maximum green compression strength for each mixture of bentonite-bonded sand have been plotted on the curves shown in Fig. 2 and 3 (indicated by X). It will be noted that in the case of both bentonites, maximum green strength falls on the dry side of the points of minimum compaction. Therefore any slight increase in tempering water beyond the point of maximum green strength would only tend to decrease further the compactibility of the sands.

There would appear to be two alternatives with respect to most efficient use of the sands. One would be to maintain the tempering water in amounts falling on the dry side of maximum green strength, thereby providing good compaction with some loss of green strength. The other would be to add tempering water in amounts greatly exceeding that of maximum green strength, passing over the hump, so to speak, beyond the point of minimum compactibility. Either alternative would result in the reduction of green strength below the maximum value, but the first alternative would also result in marked reduction in dry strength, eliminating it as a possibility. Maximum dry strength comes at moisture contents much greater than for maximum green strength. Therefore, taking green strength, dry strength, and compactibility all into consideration, maximum efficiency requires maintaining tempering water in amounts greatly exceeding that necessary for maximum green strength. This seems to be the actual practice of most operators.

In order to illustrate these points, let us consider a sand bonded with 6 per cent Wyoming bentonite. Maximum green strength occurs at 1.8 per cent tempering water and lies within the range of poor compactibility. In order to improve the compaction, this sand could be tempered at either 1.5 or 3.2 per cent. This would result in a decrease in green strength from 14 psi at maximum to 12 psi when tempered at 1.5 per cent and to 6 psi when tempered at 3.2 per cent. At 1.5 per cent tempering water, however, the dry strength would become almost negligible, while at 3.2 per cent tempering water it would exceed 100 psi. Therefore, in order to attain good compaction characteristics and at the same time provide adequate dry strength, it is necessary to maintain moisture contents greatly exceeding that for maximum green strength, even though green strength must be sacrificed.

Fireclay- and illite-bonded sands—Figures 4 and 5 show that the points of maximum green strength for fireclay- and illite-bonded sands fall well within the field of good compaction. Therefore these sands can be most efficiently utilized by tempering so as to maintain maximum green strength and at the same time provide good compaction, and close control of tempering water is not necessary.

Conclusions

The graphical data derived from compaction studies of four different molding sands exhibit constancy of certain trends and suggest significant variations on compaction characteristics of the sands studied. The data suggest the following conclusions:

1. Inasmuch as it denotes differences in distribution and gross bulk densities of the sands, the experimental method involved reflects differences in com-

pactibility characteristics.

2. Fireclay- and illite-bonded sands exhibit distinctly different compaction properties from those of

Wyoming and Mississippi bentonite.

3. The experimental data indicate that the fireclay- and illite-sand mixtures afford relatively good compaction and little variation in gross bulk density over a great range of moisture contents. The bentonDS

th

ld

11-

h,

SS

p-

X-

to

of

1e

C-

ts

h.

d

fi-

n

i-

ıl

a

e.

).

l-

is

t,

n

5

ites exhibit all variations from good to poor, depending upon the sand-clay-water ratio. This agrees well with visual observations of sands in the compaction

chamber following compression.

4. In the case of bentonite-bonded sands, compactibility is extremely sensitive to minor variations in amount of tempering water. Fireclay and illite bonds exhibit only minor variation, thus eliminating the necessity of close control in foundry practice.

5. Under the same conditions, that is, utilizing equal additions of sand, clay, and water, compactibility decreases in the order fireclay, illite, Wyoming

bentonite, and Mississippi bentonite.

6. The relation between maximum green compressive strength and compactibility of bentonite-bonded sands indicates that moisture contents greatly exceeding that required for maximum green strength are necessary in order to maintain adequate dry strength as well as compactibility. This results in appreciable sacrifice of green strength, however.

7. In the case of fireclay- and illite-bonded sands, most efficient utilization is attained by tempering so as to maintain maximum green strength and at the

same time good compaction.

Acknowledgments

This study has been carried out as a part of a cooperative research project by the Illinois State Geological Survey and the Engineering Experiment Station of the University of Illinois, under the sponsorship of the Illinois Clay Products Co., Joliet, Ill. The authors wish to acknowledge the assistance rendered by the sponsors, particularly Mr. Otis L. Jones, President, and Mr. Howard F. West, Executive Vice President of the Illinois Clay Products Co., whose interest made this study possible.

References

1. R. E. Grim, and F. L. Cuthbert, "The bonding action of clays; Part I, Clays in green molding sands: Illinois Geol. Survey Rept. Inv. 102, 1945; Univ. of Ill. Eng. Exp. Sta. Bull. 357,

DISCUSSION

Chairman: J. B. CAINE, Foundry Consultant, Wyoming, Ohio. Co-Chairman: R. H. JACOBY, The Key Co., E. St. Louis, Ill.

R. P. Schauss (Written Discussion): 1 Dr. Grim and Mr. Johns made a definite contribution to the advancement in measuring sand flowability by the method presented in this paper. The size of the sand specimen used together with the simplicity of the test make it possible for this test to eliminate some of the variables known to be present in the current flowability test.

It is interesting to note that the data presented here substantiate that presented by Woodliff, Schauss and Baley in a paper

entitled "A Study of Factors Affecting Molding Sand Density, Shrinkage, Expansion and Workability" published in A.F.S. Transactions, vol. 58, p. 217 (1950). The authors have gone one step further in that they show the effects of tempering water on the gross bulk density of molding sands. Their comments on the effect of tempering water versus the properties of Western and Southern bentonite are well taken, since this factor is important in practical foundry operation. It is a well established fact that moisture content of molding sand must be kept within close limits when Western and Southern bentonite are used as bonding mediums. On the other hand, the use of fire clay bonds either alone or in combination with the bentonites allows a wider working range of moisture with improved compaction characteristics in the sand.

CHAIRMAN CAINE: What does this test tell us that any older test does not, and what does this test tell us about the sand's behavior when in contact with molten metal? We have so many tests on sand, so many of them misapplied and misinterpreted,

that I think these questions are important.

Mr. Johns: Our purpose has been to present experimental data which we believed to be of interest to the foundryman, and certainly not to present a testing procedure designed to supplant the older well-established ones. It is for you to judge whether the interpretation of the data as given is justified and for you to evaluate the data in terms of your own individual practice. Nevertheless it has been pointed out that this procedure might more closely represent actual working conditions than some of the standard tests, inasmuch as larger quantities of sands were tested and irregularities were introduced into the testing appar-

Obviously, this test like other tests tells little of what is going to happen when hot metal comes in contact with the sand. All that any sand test can hope to indicate is something of the character of the mold itself. No rapid testing procedure can include all the variables encountered in large-scale foundry

C. W. Briggs: 8 Referring to Fig. 2, the authors used a minimum clay content of 4 per cent. I would be interested in knowing the position of the 2 and 3 per cent clay curves. How would

the lower curve come up? Would it be flat?

The steel casting industry uses 3 and 31/2 per cent bentonite with moisture contents of about 3 to 4 per cent. Curves of such bentonite and moisture contents should be back in the good flowability range. Steel foundrymen normally work sands in the range of 6 psi green compressive strength. I regret that the authors did not carry the curves to these levels. I am interested in what they look like in these ranges.

Co-CHAIRMAN JACOBY: The author mentioned piston displacement. I do not quite understand the term. How was your

original point on that piston determined?

MR. JOHNS: We just introduced the sand into the compaction chamber, then the piston and the pressure and then measured the distance the piston went down into the compaction chamber.

CO-CHAIRMAN JACOBY: The difference between dry sand practice and green sand practice was brought out and the difficulty of trying to make one sand fit both practices.

H. W. DIETERT: 8 The fire clay was in an insufficient amount for normal foundry practice. I would like to see the results with a little more fireclay. It would give us a sand more comparable to that used in the foundry. The moisture content could have been extended in all the graphs.

¹ Foundry Engineer, Illinois Clay Products Co., Chicago.

² Technical Director, Steel Founders' Society of America, Cleveland.

⁸ President, Harry W. Dietert Co., Detroit.

MANUFACTURE OF BRONZE BOILER DROP PLUGS

Ву

B. F. Kline* and J. R. Davidson**

VITALLY IMPORTANT to the safe operation of steam locomotives, boiler drop plugs serve as a warning of low water which might be the cause of a boiler explosion. When boiler water becomes low, these plugs must be capable of functioning under differential stresses of unknown magnitude and must withstand temperatures as high as 420 F at 300 psi on the water side, and firebox gases up to 2800 F on the firebox side. A further complication which adds to the severity of service is the corrosive attack of firebox gases on one end of the plug, and the attack of hot alkaline boiler water on the other. Thus, boiler drop plugs, because of the severe service to which they are subjected, require the best materials and workmanship. This article describes the various phases of their manufacture and testing and investigations of varying fusible alloy compositions under simulated service conditions.

The main part of a drop plug consists of a bronze casting (Fig. 1), in which is inserted a rolled bronze button held in place with a 0.003-in. thick film of fusible metal (Fig. 2). This fusible alloy has a softening point of about 540 F. The plugs are screwed from the firebox side into and through the crown sheet in the top of the firebox until they project not less than 1-1/4 in. above the crown sheet on the water side of

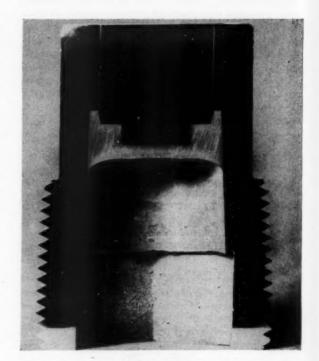


Fig. 2-Cross-section of boiler drop plug. Rolled bronze button is held in place in casting by means of a 0.003-in. film of metal that softens at 540 F.

* Chief Chemist and ** Supervisor of Foundry Operations, Southern Pacific Railroad Co., Sacramento, Calif.

Fig. 1-Bronze castings used in boiler drop plugs.



the firebox.

Functioning principle depends upon the film of bonding metal becoming fused, permitting the button to be blown by high boiler pressure into the firebox. All Southern Pacific steam locomotives have from three to seven of these drop plugs screwed into the crown sheet.

Usually, the plug on the highest part of the crown sheet is the first to let go. Discharge of steam into the firebox serves as a warning to the locomotive crew that something is wrong and that immediate action must be taken to cut the fire off.

Good castings are analogous to good cooking in that good ingredients are a primary requisite. Metals should be clean, dry and free from grease and oil. In making drop plug bodies no scrap returns are used—only new metals—with the exception of copper, since it has been found that heavy scrap trolley copper performs better than new ingot copper in that the film of oxide on the trolley metal counteracts a tendency to absorb hydrogen.

All metals are accurately weighed so as to furnish a composition conforming to the following analysis: copper, 86-88 per cent; tin, 6-7; lead, 1-2; zinc, 3-4; nickel, 1.0 max; iron, 0.25 max; phosphorus, 0.05 max. This analysis conforms to ASTM B143 for leaded tin bronze.

Melting is accomplished in a horizontal tilting furnace, using a combination fuel of oil and gas. Originally, fuel oil was used for melting, but later a combination burner was adopted, using low-pressure natural gas of around 950 BTU at 5 psi. Gas and oil valves are regulated to furnish approximately a 50-50 mixture in which the gas vaporizes the fuel oil for more efficient combustion.

Combustion is regulated so as to keep it slightly

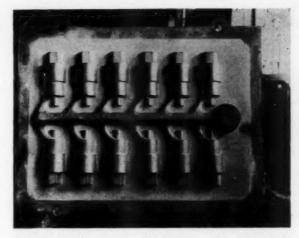


Fig. 3-Cope side of aluminum matchplate used in casting bronze boiler plugs. Drag side is identical.

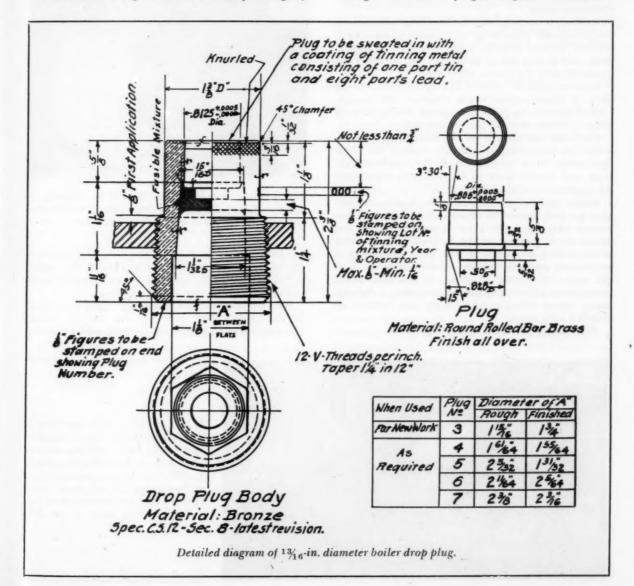




Fig. 4-Boiler plug cores are carefully baked to avoid pinhole porosity in the bore of the casting.

on the oxidizing side, but as an added safety measure the melt is oxidized with sodium nitrate and then deoxidized with phosphorus copper, which reduces porosity from hydrogen absorption. The furnace is brought up to 2200 F and its contents poured into preheated ladles.

Molding is accomplished on molding machines of the jolt-squeeze type, using matchplates of aluminum. Natural bonded molding sand is used. All pop-off flasks and jackets are of aluminum and all cores are produced with California white sand.

Since no pinhole porosity can be tolerated in the bore of these bodies, which receives the buttons, careful attention is paid to thorough baking of cores (Fig. 4), before they are placed in molds.

The body and button are machined in progressive stages on turret lathes specially tooled for these operations. After machine work is completed, the body and button move to the finishing department for application of the fusible alloy.

Before applying the fusible metal, bodies and buttons are first cleaned by immersion in a hot alkaline cleaner (Fig. 5) at 180 F for 15 min, to remove any grease or oil present on the finished surfaces. After removal they are rinsed in hot water, immersed in hot, strong muriatic acid for 10 min, removed from the acid and allowed to drain. It has been found advisable to follow this procedure to insure a chemical bond. From this point, the body and buttons are handled with tongs.

The bodies are next preheated in two electric

ovens thermostatically controlled at 650 F. with a circulating fan in each oven. Dry nitrogen gas is piped to each oven and is fed at a rate of approximately 2-3 liters per min, controlled by a flow meter to prevent oxidation at high temperature. Bodies are fed into the oven and removed so as to obtain a heating cycle of approximately 10 min. After removal from the oven they are dipped into a special flux, consisting of zinc chloride, ammonium chloride and water acidified with muriatic acid. Next, they are immersed in the fusible alloy, which consists of 11.10 per cent tin and 88.90 per cent lead, or roughly 1 part tin to 8 parts lead.

The fusible alloy is melted and kept molten in a circular cast iron pot heated electrically and maintained at 650 F-675 F by thermostatic control. The



Fig. 6-Body and button are tinned by dipping in fusible alloy. Alloy is poured into 1-in.-diameter stainless ladle to depth of 1/8 in. over button.

pot holds 95-100 lb of metal and is heavily insulated on the outside to conserve heat and insure the operator's comfort. Heating elements are cast into the walls of the pot and operate on 220 volts a-c.

Both the body and the button are tinned by dipping in the fusible alloy (Fig. 6). The film of fusible alloy surrounding the button is obtained by pouring



Fig. 5-Before applying fusible metal that holds them together, boiler plug bodies and buttons are immersed in hot alkaline cleaner at 180 F for 15 minutes to remove surface grease or oil, then rinsed in hot water and immersed in hot, strong muriatic acid for 10 min.

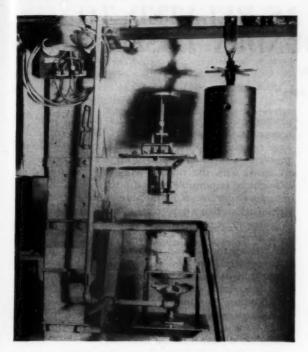


Fig. 7-Apparatus used in testing boiler drop plugs.

a predetermined volume of molten alloy to a depth of ½ in. over the top of the button. A small, stainless steel pouring ladle, slightly more than 1 in. in diameter, is used for this purpose. The fusible metal is cleaned of mechanically trapped impurities by bubbling dry nitrogen gas through it for 8 to 10 min at the end of every hour.

The whole assembly is then carefully transferred to a cooling plate to solidify. This cooling plate consists of a hot plate held at 140 F by thermostatic control. It has been found that slow, controlled cooling will prevent hairline cracks in the fusible alloy. After the metal has solidified and the plugs are reasonably cool, they are given a final wash and buff. Six out of every 125 are picked at random and sent to the laboratory for testing.

Examined For Dropping Temperature, Bond

When finished drop plugs are received in the laboratory, they are divided equally for different examinations. One-half of the plugs are tested for dropping temperature. This is done by placing the plug to be tested in a supporting saddle immersed in an agitated oil bath preheated to a temperature of 515 (Fig. 7). A weight is applied on the button of the drop plug so that the load over the button face equals 300 psi.

Temperature of the oil bath is raised slowly until the plug functions. Function should occur under such conditions at a temperature of 530 F to 555 F. After functioning of the plug, the oil bath is cooled to 515 F, at which point another plug is inserted and the cycle repeated. This cooling period between the testing of the plugs is essential. Temperature at time of plug functioning is measured with a mercury thermometer.

The other half of the test lot of plugs is examined for bond as follows: a plug is split by first sawing a ½-in. V in opposite sides of the threaded end. Then, without sawing through the button, the rest of the plug body is sawed in line with the apex of the V. The threaded end is then placed in a vise and pressure is applied until the plug body separates from the button, exposing the alloy bond area.

If the bond of any plug shows other than a uniformly frosty appearance over the face of the split bond, or if in the dropping test any plug functions outside the temperature limits 530 F to 555 F, that plug is considered to have failed. In that event, six more plugs of the same serial and operator number are obtained and either split or dropped. If any of these six should fail, that operator's lot of 82-84 plugs is reworked and resubmitted for testing. If such plugs fail to pass a second time, the lot is scrapped and a new set of plugs processed from new bodies and buttons.

Reduce Button Creep Tendency

In order to reduce tendency of buttons to creep at such boiler pressures as 300 psi, research work has been conducted on such phases of their manufacture as: (1) various fusible alloys, such as silver solder, reduced tin content and pure lead; (2) nickel plating part exposed to boiler water; (3) heat treatment and aging; and (4) design change.

Preliminary tests were run in the laboratory using four small steel bombs with each end tapped to receive a drop plug. On all tests a standard drop plug was screwed in at one end and the plug under investigation at the other end. The bomb was filled with typical boiler water and the whole assembly placed in a thermostatically controlled oven heated to produce 300 psi in the bomb. Each test was run for 168 hr. At the end of this period the bombs were taken from the oven to cool and the drop plugs removed for examination.

These tests indicated that silver solder has a tendency to corrode, while nickel plate rapidly ruptures. Lowering tin content causes difficulty in processing. The final outcome of these tests was to adopt a change in design whereby length and diameter of the button were increased ½ in. This increased the effective bonding area, but maintained the standard fusible alloy composition of 88.90 lead and 11.10 tin.

To confirm laboratory findings, road tests were run on locomotives equipped with the longer buttons and it was found this design change materially reduced tendency of the button to creep. The change was then adopted as standard.

CORE PRACTICE AS RELATED TO MALLEABLE FOUNDRY LOSSES

By

E. J. Jory*

EVERY DAY IN EVERY FOUNDRY a great amount of effort is expended in reducing casting losses and improving quality levels. The reasons for this are obvious—efficiency, costs, and customer satisfaction.

Since many scrap castings may be the result of more than one cause, it is not always easy to pin down the practice which results in foundry losses. The most important factor in overcoming any type of scrap is the use of the powers of observation. This may be a case of merely looking at a scrap casting critically and visualizing what must have happened to produce it, or by correlating operating or experimental data with resulting quality levels by statistical means. Unless an open minded approach by all concerned is adopted, and unless an intelligent analysis based on thorough observation is made, it is extremely difficult to overcome foundry losses.

Since this discussion of losses is to consider only the effect of core practice, it is well to first point out that the part which cores may play in the production of castings which are scrapped or require salvage operations, is not always obvious. Because of this, scrap produced by cores often is not charged to the coreroom. On the other hand, the coreroom supervisor may be charged with scrap which is not caused by the core practice, due to faulty analysis.

Accurate Scrap Analysis Needed

As an example of this the case of a large and complicated casting being scrapped because of a thin section in the drag can be cited. This scrap, which amounted to about 4 per cent of the castings produced, had been analyzed by the molding foreman as being caused by sagging cores. The coreroom foreman, who had observed the scrap castings with the molding foreman, had accepted responsibility for the losses and was trying in several ways to eliminate the core sag.

This situation had existed for several months with occasional improvement, but the loss was not permanently eliminated. Finally, at the suggestion of an outsider, a gage was made to check the cores for sagging. No sag was found, and in some cases the core was off in the opposite direction. Further observation showed that the thin drag sections were caused by the molders placing the drag mold on dirty bottom boards, which cause the drag mold face to be pushed up.

Incidental to this situation, a large increase in the occurrence of hot tears in this casting resulted because the coreroom had gradually increased the hardness and strength of the body core to the point where it restricted the normal contraction of the metal in the sections around the core.

The fact that situations such as that related in the

foregoing example exist, indicates that in order to reduce foundry losses it is necessary that all concerned have the ability to use the apparent facts to solve the problem. Coreroom supervision should become acquainted with the various types of scrap for which it might be responsible, and also that which the other foundry departments might cause. Since in many cases the supervision in a department comes up through that department, it often is not familiar with the problems of others.

At times in the core department the scrap is not conveniently seen, and it is advisable for core personnel to make a daily practice of reviewing the scrap, and by so doing gain knowledge and experience from the problem encountered. Some of the specific types of defects for which the coreroom may be responsible, and means which have been successful in eliminating them, are discussed in the following.

Hot tears result when the restricting forces set up by the mold or core are greater than the ultimate strength of the metal during the cooling period after solidification. In this case the desirable approaches are to increase the ultimate strength of the metal, or reduce the restricting forces.

Here the coreroom is charged with the responsibility of producing a core which will have properties such that it will offer a minimum resistance to the contracting metal. This problem has been attacked in many ways and with varying degrees of success.

One method which has been found quite effective is that of controlling core density or, more simply, core weights. This requires proper ramming of the core in hand-ramming operations. To obtain a core of minimum density requires the least amount of ramming commensurate with retaining sufficient dimensional stability. In the writer's shop it has been found that to control cracking tendencies on a particular casting, it is necessary to hold the weight of an internal core to 12 lb maximum weight.

Control Core Ramming

It has been proved that cores weighing 13 lb will produce 100 per cent hot tears. On the other hand, it has been found that cores rammed to 11½ pounds tend to sag, with resultant loss in holding dimensions. To obtain this degree of control requires that the coremaker be carefully trained, since the surface of the core must be well packed to avoid casting roughness while the body of the core be softly rammed.

In making such a core the coremaker's only tool is the strike-off. He is not permitted to use any aid to ramming other than skillful use of his hands. The sand for such cores is made by blending base sands of different finenesses so that the grain is distributed over six sieves. Sufficient cereal is used to bond the

^{*} National Malleable & Steel Castings Co., Cicero, Ill.

15

sand to about 1.2 psi green strength. With this sand practice sufficient green strength is obtained from the spread of grains and cereal for the core to maintain its shape. The grain distribution spread also helps to maintain desired casting surface finishes without the use of core washes, in spite of the soft ramming.

Control of this practice is maintained by weighing the cores at random as they come from the ovens. When core weights exceed the desired minimum, the core supervisor is immediately notified and takes corrective action. This practice has, in some cases, been found to warrant making a job by hand rather than on the coreblower.

In cases where hot tears persist in spite of controlled core weights, it has been found helpful to use a filler in the sand. So far the most effective filler found has been sawdust, which means materials having a particle size of about 20 mesh. This material used in proportions up to 25 per cent by volume has been quite effective. The use of wood flour in equal amounts, in the author's experience, had not been nearly as effective. There are undoubtedly other fillers available and in use which would be equal to sawdust, but from the standpoint of costs and results sawdust has been quite satisfactory.

Some Objections to Sawdust in Cores

There are certain objections to the use of sawdust which are well founded. The greatest disadvantage is its effects on casting surfaces, where it produces undesirable roughness. This can be overcome where conditions are most abusive to the core by the use of core washes. Where conditions in the mold are such that cores are not submitted to excessive erosion or heat, the use of sand mixtures with broad grain distribution, as previously mentioned, seems to be sufficient.

Another disadvantage often mentioned is that heaps or sand systems become increasingly filled with sawdust. The author has never seen this condition exist. However, if such a condition should arise, it would be a matter of considering the various factors involved and determining a proper course of action. Sawdust also has the undesirable characteristic of weakening the cores in the dry state and careful handling is required. This can be helped by judicious use of high dry strength facing and rods. Because of its disadvantages the use of sawdust is indicated where no other means of reducing hot tears is effective either in the foundry or coreroom.

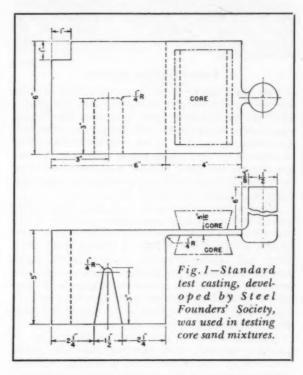
A combination of the two aforementioned practices appears to be the limit to which the coreroom can go in helping to overcome hot tears. However, there are other steps which can and should be taken in all cases. It is desirable to always use the least amount of binder commensurate with core handling. The use of extremely fine sands or sand containing clay should be held to a minimum.

Proper baking is desirable because the presence of unoxidized core oils results in an exothermic reaction which would have an adverse effect on casting hot spots. Keeping the moisture contents of core sands at a minimum is desirable, since increasing moisture contents result in increasing all core properties which would tend to promote hot tearing.

It should be remembered that not all hot tears are

the result of core conditions. In nearly all cases the causes of hot tears are the result of several factors acting simultaneously. However, careful analysis of how stresses are acting in a casting prone to hot tear will indicate whether or not cores are a contributing factor. If it can be seen that cores may be a factor, then steps should be taken to correct the condition and practices devised for that purpose be maintained by some form of control.

Strain cracks are the result of uneven cooling of different casting sections. It is doubtful that the properties of cores have any great effect on the occur-



rence of strain cracks. However, it has been found that in mechanized shops where shakeout occurs shortly after pouring, the time required for a core to shake out may have some effect on strain cracks.

It was noticed that in a hub casting where the hub core disintegrated and ran out of the casting rapidly, there was less tendency toward strain cracks than when the core sand was retained in the casting. Apparently the reason for this was that the rapidly disintegrating core allowed air to penetrate to the heavy sections of the casting, thereby cooling it faster in relation to the lighter flange. More even cooling reduced the strains usually set up with less collapsible cores. This practice was controlled by holding core sand moistures to a minimum.

Core scabs and cuts are the result of intergranular expansion of the sand grains of the core sand mixture. They are usually found in locations which are subject to abuse from the metal entering the mold, and if they are sufficiently serious may develop into what is known as a core cut.

The circumstances present when a core cut or scab appears indicate insufficient strength of the sand at high temperatures. This lack of high temperature strength can be rectified by additions to the sand mixture of silica flour in sufficient amounts, usually at least 20 per cent by weight. The use of core wash cannot be strongly relied upon to prevent scabs or washes.

For experimental or control purposes a very useful tool is the sand test block shown in Fig. 1. Experimental core sand mixtures can be made up into ramup cores, which are rammed up in the mold at cope and drag of the plate section of the pattern. Since the entire casting is poured through this plate section, the cores being tested will receive extreme abuse from the flowing metal. If a mixture is produced which resists cutting or washing in this application, the chances are great that it will perform quite satisfactorily in use.

Reduce Core Gas Pressure

Core blows may be the result of core conditions if the gas pressures developed within the core are greater than the surrounding metal pressures. The logical approach to this problem is to reduce the amount of gas-producing material in the core sand mixture, or to improve the core baking process. However, the core gas pressure and metal pressure relationship are often such that additional help in reducing core gas pressures is required through the use of vents.

When vents are necessary they must be properly placed and made. The appearance of a hole in the end of a core print should not be accepted as conclusive evidence that a core is vented. Very often blows will appear at feeders or gates of the cored heavier sections of light castings. These can often be traced to the core. The reason they appear at the gate mouth probably is that the metal here remains fluid longer than in other sections of the casting and offers less resistance to the passage of gases.

Another source of core blows is in the use of patching material. Core patching materials usually contain high percentages of gas producing materials. This concentration of gas producers in a small area where cores may be patched because of chipping or carelessness can easily result in blows. The best remedy for this condition is to make and handle cores in such a manner that patching is not necessary. Where the use of mud is necessary, as around paste-in cores, it should be used in as small amounts as possible and carefully dried.

Finer Core Sands Improve Casting Surface

The use of core washes sometimes resulted in the presence of very small blows about 1/8 in. in diameter throughout sections of a casting. When washed cores were applied to the plate section of the test block (Fig. 1), these small blows were often found in the plate sections of the casting. Wherever possible it has been found desirable to improve casting surface conditions by the use of finer sands in the core mixture.

Penetration and Rough Surfaces: Casting buyers have become more and more demanding for smooth surfaced castings free from adherent sand grains. The use of washes with cores certainly improves surface conditions; however, aside from the fact that this adds costs of labor and materials to the finished product, it often results in the presence of small gas holes in the casting which are not uncovered until the machining operation.

Except in most severe cases surface finishes can be satisfactorily controlled through the sands used in the core mixtures, without appreciable cost increases. The judicious blending of base sands in such a manner as to produce a five or six sieve sand, with 10 to 15 per cent of the grains below the No. 100 sieve, will result in desirable finishes in most cases.

The use of excessive fine material or clay containing sands will result in the need for more binder, thus increasing costs and the possibility of blows. However, a study of the sands available and sufficient experimentation has provided core mixtures requiring no more than the usual amounts of binders, and highly satisfactory finishes.

Poor packing or ramming of cores results in poor casting surfaces. There is no substitute for proper ramming, and no other steps to improve finish should be considered until satisfactory ramming is attained.

Watch Gating System Cores

Dirt appearing in cope surfaces of castings can often be attributed to core failures. This dirt may be the result of a core cut or scab, and where this is the case it is not difficult to uncover the source. However, if the dirt occurs with no appearance of a source, it is wise to examine the sprue of the casting at points where cores are used in the gating system. It is sometimes found that skim cores or gate cores may be disintegrating under the flow of metal.

While gate cores may appear insignificant, upon failure they suddenly become the most important core in a mold. It has been found that the best insurance against this possibility is the use of silica flour in the sand mixtures for gate cores, especially for heavy work.

Dimensional Variations in Castings: The manner in which a core is handled may result in variations in casting dimension. Where this dimensional instability is great enough it may cause misruns or crushes, which are easily identified in a hard iron casting. However, these discrepancies may be such that a casting will pass hard iron inspection and be broken in finishing operations in an attempt to make it meet the gages, or it may fail in service.

These conditions may be caused by excessive jolting of core boxes after the core is rammed and before the box is drawn, soft ramming, inaccurate dryers, weak sand, core box wear, or careless clamping of core boxes prior to making the core. Where a critical casting dimension is determined by a core, the use of gages or rubbing jigs is mandatory, at least for spot checking.

Shrinks: Cores can sometimes be an indirect cause of shrinks by creating hot spots at undesirable points in castings. It has long been known that the presence of unoxidized core oil may produce an exothermic reaction when it comes in contact with molten metal. Thus cores which have not been properly baked can produce hot spots with, resultant shrinks. It has been found that this heat-producing reaction does not occur in resin-bonded sands, which normally require less baking than oil-bonded sands.

Proper baking of oil cores in dryers requires that the vent holes placed in the dryer at the core surface have access to the atmosphere. Sometimes in the making of a dryer the core supporting surface is well vented with holes, but access of these holes to the atmosphere is prevented by the side and end supporting members of the drier. If some means of introducing air to the surface of the core in a form dryer is not available, that surface of the core can never be properly baked.

Misruns and Cold Shuts: It has been said many times that molten metal will not "lay to" a hard, smooth core surface, and that hard core surfaces will result in misruns and cold shuts. A reasonable explanation would seem to be that hard, smooth cores usually have a high ratio of gas content to volume of core. When this gas is freed by the action of heat, the pressure produced in the mold and the chilling action of the gases must certainly affect the ability of the iron to run the casting.

Unfortunately, we often have this condition in castings for which the cores are intricate and easily broken. Where this is the case, the binder content of the cores is increased to overcome excessive core breakage. When this is done, the conditions affecting misruns are naturally worsened. If possible, cores for such castings should be made as weak as possible and handled with great care. Scrap cores are undesirable, but less so than scrap castings.

Sometimes this condition is improved by roughening cores with sand paper. This results in rough surfaces and is not a satisfactory solution where surface finish is important.

Where core conditions as above cannot be conveniently obtained, the foundry should consider improved gating as a means of bettering the condition.

Summary

The author realizes that the foregoing discussion of defects resulting from cores is not complete, although the more common ones are presented. It is suggested that the following steps be taken in attempts to reduce losses resulting from core practices.

1. A systematic method of scrap analysis should be set up and should include the coreroom personnel. Failure to accurately determine causes or possible causes of scrap only confuses the situation.

2. Possible causes of defectives should be investigated and positive causes determined.

3. A practice should be decided upon for correction of the defects.

4. After the above steps are taken, the most important phase follows—controlling the practice within the limits necessary to eliminate scrap. In core departments where the supervisory load is heavy, it may be desirable to install personnel whose primary function is that of controlling quality. Any aid to visual inspection of cores such as gages, jigs, and testing equipment should be available for quality control.

Acknowledgment

The material herein presented is not entirely original with the author, but has become available through the work of many foundrymen whom the author desires to acknowledge.

DISCUSSION

Chairman: E. C. ZIRZOW, Deere & Co., Moline, Ill.

Co-Chairman: F. J. Pearr, Lake City Malleable Co., Cleveland. Wm. Farrell: I wonder why it is that so often it is thought that cores and molding sand are the cause of hot tears. I rather think it is neither one. Hot tears are unknown in our plant. It used to be a troublesome problem with us until we installed moisture control. Then the hot tears disappeared.

CHAIRMAN ZIRZOW: A few years ago we tried to produce hot tears experimentally in a casting using hard cores. We used a T-bar with a notch in the center and made a reinforced core by using rods. We were not able to produce hot tears in this particular casting. The only way you could produce a hot tear in this casting was by using steel bars in place of the cores. The steel bars would prevent any contraction which would produce hot tears.

¹ Vice President, Auto Specialties Mfg. Co., St. Joseph, Mich.

DIMENSIONAL CHECKING AND PRESSURE TESTING OF GRAY IRON CASTINGS

By

Kenneth M. Smith*

Introduction

GRAY INON CASTINGS which must be machined accurately on a number of surfaces should be dimensionally checked at the foundry to insure minimum scrap loss in the machine shop from foundry molding and core setting variations and from casting pouring and cooling variations. Similarly, castings which must be liquid and gas tight should be pressure tested in the foundry to reduce losses of machine shop time. Dimensional checking and pressure testing also provide convenient indications of undesired variations in foundry practices before any large scrap loss has taken place in the foundry.

Dimensional Checking

The first reason for dimensionally checking a casting is to be certain that the pattern and core boxes will produce a casting with the proper dimensions. Since casting dimensional standards are determined by the type of service and market for which the castings are intended, casting variations which can be compensated for in a very small lot of castings assume a major cost importance if thousands of the same casting are to be made. Sample castings require a complete layout on a surface plate both for checking the pattern equipment and for checking the inspection fixture which may be used for routine check-

ing of additional castings. Figure 1 shows an engine cylinder block which has been set up on a surface plate in preparation for dimensional checking.

Sample castings with inaccessible inner walls must be sawed through at the proper places to check pattern and core box accuracy and core assembly accuracy. Figure 2 shows a cylinder head which has been sawed for checking internal wall thicknesses and passageway contours. In high production operations it is advisable to saw up periodically sample compli-

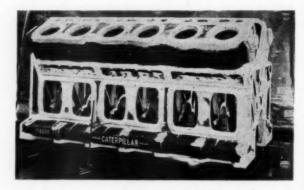


Fig. 1-Surface plate with engine cylinder block set up for dimensional checking.



Fig. 2—Cylinder head casting sawed to show internal wall sections.

tio di

in

to

^{*} Foundry Engineer, Caterpillar Tractor Co., Peoria, Ill.



Fig. 3—Typical fixture for dimensional checking of an unmachined casting.

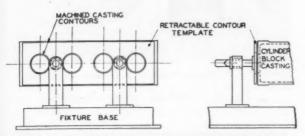


Fig. 4-Contour gage detail for checking fixture.

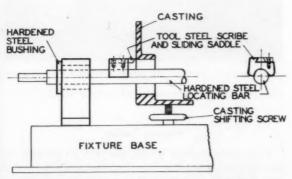


Fig. 5-Locating bar and bore scribing gage for checking fixture.

cated castings in order to detect pattern or core box wear or foundry practice variations. The wall sections of simple castings are easily checked with indicating calipers.

The most important reason for dimensional checking of castings in the foundry is to make sure that sufficient finish allowance is available for all surfaces to be machined. Since molding, core setting, pour-

ing, and casting cooling operations produce dimensional variations, the checking operation can be done most conveniently with individually designed permanent checking fixtures in which the casting can easily be shifted to the proper location which will provide the most uniform machining allowances. After the proper casting location has been established in the checking fixture, bosses which have been provided on the casting in the proper locations to support it in the machining fixtures are ground or milled to the proper height. The locating bosses can be hand ground to the proper height as shown in Fig. 3 where simple swing-type gage arms are used to establish the proper boss height, or they can be machined to the proper height by individual end mills each held in a spindle mounted in an eccentric bushing which can be turned by a hand lever to swing the end mill across the face of the locating boss. The end mill assemblies should be mounted in rugged brackets which can be swung aside when the casting is placed in the fixture. The end mills can be driven by a portable air or electric motor.

A typical fixture for the dimensional checking of castings on a production basis is shown in Fig. 3. These dimensional checking fixtures may seem quite complicated, but fortunately they are built up of quite simple basic elements. The first requirement for a good checking fixture is a base rigid enough to support itself and the casting without deflection when the base is supported on only three corners. A box with a flat machined top and multiple internal diagonal ribs makes an excellent rigid fixture base which also provides a convenient reference surface for set-

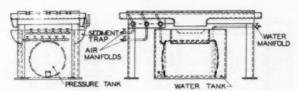


Fig. 6-Pressure testing table details.

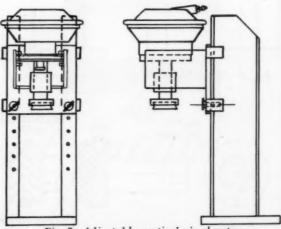


Fig. 7-Adjustable vertical air clamp.

ting up the fixture components. All brackets carrying gaging elements should be rigidly constructed.

Finish allowances on many casting surfaces can be checked by contour gage plates (Fig. 4) mounted on substantial plungers so the gages can be pushed up to the casting profile. For surfaces difficult to reach two identical contour gage plates can be mounted with spacers between them so the contour plates become line up sight gages.

For checking internal bore finishes on castings such as engine cylinder blocks and transmission cases a reference bar can be inserted through the casting after it is placed in the fixture. Scribing gages with locating shoes can then be rotated against this bar (Fig. 5) for checking bore finishes. Scribing gages should have renewable scribing elements so a clear cut scribe line can always be made. In some cases short retractable locating bars can have permanent swinging scribe gages attached to the unsupported end of the bars.

Both cam and screw type casting lifting and maneuvering devices can be used for positioning and holding the castings in the fixtures. Cam type devices must be self locking or have friction locks to prevent casting movement during milling or grinding of the locating pads. The use of ample size hardened steel hinge pins, bushings, work rests, and checking bars will provide maximum accuracy and wear life for all moving parts.

Accurate, rugged dimensional inspection fixtures in the foundry will insure the minimum casting loss from foundry practice variations, and they will also provide a convenient check on these variations.

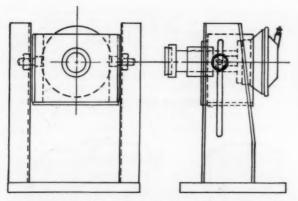


Fig. 8-Adjustable angle and horizontal air clamp.

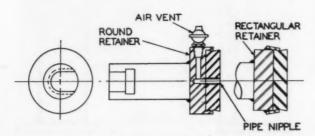


Fig. 9-Extension and sealing pad details.



Courtesy Mechanical Products Corp., Chicago.



Fig. 10-T-head self sealing test plugs.

Courtesy Mechanical Products Corp., Chicago.

to th de to ju in

Fig. 11-Air cylinder expanded rubber sealing plugs.

Pressure Testing

Gray iron castings which are used to contain or convey liquids or gases can be pressure tested by several methods. The method is determined chiefly by the quantity and variety of castings to be pressure tested. Where production rates are high, specially designed immersion type machines having the following typical cycle are frequently used: load casting into fixture, clamp casting and seal all casting openings, apply air pressure to inside of casting, submerge casting in water, check for air leakage through casting, remove fixture from water, unclamp casting, and remove casting from fixture. These special machines may pressure test one large casting at a time, or multiple fixtures can be mounted on a wheel which carries the casting under air pressure into the water tank and back to the loading station. These special machines frequently are used in the machine shop for testing castings after they are fully machined rather than for testing the unmachined castings before they leave the foundry.

Immersion type testing machines have the problem of illumination of the casting when submerged plus the fact that it is difficult to detect small leakage rates on complicated castings especially those which have many sealed openings.

A more sensitive method of pressure testing consists of sealing the casting openings before filling the interior of the casting with water under pressure. The air trapped in the casting is vented as the casting is filled with water. The part is kept under pressure sufficiently long for the liquid to pass through any porous areas. A careful visual inspection of the casting surfaces will disclose very slow leakages through the casting walls.

Typical details of pressure testing equipment can best be explained by describing an internal water pressure testing system designed to accommodate a wide variety of simple and complicated unmachined castings using the minimum number of special fix-

Castings are pressure tested in fixtures mounted on a 34-in. wide by 80-in. long cast iron table (Fig. 6) which has lengthwise T-slots at 61/4-in. centers machined into its surface to permit clamping the fixtures to the table in any desired arrangement. A water drain trough runs around the outside of the table and discharges to a sediment trap mounted on the side of an open top rectangular water tank below the table. Submerged in the open top tank is a 30-gallon pressure tank which fills with water from the open top tank when air pressure is released from the pressure tank. After a casting is properly clamped in the pressure testing fixtures, compressed air at any desired pressure up to full line pressure is admitted to the top of the water pressure tank through an adjustable air pressure regulator. The water trapped in the pressure tank by a check valve on the tank inlet is carried by hoses to the pressure testing

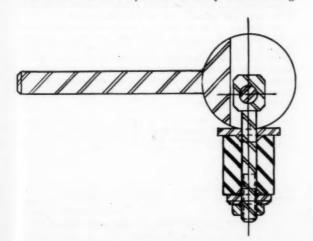


Fig. 12-Cam-expanded rubber sealing plug.

fixtures where the water drives trapped air out of the casting through the testing fixture air cocks located at the high points of the casting. When solid streams of water flow out of the air cocks, they are closed and the casting is held under pressure a sufficient length of time to disclose any leakage through the casting. After the casting surface is thoroughly checked for leaks, the water pressure is released, and the casting is removed from the fixture.

The casting clamping and support fixtures are designed for the interchangeable use of as many fixture parts as possible. In most cases casting clamping actions and sealing pad motions are provided by air diaphragm units of the type used for air brakes on highway truck trailers. Five sizes of these air diaphragms having from 12 to 36 sq in. of effective diaphragm area are used. All sizes of the air diaphragms are mounted on the same size nose plate and have the same size T-nut mounted on their stem. This permits the easy insertion into the fixtures of the proper sizes air diaphragms needed to supply the clamping pressure on any casting. The operating air for these diaphragms is supplied by two manifolds each having six outlets and its own control valve and pressure regulator. Two manifolds permit the separation of vertical and horizontal clamping motions. All air and water hoses are provided with quick disconnect couplings on both ends.

Part of the fixtures are provided with mounting slots for the air diaphragms, but adjustable vertical clamps (Fig. 7) or adjustable angle or horizontal clamps (Fig. 8) are used wherever possible. The actual

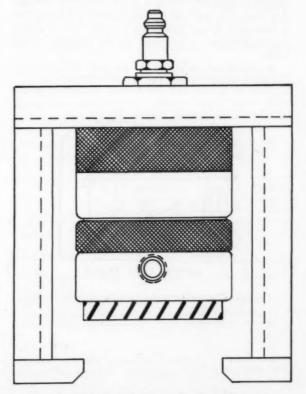


Fig. 13-Air-operated flange clamp sealing pad.

casting clamping and sealing is done with extensions which slide through the guide bushings of the vertical or horizontal clamps or fixture bases. One end of the extensions is machined to match the air diaphragm T-nut so the spring in the diaphragm unit will retract the extension, and the other end of the extension carries either pressure fingers or sealing pad retainers. These sealing pad retainers (Fig. 9) have a 3/8-in. deep recess, and are made with a 1/16-in. taper on each side of the recess to grip the 3/4-in. thick gum rubber sealing pads. The rubber sealing pads may have a circular, rectangular, or irregular shape to match the dimensions at the bottom of the retainer recess. Short pipe nipples are inserted through the rubber sealing pads for both water inlet and air vent connections to prevent collapsing the hole in the rubber sealing pad when it is clamped against the casting.

Many pressure testing set-ups are simplified by using T-head self sealing test plugs (Fig. 10); air cylinder expanded rubber sealing plugs (Fig. 11); cam expanded rubber sealing plugs (Fig. 12); and air operated flange clamp sealing pads (Fig. 13) which use air cylinders identical with the ones shown in

Fig. 11.

The casting support fixtures are individually designed with supports, locating stops, sealing pads, water inlets, and air vents as needed. Wherever possible the support fixtures are designed for testing a group of similar parts by the use of adaptors and adjustable fixture parts. A series of several sizes of flat rubber sealing plates provided with adjustable locator fingers is used for testing a great variety of castings. For clamping large flat castings against a sealing plate a fabricated bridge which holds up to ten air clamping diaphragms is mounted on the pressure testing table.

A typical fixture set-up for pressure testing a pipe casting is shown in Fig. 14.

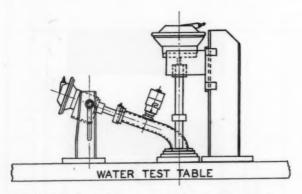


Fig. 14-Typical pressure testing fixture set-up.

Many of the above methods of construction can be used in designing quite simple single purpose pressure testing fixtures and machines. For example, air or liquid passages in large castings can be pressure tested in quite simple self-contained fixtures by the addition of a drain pan and gravity filled pressure tank to the fixture. When this is done an air cock is used between the drain pan and pressure tank for manual control of the water return to the pressure tank in order to avoid trouble from sediment in a check valve.

Air-operated arbor presses or jolt-squeeze molding machines are quite readily convertible to specialty pressure testing machines by adding a drain pan and a pressure tank. Converted jolt-squeeze molding machines are very useful for pressure testing many flat bottom parts such as cylinder heads because the casting sealing pads can rise up through specially designed roller conveyor tables which permit the installation of the machine in or next to roller conveyor lines.

Pressure testing of castings in the foundry provides a convenient check on foundry practices, and saves the machine shop time which would otherwise be wasted on defective castings.

DISCUSSION

Chairman: James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman: H. W. JOHNSON, Wells Mfg. Co. Skokie, Ill. H. H. Lurie: ¹ How frequently do you cut up castings for gaging critical sections to determine whether there is any wear

on the patterns?

F. W. Shippey: 2 That is rather difficult to answer. Our Inspection Dept. is quite large and castings are constantly being layed out for dimensions and cut up for examination of section size. So it would be rather difficult to say what per cent of the castings are cut up or how often it is done. But I can assure you that as soon as any indication of pattern wear becomes evident the Inspection Dept. will find it quickly. In addition to this, the quality control group in the foundry itself constantly breaks up scrap castings to determine the causes of defects and at that time any appreciable pattern wear would also be discovered.

Co-Chairman Johnson: What is the maximum water pressure

you use for testing these castings?

MR. SHIPLEY: We normally use 40 to 50 lb pressure on our water test machines when we check for pressure leaks.

Co-Chairman Johnson: Have you ever found that castings which showed no leak when tested at high pressure, would show leaks under less pressure due to a sealing action?

MR. SHIPLEY: I do not think so because we use plain water.

MR. LURIE: We have an aluminum pan which sometimes does not leak during water pressure testing, but due to the hot oil in the engine it will open up when in operation. Hot oil seems to go through where water will not.

MR. SHIPLEY: We have castings which we pressure test with kerosene and not water; kerosene will go through in instances

where water may not.

¹ Chief Metallurgist, Cummins Engine Co., Columbus, Ind.

² Foundry Manager, Caterpillar Tractor Co., Peoria.

CHOOSING EQUIPMENT FOR NONDESTRUCTIVE TESTING*

By Carlton H. Hastings**

Introduction

CHOICE OF NONDESTRUCTIVE TEST EQUIPMENT for practical application in the production foundry involves consideration of four principal tests: 1. Radiography, 2. Magnetic particle tests, 3. Penetrant tests, and, 4. Ultrasonic or acoustical tests. Although there are many other types of nondestructive tests for metallic materials, some of which may have limited application to the examination of castings, the four methods just mentioned have been most widely used and accepted. It is the purpose of this paper to discuss first the important aspects of applicability of these four methods in particular, and subsequently of nondestructive tests in general, with regard to the choice of test equipment.

Principal Methods

Let us consider the principal methods of test in the order listed above. They are all tests used to determine soundness of castings.

Radiography

IG:

be esair he re is or

a

19

id

at at

n

)e

ar

g

n

e

11

ıt

1

Choice of radiographic test equipment to suit foundry application should be predicated upon a knowledge of (a) the penetrating power required of the X- or gamma-radiation source; (b) the degree of flaw detectability desired; and (c) the portability or convenience of use associated with the equipment as it relates to the particular foundry production operations involved.

Penetrating Power—With regard to penetrating power and thickness limitations of the radiographic method of test, there are very few X-ray units available today for the examination of steel sections greater than 6 in. Depending upon the length of exposure time which is deemed feasible or economical, one might establish arbitrary thickness limits for X-ray examination of castings as shown in Table 1. Photographic recording techniques only are considered in this table.

In addition to the above limitations it should be noted that radiography employing radium or the radioactive isotope Cobalt⁶⁰ as a source requires greater exposure times for equivalent steel thickness.

A fair comparison cannot be made between these gamma-ray sources and X-ray units because of this fact. If one neglects to consider the exposure times, which, incidentally, are usually measured in hours or days (sometimes up to one week) for radium and cobalt⁶⁰ sources rather than in minutes as for X-ray sources, then an arbitrary thickness limitation based on picture quality alone would be about 8 to 10 in. of steel for gamma-ray exposures.

Considering aluminum alloys, the thickness which can be satisfactorily radiographed using 250-kv equipment is about 4 in. Since copper-base alloys are more opaque to penetrating radiations than steel, the thickness limitations are correspondingly decreased. Thus, where $1\frac{1}{2}$ in. of steel can be readily examined with 250-kv equipment, the corresponding limit for brass is about $1\frac{1}{8}$ in.

Flaw Detectability—One of the technical details of the radiographic method of test, which is often ignored, and which should have a considerable influence on the selection of radiographic equipment for

TABLE 1-MAXIMUM THICKNESS OF STEEL TO BE RADIOGRAPHED WITH X-RAY TUBES OF A GIVEN KILOVOLTAGET

	11/4
	1.1/4
	3
	4
	6
	14*
	20*
reens with	reens with long e

^{*} The statements or opinions expressed in this paper are those of the author and do not necessarily represent the views of the Ordnance Corps.

^{**} Physicist, Watertown Arsenal, Watertown, Mass.

[†] From Ordnance Inspection Handbook on Radiography, ORD-M608-3, Ordnance Department, U. S. Army, May 1949.

the foundry, is its variable sensitivity to defects. It does not seem worth-while to require an inspection of any type and to subsequently tolerate a testing technique which defeats the purpose of the inspection.

In other words, radiographic equipment and techniques should be chosen with the objective of detecting flaws, not with the objective of merely blackening the film. Although a large part of the variability of the sensitivity of the radiographic method can be attributed to the incorrect handling of equipment, which is itself capable of producing good results, the choice of the equipment for particular jobs is often to be blamed. As an example of this situation, the writer is acquainted with an inspection problem involving a steel casting of about 3/4-in. thickness. Probably for economic reasons, the foundry performing the required radiography chose to rent a radium capsule. It was found that satisfactory radiographs could not be made with such a source on this particular casting. This finding was based on the appearance of the penetrameter image on the radiograph. Satisfactory radiographs will reveal clearly the details of a penetrameter test object placed on the casting, so as to cast its image onto the film. In this instance, the details of the penetrameter image could not be seen, thereby indicating an inferior radiograph.

It is difficult to get adequate defect detectability on thicknesses of steel less than 1 in., or light alloys less than 4 in. in thickness when using high voltage X-rays, radium, or cobalt⁶⁰ as a source. As a general rule, when considering equipment selection for a particular job, the lowest kilovoltage X-rays which can form a picture in 15 min or less should be employed. In following this rule, one should remember that radium and cobalt⁶⁰ are considered to be roughly equivalent to a 2,000-kv X-ray machine. In the example just described, a 250-kv X-ray machine should have been used for satisfactory radiography.

In addition to considering the kilovoltage of the



Fig. 1—This unit (150 kv production type) is ideally suited for inspecting thousands of small parts per day.

Cabinet is completely rayproof.

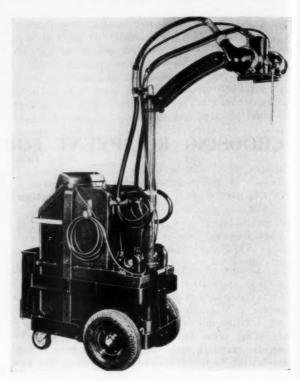


Fig. 2—This unit (150 kv portable x-ray type) is well adapted for use in a shop or on location to radiograph welds in steel plate up to 1 in. thick. The cooling system is self-contained and the unit will operate wherever a 220-volt power source is available.

source when choosing equipment, careful attention should be given to the selection of film. If only gross defects are sought, most any industrial radiographic film which permits reasonable exposure time can be made to do the job. However, if it is required to detect cracks or some of the more elusive defects, such as microshrinkage in magnesium castings or fine shrinkage in cast bronze or steel, then high-contrast, fine-grained radiographic films must be employed.

It should be pointed out that deviations from the above principles may yield films which are of no use as far as detecting flaws is concerned. In such cases, the radiographer is only going through the motions of making a radiograph, will obtain a meaningless picture, and is literally wasting his time.

Portability or Convenience of Use—Of the three factors upon which a choice of radiographic equipment depends, portability or convenience rightfully belongs at the bottom of the list. Although this factor is important to the economic and expeditious use of radiography, its importance must be subordinate to the consideration of (1) equipment capable of penetrating the casting and (2) equipment capable of forming a good picture of the casting.

Figures 1 to 5 are intended to illustrate various types of X-ray equipment designed for convenience in different applications. Figure 6 shows for comparison a radium "pill." Its small size and related porta-

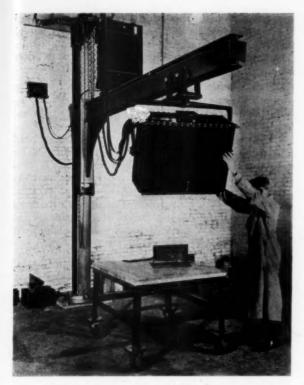


Fig. 3—This is a 250 kv x-ray type unit on jib crane. Entire high voltage system and tube are enclosed in one oil-filled housing. Water cooler is mounted on wall. Control panel is outside of exposure room. Jib crane mounting provides easy manipulation of tube for a variety of jobs.

bility is the strong point in favor of this type of source.

Magnetic Particle Tests

Whereas the radiographic method of test is applicable to all cast metals, the magnetic particle tests for soundness can be successfully applied only to castings of ferromagnetic alloys such as the ferritic cast irons and steels. The austenitic alloys of iron are not ferromagnetic and cannot be tested successfully by magnetic particle methods. The choice of magnetic particle test equipment should be based upon (a) the surface condition of the casting; (b) the degree of flaw detectability required; (c) the possibility of injuring the casting by overheating or burning the metal; and (d) the convenience of operation and compatability with other production equipment.

Surface Conditions—If the casting has the usual ascast finish (fairly rough) the dry-powder technique is advisable since the light oil suspension of magnetic particles employed in the wet technique is difficult to apply and interpret properly under these conditions. An exceptionally smooth as-cast surface or the machined surfaces of castings may be examined with the wet technique, if greater sensitivity to fine cracks is desired. Scale, paint or oxide coatings on the surface of a casting may require the use of a higher voltage

associated with the magnetizing equipment than clean or sand blasted surfaces. This situation may be encountered when using direct magnetizing techniques (passing the magnetizing current directly through the casting by means of electrical contacts). It should be emphasized, however, that the use of the higher voltages (50 volts or more) in order to arc through such surface coatings may seriously injure the surface of the casting. Adequate surface cleaning and the use of low voltage equipment (6 to 12 volts) is to be preferred. Precautions against permanent injury of the type just mentioned will be discussed at greater length in a subsequent paragraph.

Flaw Detectability Required—Beginning again with a discussion of the magnetic particles themselves, the greatest sensitivity to small surface defects, such as quench and grinding cracks, is afforded by use of the wet suspension of particles in a light oil such as kerosene. Depending upon the size of the castings to be examined, the suspension may be applied by spraying from a hand or machine-operated spray gun, pumping from a supply reservoir through a hose and nozzle or by immersion of the casting in a gently agitated bath of the suspension.

The dry-powder technique provides greater sensitivity to more deeply seated gross defects such as shrinkage cavities, elongated gas cavities, inclusions, and hot tears. Although the magnetic particle testing method is not held entirely reliable for the detection of subsurface flaws, it should be clearly understood that the method is capable of doing a great deal in this direction. One should not expect the method to detect all subsurface flaws but if they are of interest to the inspector, use of the dry-powder technique will provide real information in the hands of skilled personnel.

Dry magnetic powder may be applied from either hand or pneumatic applicators with about equally good results, the difference being largely one of convenience. With regard to the selection of magnetic

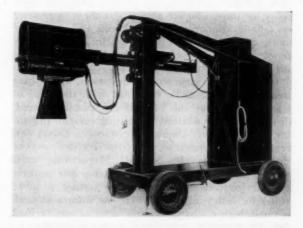


Fig. 4—This 260 kv mobile x-ray unit is completely self-contained and requires only a 220-volt power supply to operate. Operator's booth is lead-lined, but all other personnel must be kept at least 50 ft away while unit is operating.

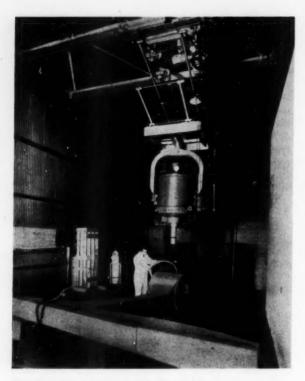


Fig. 5—This is a 1,000 kv resonance-type x-ray generator suspended from bridge crane by a pantagraph device which prevents sway. Compare size of unit with older 400-kv unit shown in crate at left.

particles, the writer considers the pastes or particles specially prepared for use with the wet technique most convenient and feasible when this technique is indicated. Specially prepared dry powders, available commercially for this test method, offer the advantages of several colors for ease of visibility on scaled as-cast surfaces, or sand-blasted finishes. Other finely divided iron powders, however, can provide adequate sensitivity for those who are able to test them in practice on samples containing known defects and may prove more economical than the specially prepared dry powders when a sufficient volume of inspection testing must be accomplished.

With regard to magnetizing equipment, when heavy steel castings are to be examined with the hope of learning whatever the method can possibly reveal, direct current or rectified alternating current (pulsating direct current) should be employed. Direct current welding generators, dry disc rectifier sets or similar units may be utilized. Magnetizing units should preferably supply up to about 800 amperes current (continuous rating) at low voltage (6 to 12 volts) for heavy castings. Where the volume of work is small, high discharge storage batteries may be used but are generally considered to be a nuisance to maintain in a fully-charged condition. For small casting of section thicknesses up to 1/2 in., sources supplying currents up to about 350 to 400 amperes should be adequate. Where only surface flaws are sought, alternating current sources may be employed. For alloys having a high magnetic retentivity and when the residual method of test is indicated as acceptable, short duty cycle magnetizing units supplying up to 5,000 amperes direct or alternating current are often employed.

When considering magnetizing techniques and the equipment associated therewith, it is believed worthwhile to mention that with the exception of the so-called central conductor technique, cable wrapping or solenoid (indirect magnetization) techniques are held inferior in sensitivity to direct magnetization. The direct methods require some form of electrical contact plate or electrodes while the indirect methods tend to require varying quantities of flexible cable. All cables must be of sufficient current capacity to handle the heavy magnetizing currents without excessive heating. Welding cable is commonly employed for this purpose.

Injury by Overheating-Certain precautions involving special equipment features must be taken when employing the magnetic particle test on finished surfaces or on the surface of air-hardening alloys. The high currents employed for magnetization may cause injury due to heating of the metal or by arcing so as to cause pits on machined surfaces. In the case of airhardening steels, the local overheating which can occur at the contacts when direct magnetization is employed, may even initiate small quench cracks. Such difficulties may be avoided by using larger contacts made of soft copper gauze. Contact electrodes of this type are available commercially or may be improvised by attaching a stub of welding cable to the ends of the copper contact electrodes or prods usually employed. This type of soft contact material decreases the danger of arcing and the possibility of obtaining burned or hot spots on the surface of the casting.

Convenience of Operation—The large volume of heavy castings inspection is accomplished using port-

tı

11

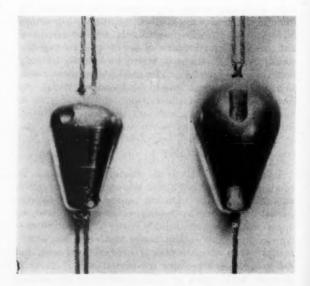


Fig. 6-Radium pills for radiographic inspection.

0

d



Fig. 7—Bench-type magnetic particle inspection unit for wet method examination of small castings.

able rectifier units and contact prods for magnetization. Such mobile equipment can be moved about the shop readily for inspection at the scene of other operations. If smaller castings, which can be easily manipulated by hand are to be examined, more convenient bench-type inspection units are available. A typical unit of this form is shown in Fig. 7. If a great volume of production of small castings must be inspected by this method it might be profitable to investigate the procurement of a special inspection machine designed to handle the particular product. Special equipment has been constructed for the rapid testing of aircraft cylinders for reciprocating engines.

Penetrant Tests

Penetrant tests include a variety of old and new versions of the so-called "Oil and Whiting Method." Many of the newer techniques are patented and materials for their accomplishment sold under numerous trade names. The choice of equipment or materials for implementation of penetrant tests seems to be mostly dependent upon convenience of use rather than upon any specific technical factors. Because of the fact that this group of tests involves patented techniques and because their choice involves mostly arbitrary decisions pertinent to particular types of casttings, it is difficult to make any statements that would apply to castings, generally. It is therefore suggested that prospective users of this method request a demonstration of the materials and equipment for the particular inspection problem involved. The writer believes that ordinary good judgment and a familiarity with the particular foundry application are the only requisites for a proper choice in this instance.

Ultrasonic or Acoustical Tests

When applicable to the examination of castings, the ultrasonic techniques may be employed to measure the thickness of metal involved, even if only one side of the metal is accessible, as well as to detect unsound conditions.

A wide range of equipment is not currently available for ultrasonic testing because of the newness of the method and the patent restrictions. The choice which is presented, however, concerns equipment primarily employed for thickness measurement. The echo techniques are to be preferred at the present time for thicknesses of over 4 in. For the measurement of thinner sections, and particularly for those thicknesses less than 1 in., the so-called resonance technique, employing different apparatus than the echo techniques, is more desirable. Considering the resonance technique for thickness measurements under 4 in., there are a number of available instruments. These are best selected on the basis of convenience of operation, portability, etc. At least one instrument is constructed for operation from self-contained batteries, and is therefore independent of power lines. This unit provides an audible indication of proper adjustment by means of a pair of earphones, leaving the hands of the operator free for other tasks. Other types of ultrasonic resonance thickness meters employ a calibrated cathode-ray oscilloscope for visual observation of data.

For the detection of flaws in castings, the echo technique and its associated equipment is the more versatile.

There are a large number of inspection problems, associated with the testing of various shapes and sizes of castings, which cannot be approached with ultrasonic techniques at all, considering the present state of the art. A complete discussion of the reasons why the method may or may not apply is considered to be beyond the scope of this paper. It should be pointed out that, in general, crystals of lower fundamental frequency should be employed for detecting flaws by reflection or echo techniques in castings in order to overcome the acoustic opacity associated with dendritic, coarse-grained, or porous materials. This acoustic opacity, which may prevent the successful application of ultrasonic methods to castings, becomes progressively greater at higher frequencies.

In view of the foregoing statements, it is strongly recommended that ultrasonic equipment be demonstrated to be satisfactory for the particular inspection problem by actual trial prior to purchase.

Nondestructive Methods-General

Since the test equipment or other apparatus represents only physical implementation of test methods, information of the type presented in the previous parts of this paper can only serve as a rather poor substitute for an adequate understanding of the various test methods themselves. In order to increase the usefulness of the information which has been presented, it is believed desirable to discuss some important factors associated with the correct selection and application of nondestructive test methods.

It is difficult to present all points of view regarding a subject. It is recognized also that the problem associated with inspection methods in the foundry must be considered from the point of view of the purchaser and the producer.

The purchaser's problem is to specify inspection

methods or technique properly and completely. It is the producer's problem to interpret these specification requirements in terms of production and cost considerations. Actually, the most serious inspection difficulties are due to incorrect selection or application of test methods rather than test equipment. Difficulties of this type can only be avoided by strict vigilance on the part of both purchaser and producer. Although the following factors are discussed from the point of view of the purchaser, or actually the specification writer, the same factors should be considered by the producer during his study of specification requirements prior to starting production.

1. The engineer responsible for the selection of one or more nondestructive inspection test methods for the examination of a particular casting design should have a sound knowledge of the physical principles underlying the available methods. It is often the case that personnel preparing or applying specifications do not have the time or the background of training necessary for them to acquire this basic knowledge. As an alternative to being expert in the principles of nondestructive testing, the engineer should require that methods and equipment be demonstrated, by a nondestructive testing specialist or the test equipment vendor, if he is qualified, to perform satisfactorily on the particular inspection problem. Satisfactory performance primarily involves the reliable evaluation of some chosen characteristic of the material being tested.

2. The chosen characteristic mentioned in (1) above must be proven to be (a) a necessary quality characteristic for satisfactory service performance, or (b) some other characteristic which is measurable by nondestructive techniques, and which has been proven to correlate with a necessary quality characteristic. The correct choice of a proven correlation cannot be overemphasized, and should be made through cooperative effort between competent design and materials engineers.

3. The method should be synchronized, insofar as it is possible to do so, with the manufacturing operation. This is necessary if the greatest economic advantage is to be realized from the nondestructive nature of these tests. Synchronization, as it is here referred to, implies the application of the method as

early in the manufacturing process as is feasible. For example, methods of detecting subsurface or internal flaws should be applied as soon after breaking the casting out of the mold as is convenient for handling. Thus rejectable material will be detected before any subsequent labor can be wasted thereon. Surface inspection methods should be applied only to surfaces which are significant in the finished product. There is no point in detecting surface flaws on a rough casting and subsequently machining to provide a new surface that must also be examined. The surface examination in such a case should be withheld until the final approximate surface is established.

Probably, the greatest errors of commission or omission in inspection operations do not directly concern the test methods or the equipment, but frequently reflect on them in such a way that either the method or the equipment is held unreliable. These errors revolve around the decisions regarding acceptance or rejection of material examined. Frequently, acceptance standards which can be used as a guide in making the decisions based on test results, are entirely forgotten in the process of writing a specification. If not forgotten, standards which have no basis founded on a required quality characteristic, may be supplied. In the former case, test equipment operators, not sufficiently qualified to do so, are forced to make "on the spot" decisions regarding the acceptability. In the latter case, unjustified rejections or dangerous acceptances may result. The fault, in either case, is the failure to recognize that the setting of acceptance limits or the making of decisions in the absence of established standards is a high level engineering problem, and must be handled by expertly qualified engineers who are familiar with every important detail of service performance and the product quality required thereby.

The mishandling of reliable inspection data or the procurement of reliable but irrelevant data because of an incorrect choice of test method should not be blamed on the method or its associated equipment. It is believed that the careful consideration of the factors discussed in this paper can greatly reduce the expense of wasted man-hours and unnecessary rejections, which can result from the misuse or improper choice of nondestructive testing equipment.

W

C

of

ve

in fir me ve pr

tha

to

PRECOAT MATERIALS FOR INVESTMENT CASTING

By

Wm. F. Davenport, 1st Lt. USAF and Adolph Strott*

ABSTRACT

This project was initiated to evaluate the more plentiful ceramic oxides which offer a possibility of use as a precoating material in the investment casting process. Various ceramic oxides, alumina, silica, zirconium silicate, and magnesia were investigated as possible precoating materials for the precision casting process. These materials were evaluated by casting a heat-resistant alloy, a stainless steel, and an alloy steel into the precoated mold. The castings were examined to determine the best precoating material for each alloy cast. The precoats were evaluated on the basis of precoat adherence, surface smoothness, and degree of oxidation. The most satisfactory materials for use as precoats were found to be of the alumina-silica composition.

Introduction

Production of smooth surfaces is one of the more pronounced advantages of the precision investment casting process. To obtain a smooth surface on the casting, the mold material in contact with the cast metal must have a very fine particle size. There are two ways of insuring that the material lining the cavity will be of sufficient fineness to produce a smooth surface: (1) The use of an investment material which is composed wholly of fine particles, or (2) Coating of the disposable pattern with a thin layer of fine particles before surrounding it with an investment.

At the present time the authors know of no one-step investment, one which is composed entirely of very fine particles, that is used commercially for alloys melting over 2000 F. Until such an investment is developed it will remain necessary to precoat patterns prior to investing.

At the present time, silica flour is the most commonly used precoating material. However it has certain disadvantages, the most important of which is that it reacts with some metals, and has no effect on inhibiting the formation of oxide on the casting.

Therefore, it was felt that it would be worthwhile to undertake a study of possible precoat materials with an aim toward the development of more satisfactory coatings than those in current use.

On the basis of present knowledge it was decided that the ideal precoat would have the following properties: (1) Simplicity in preparation, (2) Economy in material costs, (3) Physical and chemical inertness, (4) Resistance to thermal cracking, and (5) Adherence of reaction products, if any, to the investment rather than the casting.

Experimental Work

The ceramics selected as possible precoat materials were chosen on the basis of refractory properties and reasonableness in cost. The materials finally selected by the above criteria were silica, alumina, zirconium silicate, and magnesia. They were acquired as commercially-milled flours finer than 150 mesh. (See Table 1 for their properties, and sources).

The normal practice used in most commercial shops making precision investment castings of alloys melting over 2000 F was followed very closely. This practice consists of making a wax replica, and then cleaning it in a solvent, such as acetone, by dipping. When the pattern has dried, it is dipped into the previously prepared precoat slurry until a coating of desired thickness is built up. The coating thickness to be used will depend on the producer's previous experience. Some producers prefer a relatively thin coating, two or three thousands of an inch thick, while others prefer a heavy coating three or four hundredths of an inch thick.

Theoretically the precoat must be only thick enough to afford enough strength to withstand the erosion of the metal entering the mold, if the investment surrounding it is sufficiently strong. By strong is meant that the investment is able to resist any deformation which may occur during casting, or previous handling of the mold. Most known investments more than meet this required property. A thin precoat will act physically as the back up investment material, while thick precoats behave as an entity in physical characteristics such as shrinkage, thermal shock resistance, etc. Since this work deals with possible precoat compositions of untried ceramics it was

^{*} Materials Laboratory, Air Materiel Command, Wright-Patterson Air Force Base, Dayton, Ohio.

TABLE 1-MATERIALS USED FOR PRECOAT INVESTIGATION

Ceramics	Apparent Density	Average Grain Size	Delivered By			
Alumina	1.15 g/cc	1.5	I. T. Baker Chemical Co., Phillipsburgh, N. J.			
Silica	1.64 g/cc	5	Ransom and Randolph Co., Toledo, Ohio			
Magnesia	0.55 g/cc	1	City Chemical Corp., New York 11, N. Y.			
Zirconium Silicate	2.70 g/cc	2	,			
Silica Sand	1.62 g/cc	40 mesh	Ransom and Randolph Co., Toledo, Ohio			
Investment	I. C. 711-G1		Ransom and Randolph Co., Toledo, Ohio			
Chemicals			From			
Sodium Silicate	U. S. A. Specificat: Amend1, Class A (AAF Stock No. 8	1	Industrial Distributors, Inc., New York 29, N. Y.			
Polyvinyl Alcohol	Low viscosity—Typ Grade RH-393		E. O. DuPont de Nemours & Co., Niagara Falls. N. Y.			
Casting Alloys						
Stainless Steel 302	Specification AN-S- Composition, G,		USAF Stock			
Alloy Steel	A. I. S. I. 4340		USAF Stock			
Heat Resistant Type Alloy	Approximate Anal	ysis	Scrap from Materials Laboratory Alloy Develop-			
	20% Cr, 40% Ni, 1.0% Mn, Bal. (5% C, .5% Si,	ment Program			

thought best to use a heavy precoat of approximately two hundredths of an inch.

It was known from previous work that the addition of 40-mesh silica sand to the precoat mix as a filler helped in the prevention of shrinkage, and increased the thermal shock resistance. This work consisted of making different percentage additions of 40-mesh silica sand to precoat mixes. The results showed that a 13 per cent by volume addition to the fine material produce the optimum in results from the standpoint of thermal shock resistance, shrinkage and the maintaining of a smooth surface. This sand was added to the dry ceramic flours, and mixed prior to the preparation with the liquid binders of the dip slurry.

It was necessary to conduct some preliminary experiments to determine the most suitable binder for each material to be investigated, for possible use as a precoat. These experiments showed that a 20 deg Baume sodium silicate solution could be used for most of the materials investigated, silica, magnesia, and zirconium silicate as a binder, and it was satisfactory as a suspension agent. However, it was found that when alumina was mixed with sodium silicate solution it coagulated. Therefore another binder had to be found for use with alumina. A 2 per cent polyvinyl alcohol solution was found to be satisfactory

as a binder and suspension agent for it. Further investigation showed that zirconium silicate could also be prepared with polyvinyl alcohol, but silica and magnesia could not, because they coagulated when added to the binder.

The aforementioned results indicate that if alumina is to be prepared with any material other than zirconium silicate it will be necessary to prepare the alumina with polyvinyl alcohol and the other material with sodium silicate separately. The materials after preparation with their different binders may then be mixed together to produce a dip coat slurry. In Table 2 is shown the compositions prepared and evaluated, also the amount of ceramic, filler sand, and binder used for each sample precoat preparation. The amount of binder added was not critical, the governing factor in the addition of binder was the formation of a slurry of the desired consistency. To all the compositions, 13 per cent by volume of sand (40mesh silica) was added prior to mixing with the liquid binder.

The method used in preparing the alumina-silica series of compositions was as follows: First the sand addition of 13 per cent by volume was made to both the alumina and silica. Then the silica portion was mixed with the sodium silicate solution (20 deg Be) and the alumina was mixed with the polyvinyl alcohol

TABLE 3-COMPOSITION OF SILICA-ZIRCONIUM SILICATE-ALUMINA PRECOATS

									Amount of	Ceram	ics, grams			of Binder
No.	% Wt	lica % Vol	Compos Zirconiur % Wt	n Silicate % Vol		mina % Vol	Silica	Sand*				Sand*	Sodium Silicate 20° Be	Polyvinyl Alcohol 2%
61	29.8	33.3	49.0	33.3	21.2	33.3	44.7	9.7	73.5	9.7	31.8	9.7	43.5	33.5
62	24.5	25.0	40.5	25.0	35.0	50.0	36.75	8.0	60.8	8.0	52.5	16.0	35.5	55.0
63	15.4	17.7	59.0	41.2	25.6	41.2	·23.1	5.0	88.5	11.7	38.4	11.7	38.5	40.0
64	13.9	15.0	53.1	35.0	33.0	50.0	20.9	4.5	79.6	10.5	49.5	15.0	34.5	52.0
65	18.0	23.0	69.2	54.0	12.8	23.0	27.0	5.9	103.8	13.7	19.2	5.9	45.5	20.0
Sand	d addition	on to sili	ca conium-sil	icate =	21.7% t 13.2% t	tion to all by weight by weight by weight	materia	l was l	3% by volu	me.				

TABLE 2-COMPOSITION OF PRECOAT MATERIALS INVESTIGATED

													(co	of Binde c) Polyviny
		Composi	tion (%)			Amount o	of Ceramio	s (gran	ns) Zirconium				Silicate 20° Be	Alcoho
io.	Silica	Alumina	Zirconium Silicate	Magnesia	Silica	Sand*	Alum ina	Sand*	Silicate	Sand*	Magnesia	Sand*	Solution	Solutio
_	100		,		100		Alumina (Composi	tions		,			
0	100	100			120	26.1	100	90 8					52	***
0	10	90			12	2.6	120 108	36.5 32.8					5	125
2	20	80			24	5.2	96	29.2					10.5	112
3	30	70			36	7.8	84	25.5					15.5	87.5
4	40	60			48	10.4	72	21.7					21	75
5	50	50			72	15.6	72	21.7					31	50.5
6	60	40			84	18.2	56	17.0					37	47.5
7	70	30			84	18.2	36	10.9					37	37.5
8	80 90	20 10			96 108	20.8 23.4	24 12	7.3					42	25
9	90	10						3.7	asitions				47	12.5
0			100				um Silicate	e Comp	200	26.4			30	34
1	10		90		20	4.3			180	23.8			40	31
2	20		80		38	8.2			152	20.0			39	30
3	30 40		70 60		54	11.7			126	16.6			36	30
5	50		50		68 .80	14.8 17.4			102 80	13.5 10.5			35 35	30 24
6	60		40		90	19.5			60	8.0			39	18
7	70		30	-	98	21.3			42	5.5			43	13
8	80		20		104	22.6			26	3.4			45	8
9	90		10		117	25.4	-;		13	1.7			51	4
				***		Silica - M	agnesia Co	mpositi	ons					
0	10			90	8.5	1.8					75 75	48.8 48.8	150 154	
2	20			80	19	4.1					75	48.8	158	
3	30			70	32	7.0					75	48.8	164	
4	40			60	50	10.8					75	48.8	172	
5	50			50	50	10.8					50	32.5	122	
6	60			40	75	16.3					50	32.5	133	
7	70			30	117	25.4					50	32.5	151	
8	80 90			20 10	120 135	26.0 29.3					30 30	19.5	112	
9	90			10			nomeric Co				30	19.5	119	
0				100	Alui	nina - M	agnesia Co	mpositi	ons		75	48.8	150	
1		10		90			8.5	2.5			75	48.8	150	9
2		20		80			19.0	5.7			75	48.8	150	20
3		30		70			32.5	9.8			75	48.8	150	34
4		40		60			38.5	10.0			50	32.5	100	35
5		50		50			50	15.2			50	32.5	100	52
6		60 70		40 30			45 70	13.7 21.3			30 30	19.5	60	47
8		80		20			60	18.3			15	19.5 9.8	60 30	73 63
9		90		10			67.5	20.5			15	9.8	30	60
					Alumina	- Zircon	ium silicat		ositions					
0		100	10				120	36.5	100				***	125
1		90	10				120	36.5	13.5	1.8			4.5	125
2 3		90 70	20 30				110 95	33.4	27.5	3.6			9	115
4		60	40				80	28.9 24.3	41 53.5	5.4 7.0			13.5 17	99 83.5
5		50	50				60	18.2	60	7.9			19.5	62.5
6		40	60				50	15.2	75	9.9			24	52
7		30	70				40	12.1	93	12.3			30	42
8		20	80			*	30	9.1	120	15.8			39	31
9		10	90				20	6.1	180	23.8			58	21
				40-mesh si	lica) addit	ion to all	mater ials	was 13	% by volu	me				
		dition to			= 21.79	by weight by weight by weight	ht							
		dition to		.111	= 30.49	o by weig	nt							
	no Ade	uillon to	Zirconium	-suicate	= 13.2% = 65%	by weight	gnt							

solution (2 per cent). Both slurries were allowed to stand for approximately one-half hour to allow any entrapped air to escape and to make vacuuming unnecessary. The two slurries were then added together, and stirred until the resulting slurry was uniform. As the slurry was stirred it became thicker, approaching a gel. This resemblance to a gel was probably due to the polyvinyl alcohol going out of solution in very fine, almost colloidal size, particles, and these par-

ticles being interlaced with each other in an arrangement resembling a veil or spider web. This reaction described is ideal for precoat preparation in that the particles will be properly suspended so that continuous mixing is not necessary to prevent settling, or separation due to varying weights, and densities of the particles.

In those compositions where polyvinyl alcohol could be used together with sodium silicate as the







Fig. 1—Cast specimens of the 10, 20, 30 and 40 series of precoat compositions. The casting number is the and 41, A=26315-P:

20, 30 and 40 series second digit of the precoat mix number. Example: casting number is the casting No. 1 contains specimens numbered 1¹, 2¹, 3¹, and 4¹. A-26315-P; B-26314-P; C-26316-P.

binders, it was believed advisable to do this to gain the maximum in particle suspension. As an example zirconium silicate which could be prepared with either polyvinyl alcohol or sodium silicate was prepared with a 20-deg Be sodium silicate solution, while

the alumina had to be prepared with a 2 per cent polyvinyl alcohol solution. In the silica-zirconium silicate series of compositions, zirconium silicate was prepared with a 2 per cent solution of polyvinyl alcohol, and the silicate had to be prepared with sodi-







Fig. 2—Cast specimens of the 50 and 60 series of precoat compositions. A-26559-P; B-26458-P; C-26460-P.

um silicate. Although the combination of two types of binders in one composition complicated the mixing procedures, it was worthwhile, because even though the density of zirconium silicate is approximately two times that of silica, the gel-like effect obtained when the two slurries with different binders were mixed together prevented any separation or settling of the particles.

The silica-zirconium silicate compositions were mixed in the same manner as the previously discussed alumina-silica group of compositions except zirconium silicate was prepared with polyvinyl alcohol and the silica with sodium silicate.

The silica-magnesia series of compositions was prepared as follows: The magnesia and silica after the sand addition was made were dry mixed and added to the required volume of sodium silicate solution. The mix was then stirred to produce a uniform slurry. This group could not be prepared with both binders, polyvinyl alcohol, and sodium silicate, because of coagulation, but the slurry was satisfactory for dip coating.

The alumina-magnesia compositions were prepared in the same manner as described for the alumina-silica series. The magnesia being prepared with sodium silicate, while alumina had to be prepared with polyvinyl alcohol.

The alumina-zirconium silicate compositions were prepared as described for the alumina-silica series. The zirconium silicate was prepared with sodium silicate, while the alumina had to be prepared with polyvinyl alcohol.

An additional series of compositions, silica-aluminazirconium silicate was made because the results on the previously described compositions showed evidence of a volumetric relationship. This relationship indicated that the compositions where the ceramics were in the ratio of one to one to each other were the best precoats. Therefore in this series the compositions were prepared so that the constituents would be in the approximate ratio of 1:1:1, or 1:1 of the sum of the other two, or 1:1 with a small addition of the third constituent. Table 3 shows the compositions, per cent by volume, and weight, also the amounts of ceramics, filler sand and binder used for each sample precoat preparation. In this group, zirconium silicate and silica were dry mixed with their sand additions and prepared together with sodium silicate. The alumina was mixed with its sand addition and prepared with polyvinyl alcohol. The slurries were then mixed together as described in the preparation of the alumina-silica compositions.

The wax patterns, consisting of specimens 1 in. by $2\frac{1}{2}$ in. by $\frac{3}{16}$ in. attached to sprues, were cleaned by dipping in acetone. When dry, three patterns were

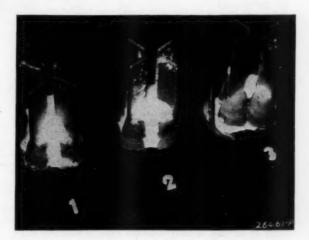


Fig. 3-Alumina-silica No. 5 precoat prepared with polyvinyl alcohol and sodium silicate solutions in the ratio of 10 to 2.5. 1. Heat resistant alloy casting; 2. Stainless steel casting; 3. Alloy steel casting.

TABLE 4-RESULTS OF PRECOAT COMPOSITIONS INVESTIGATED

No.		Con	posi	tion,	%		Resistant T			Stainless Ste			Alloy Stee	
						Cerami	c Surface	Surface* Smoothness	Ceramic Adhered		Surface* Smoothness	Ceramic Adhered?	Surface Oxidation?	Surface Smoothn
								ilica-Alumina	Compos	sitions				
0		0 sili				Some	Yes	Good	Some	Yes	Good	No	Yes	Good
10	100	0 aku	ımin	a		Yes	Yes	Poor	Yes	Yes	Poor	No	Yes	Fair
11		silic		alum	ina	Yes	Yes	Poor	Yes	Yes	Poor	Some	Yes	Fair
12	20	29	80	9.9		Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Fair
13	30	22	70	29		Yes	Yes	Poor	Yes	Yes	Poor	Some	Yes	Poor
14	40	22	60	22		Yes	Yes	Poor	Some	Some	Poor	Some	Yes	Poor
15	50	22	50	29		Yes	Yes	Fair	Some	Some	Poor	No	Yes	Fair
16	60	**	40	99		No	Edge Re- oxidation	Good	No	Some re- oxidation	Fair	No	Yes	Fair
17	70	22	30	22		No	Reoxidation	Good	No	19	Good	No	Yes	Good
18	80	93	20	22		Some	Some	Fair	No	19	Good	No	Yes	Good
19	90		10	27		Yes	Yes	Fair	Some	Yes	Fair	No	Yes	Good
							Silic	a-Zirconium	Silicate	Composition	ıs			
20	100	zire	oniu	m sili	icate	No	Yes	Good	No	Yes	Fair	Yes	Yes	Poor
21					ium silicate		Yes	Good	No	Yes	Fair	Some	Yes	Fair
22	20	SHILL	80	ZIICOII	"	No	Yes	Good	No	Yes	Fair	Some	Yes	Fair
23	30	22	70		21	No	Yes	Fair	Yes	Yes	Poor	No	Yes	Fair
	40	22	60		grs	No	Yes	Fair	No	Yes	Poor	No	Yes	Good
		**			**									Good
25	50	22	50		10	No	Yes	Good	No	Yes	Fair	No	Yes	
	60	**	40		**	Some	Yes	Good	No	Yes	Poor	No	Yes	Good
	70	**	30		**	Yes	Yes	Poor	No	Yes	Fair	No	Yes	Good
	80	**	20		**	Yes	Yes	Poor	Some	Yes	Good	No	Yes	Good
29	90	**	10	*		Yes	Yes	Poor	Some	Yes	Good	No	Yes	Good
								ica-Magnesia	Composi					
30	100	mag	gnesi	a		Yes	Yes	Poor	Yes	Yes	Poor	Some	Yes	Fair
31	10	silica	-90	magne	esia	Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
32	20	22	80	20		Yes	Y _t es	Poor	Yes	Yes	Poor	Yes	Yes	Poor
33	30	**	70	22		Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
34	40	89	60			Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
35	50	27	50	22		Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
	60	9.9	40	22		Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
	70	9.9	30	9.2		Yes	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
	80	11	20	2.9		Yes	Yes	Good	Some	Yes	Good	No	Yes	Fair
	90	22	10			Yes	Yes	Good	Some	Yes	Fair	No	Yes	Fair
							Alumi	na-Magnesia	Composi	tions				
10	100	mag	mesi	a		Yes	Yes	Poor	Yes	Yes	Poor	Some	Yes	Fair
11	10 a	alum	ina-	90 ma	gnesia	Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
	20	9.9		80	**	Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
	30	22		70	**	Yes	Yes	Poor	Yes	Yes	Poor	Yes	Yes	Poor
	40			60	**	Yes	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
	50	25		0	0.0	Yes	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
	60	22		10	99	Some	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
	70	200		10	11	Some	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
	80	213		20	99	Some	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
	90	15		0		Some	Yes	Poor	Some	Yes	Poor	Yes	Yes	Poor
								a-Zirconium		Compositions				
0	100	alun	nina			Yes	Yes	Poor	Yes	Yes	Poor	No	Yes	Fair
				0 zire	conium silicate	Yes	Yes	Poor	Yes	Yes	Poor	No	Yes	Good
2 8	80	2.0		20	**	Yes	Yes	Poor	Yes	Yes	Good	No	Yes	Good
	70	20		30	99	Yes	Yes	Poor	Yes	Yes	Good	No	Yes	Good
	50	99		10	**	Yes	Yes	Poor	Yes	Yes	Good	No	Yes	Good
	50	22		50	**	Yes	Yes	Poor	Yes	Yes	Good	Yes	Yes	Good
	10	90		50	**	Yes	Yes	Poor	Yes	Yes	Good	Yes	Yes	Good
		**			99				Some	Yes	Good		Yes	Good
	30	21		70		Yes	Yes	Good				No		
	20	22		30	**	Some	Yes	Good	Some	Yes	Fair	Yes No	Yes	Good Good
9 1 Syst	l0 em	11800		90 grad		No smoothr	Yes ness was as f	Good follows:	Some	Yes	Good	No	Yes	Good
Jyat	CILL		ade	State	Profilon	neter Re	ading (RMS	Units)						
		Go				Less that								
		Fa	ir			100 to 2								
		Po				Creater	than 200							

precoated, by dipping, per composition. It was necessary in order to fully evaluate each precoat composition to prepare three specimens of each, one for each alloy to be cast, a heat-resistant type, a stainless steel, and an alloy steel. When the precoat had set, it was waterproofed by dipping in Tygon paint. On drying the patterns were assembled in groups of three or four different compositions to a wax-coated plate by welding. The reason for assembling in groups was

economy in use of investment. The investment material (Ransom and Randolph's IC-711-G1) was prepared with water according to the maker's directions and slip cast into flasks surrounding the assembled patterns.

th

70 71 72 73 * Sa Sa

When the investment had set, the wax-coated plates were removed from the molds, and the molds placed in the wax melt-out oven overnight at a temperature of 180 F for wax removal. The following morning

the molds were transferred to the firing furnace where the temperature was slowly increased to 1600 F. At this temperature they were allowed to soak for several hours to insure that all organic material had been oxidized.

A stainless steel (Type 302) and alloy steel (4340) and a heat resistant-type alloy (this alloy was composed of scrap from the alloy development program, and had an approximately analysis of 20 per cent Cr, 40 per cent Ni, 0.5 per cent C, 0.5 per cent Si, 1.0 per cent Mn and balance Co), were melted by induction heating and poured by gravity into the molds. Several 5-lb heats were made of each alloy. Chemical analysis was not taken on the various heats because the stock was the same and melting conditions were constant for all melts. Pouring rates were also controlled, by having one melter make all castings.

The mold temperature used for all casting was 1600 F, and the pouring temperature was approximately 2700 F plus or minus 50 F as measured by optical pyrometer, for all alloys except the alloy steel (4340) which was poured at approximately 2850 F plus or minus 50 F. The alloy steel castings poured in the first group (Fig. 1C) exhibited gas holes, probably due to insufficient deoxidation. However, the surface condition was such that it could be used for precoat evaluation. A silicon addition of 0.5 per cent was made to the second group of alloy steel castings (Fig. 2C). This melt produced sound castings. Therefore the assumption that the large holes in the first group were due to insufficient deoxidation was believed correct.

Each precoat was evaluated for its reaction with the aforementioned alloys on the basis of precoat adherence, surface smoothness, and degree of oxidation. The results were obtained by visual examination of each casting in the as-cast condition. The castings were examined as they came from the investment mold for precoat adherence, surface oxidation, and surface smoothness prior to any type of finishing operation. Questions regarding ceramic adherence and surface oxidation were answered, and are included in Table 4. Each casting was examined for surface smoothness, and given a grade of good, fair, or poor. Profilometer readings of representative castings of each grade were taken to give the reader a value to compare with the grades. Figures 1 and 2 are photographs of the castings on which the aforementioned macroscopic examinations were made.

An analysis of the results, as tabulated in Table 4, indicated that the alumina-silica group of compositions in the range from 60 per cent silica—40 per cent alumina, to 90 per cent silica—10 per cent alumina offered the most promise of fulfilling the requirements for the ideal precoat as set forth in the introduction of this report. However the mixing procedure for these compositions is involved. Therefore it was thought advisable to conduct some additional experiments to attempt to develop a one-solution binder, which would give the same suspension and binding characteristics as the separate solutions of polyvinyl alcohol, and sodium silicate when used as previously described for alumina-silica precoat preparation.

The precipitation of polyvinyl alcohol in the presence of the silicate ion prevented the mixing of the two solutions in all proportions. However, it was found that a stable solution of the two could be made in the ratio of 2.5 parts sodium silicate solution (20 deg Be) to 10 parts polyvinyl alcohol solution 2 per cent. This binder solution was evaluated by preparing precoats and making castings. The composition selected for binder evaluation was 62.5 per cent

TABLE 5-RESULTS OF SILICA-ZIRCONIUM SILICATE-ALUMINA PRECOAT COMPOSITION

No.	Comp	osition (%	vol.)	Heat 1	Resistant Ty	pe Alloy		Stainless Ste	el			
	Silica	Zirconium Silicate	Alumina		Surface Oxidation?	Surface* Smoothness	Ceramic Adhered?	Surface Oxidation?	Surface* Smoothness	Ceramic Adhered?	Surface Oxidation?	Surface* Smoothnes
61	33.3	33.3	33.3	No	Some re- oxidation	Good	Yes	Yes	Poor	No	Yes	Good
62	25	50	25	Some	Little	Fair	Yes	Yes	Poor	Some	Yes	Fair
63	17.7	41.2	41.2	Some	Some re- oxidation	Fair	Yes	Yes	Poor	No	Yes	Fair
64	15	50	35	Some	23	Fair	Yes	Yes	Poor	Yes	Yes	Fair
65	23	23	54	No	Edge oxi- dation	Good	Yes	Yes	Poor	No	Yes	Fair
The		n used for rade				RMS Units)						
	Ge	ood	_		than 100							
	-	nir oor			to 200 iter than 20	00						

TABLE 6-ALUMINA-SILICA PRECOAT PREPARED WITH DIFFERENT BINDER SOLUTIONS

	Comp	osition, %	Amou	int of C	eramics (g	rams)			4	
No.		Alumina	Silica	Sand*	Alumina	Sand*	Sodium 20°		Polyvinyl Alcohol (2%)	Sodium Silicate/Polyvinyl Alcohol 20° Be / (2%) = 25/10
70	62.5	37.5	93.75	20.4	56.25	17.1		41	58.5	
71	62.5	37.5	93.75	20.4	56.25	17.1			99	**
72	62.5	37.5	93.75	20.4	56.25	17.1		88		
73	62.5	37.5	93.75	20.4	56.25	17.1				80.5
San	d Addi	tion—The sa tion to Silic tion to Alu	a _2	1.7% by	weight	on to al	ll materia	il was 139	by volume.	

silica—37.5 per cent alumina. This composition was prepared with various binders as shown in Table 6. First with separate solutions of sodium silicate and polyvinyl alcohol as described for the preparation of the alumina-silica series of compositions, secondly with polyvinyl alcohol alone, thirdly with sodium silicate, and fourth with the stable solution of sodium silicate and polyvinyl alcohol in the ratio of 10 to 25 by volume.

The results of this experiment show that a combination of polyvinyl alcohol and sodium silicate solutions in the ratio of 10 to 2.5 can be used as a binder. A photograph of the castings is shown in Fig. 3. The use of this new binder solution will elminate one mixing step in the preparation of aluminasilica precoats. It is now possible to dry mix the ceramics, and add them directly to the binder, and stir to produce the dip-coat slurry, which will produce results comparable to those obtained when the materials were mixed separately.

Discussion of Results

Preliminary observation of the castings, produced to study the effect of different precoat compositions, showed that no one precoat composition of those investigated filled the role of the ideal precoat for all the alloys cast. Therefore the castings were compared in alloy groups so that the most satisfactory precoat compositions could be selected for each alloy cast. The criteria for examination were precoat adherence, surface smoothness, and degree of surface oxidation.

Examination of the heat-resistant alloy castings indicate that compositions, No. 16 (60 silica-40 alumina) and 17 (70 silica-30 alumina) yielded the best results for this alloy. These compositions produced a smooth, shiny, oxide-free surface on the casting. The compositions of No. 16 and 17 are similar to the precoat alumina-silica No. 5 (62.5 silica-37.5 alumina previously reported by the authors in USAF Technical Report No. 5925.

In the case of the stainless steel castings it was found that composition No. 17 (70 Silica-30 alumina) was the most satisfactory, producing a casting with a smooth surface, and little surface oxide present. However, compositions No. 16 (60 silica-40 alumina) and 18 (80 silica-20 alumina) should not be overlooked, because they also produced castings with very good surfaces.

Precoat compositions No. 18 (80 silica-20 alumina) and 19 (90 silica-10 alumina) were found to be superior to the other compositions for use with alloy steel casting. These castings were very smooth, but the oxide layer was of the tightly adhering type rather than the scale-like layer obtained with the standard silica precoating. However, this oxide can be removed in the usual manner by a blasting operation, or chemical treatment.

The probable reason for the beneficial effect of alumina-silica precoat in minimizing surface oxida-

tion, and adherence of coating to castings of stainless steel, and heat treatment alloys, may be due to the following mechanism: When the molten metal (approximately 2700 F) is cast against the precoat a reaction takes place to produce a glass. This glass absorbs by diffusion the surface oxide, which has been formed by reaction of the metal with the oxygen contained in the mold cavity, forming a slag. The differential contraction rates of the metal and slag would then account for the lack of adherence, because at the mold cools to room temperature the casting contracts at a much faster rate than the coating, and therefore the casting and the precoat part. The casting consequently can be taken from the mold without precoat adherence, and with only the oxidation which may have occurred after the separation of the casting and pre-

The other materials investigated, zirconium silicate, and magnesia, did not show as promising results as the combination of alumina and silica. However, the zirconium silicate coatings produced results comparable to the standard silica precoat. Magnesia coatings were all very poor. This was probably due to the high shrinkage of the magnesia. The use of fused magnesia which has a higher apparent density might alleviate the aforementioned condition.

Another observation of this work was that no precoats cracked during setting or firing. The explanation for this is believed to be related to the addition of 13 per cent by volume of 40-mesh silica.

Conclusions

Of the various ceramics, and preparation procedures investigated, alumina-silica precoats offer the most promise of fulfilling the requirements of the ideal precoat. The surface oxide on heat resistant alloy and stainless steel castings is almost eliminated by the use of this precoat composition. In the case of alloy steels, the only improvement noted was that the surface oxide was of the tightly adherent type rather than the scale-like layer found on castings when the standard silica precoat was used.

Precoats of zirconium silicate were found to be comparable to the standard precoat, while the magnesia compositions all produced poor results.

DISCUSSION

h

d

p

cı

Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

Co-Chairman: H. W. DIETERT, Harry W. Dietert Co., Detroit. G. N. ZIEGLER: 1 What back-up investment material did the authors use against these precoats?

Mr. Strott: We used silica sand back-up material.

CHAIRMAN RASSENFOSS: Steel foundrymen strive to attain better surface finish and decrease metal penetration. What the authors used for these investment castings might have a possible use in that field. Do you think the results would still hold if the weight of the casting increased?

Mr. Strott: We believe the results would be similar on castings weighing up to 50 lb.

Engineer, The International Nickel Co., Inc., Bayonne, N. J.

IMPORTANCE OF SLAG CONTROL IN BASIC CUPOLA OPERATION

By

R. A. Flinn and R. W. Kraft*

ABSTRACT

The basic cupola is becoming of great importance because of the scarcity of pig iron for normal gray iron production and the growing need for low sulphur, high carbon base metal for ductile iron. The success of basic melting depends largely upon the control of the analysis and properties of the slag. The object of this paper is to analyze previous equilibrium data for desulphurization, dephosphorization and carburization as it affects these reactions in basic cupola operation.

Sulphur removal depends upon the attraction of the slag for sulphur compared to the attraction of the metal expressed as: %8 slag/%8 metal or more accurately, activity S slag/activity S metal. The ratio, % S slag/% S metal for 3.6 per cent C, 1.5 per cent Si iron can be varied from 80:1 to 8:1 in basic cupola slags. The best conditions are obtained with high basicity ratio, CaO+2/3 MgO/SiO₂+Al₂O₂, and low iron oxide content. The quantitative variation of this effect is discussed first from an equilibrium standpoint and then a nomograph for cupola operating conditions is presented.

Phosphorus removal is expressed by the factor: 4CaO P₁O₁ slag/9 Pⁿ metal. Best dephosphorization is obtained at high iron oxide content, a condition opposing optimum desulphurization and carburization.

Carburization is normally greater with basic than with acid practice, probably due to the lesser adherence of slag to the coke and the use of more reducing slags.

Operating experiments indicate that a 60 per cent steel, 40 per cent shop scrap charge may be melted to provide 3.5 per cent carbon, 0.02 sulphur hot metal at 2750 F.

IN ACID CUPOLA PRACTICE slag has generally been treated merely as a necessary evil. As long as the slag has been sufficiently fluid and of the lowest possible volume, foundrymen have shown little interest except in its disposal.

In sharp contrast, slag control is the heart of basic cupola practice. This is because the basic cupola slag can refine while the acid cupola slag is usually only an incidental in melting metal. In the basic method desulphurization or dephosphorization can be accomplished while sulphur and phosphorus usually increase in the acid method. In addition, greater carburization can be accomplished with basic practice. These advantages are summarized in Table 1.

TABLE 1—COMPARISON OF OPERATIONS IN 31/2 CUPOLA LINED SUCCESSIVELY WITH ACID AND BASIC MATE-BIALS FOR DUCTILE IRON PRODUCTION

	Acid	Basic
Charge needed to pro-		
duce 3.5% C iron:	40% Low P & S pig iron	60% steel
	40% shop scrap	40% shop scrap
	20% steel	
Sulphur Content	0.07/0.09	0.02
Phosphorus	0.07/0.09	0.05/0.06

It should be evident from Table 1 that basic operation not only produces ideal hot metal for ductile iron, but also, in the case of standard grades of cast iron, provides an answer to the shortage of highcarbon melting stock such as pig iron and cast scrap in high production periods.

Under these strong incentives many investigations of improved basic practice are under way. In general, these are in two directions: (1) Improvement of patching and lining materials such as the monolithic lining installed with an air gun, (2) Water cooling of the melting zone using a thin lining or no lining. While these studies will lead to cost reduction and greater facility of operation, the action of the slag will remain the key to analysis control and will be governed by the same fundamentals.

It seems worthwhile then to summarize now our present knowledge of cupola slag and to show quantitatively how the operation of a production cupola is affected by these principles. The discussion is divided into two sections:

I-Reactions between slag and metal governing desulphurization, dephosphorization and carburization.

II-Slag control under operating conditions.

In other words, the maximum refining power of a given slag and the mechanism of refining will be considered first, along with a number of actual cupola slag-metal combinations. Secondly, the influence of the slag in production will be discussed.

I-Reactions Between Slag and Metal

Investigation under equilibrium conditions is the simplest method to study slag-metal relationships. If

^{*} Metallurgical Department, American Brake Shoe Co., Mahwah, N. J.

certain amounts of slag and metal are agitated under controlled conditions in a closed crucible, a point is reached where no further change in composition of either slag or metal occurs. The experiment may be repeated and duplicate results obtained. In contrast, the relationship of slag to metal analysis in a cupola may change simply if the number of shutdowns is changed. Equilibrium conditions permit, therefore, the pure study of the ultimate refining capacity of a given slag, unconfused by operating variables. This technique has been highly developed by Chipman and others over the past 20 years.^{1, 2, 3}

Three reactions are of principal interest: desulphurization, dephosphorization and carburization. In each case the equilibrium data will be discussed first and then the cupola slag data.

A-Desulphurization

1. Equilibrium Conditions — Rocca, Grant and Chipman have demonstrated recently² that sulphur is removed from liquid metal by two mechanisms:

1. The slag dissolves iron sulphide from the metal until equilibrium is reached, that is, the tendency of FeS to escape from the metal equals the tendency of the FeS in the slag to escape to the metal:

(1) FeS metal ≒ FeS slag

Iron sulphide reacts with lime (CaO) at the slag metal interface producing calcium sulphide and iron oxide:

(2)
$$FeS + CaO \Rightarrow CaS + FeO$$

If the iron oxide of the slag is already high, very little CaS can be found in the slag and reaction (1) predominates.

Since the desired end result is sulphur in the slag, whether as FeS or CaS, the desulphurizing power may be expressed, after analyzing a given metal-slag combination for sulphur, as %S slag/%S metal. This ratio may vary from 400:1 to as low as 1:1.*

What are the conditions affecting the desulphurization ratio? First of all, the ratio is not constant for a given slag if the analysis of the metal is changed. In other words, the activity of the FeS, its tendency to leave the metal, varies with metal analysis. For example, when carbon and silicon are high, as in cast iron, FeS fortunately has a greater tendency to leave. These effects are described by Hatch and Chipman¹ and Morris and Buehl.⁴ In a 3.6 per cent C, 1.5 per cent Si cast iron, the ratio %S slag/%S metal with a given slag will be four times higher than for pure iron with the same slag. To compare data obtained with different metal analyses "%S metal" is multiplied therefore by the proper factor to yield the activity

" a_8 metal." The ratio ${}^{\circ}_{0}S$ slag/ a_8 metal obtained for different metal analysis with the same slag will then be constant and the action of the slag alone may be studied.

The two principal factors affecting slag desulphurization are iron exide content* and basicity,* as shown in Fig. 1 by the data of Rocca, Grant and Chipman.

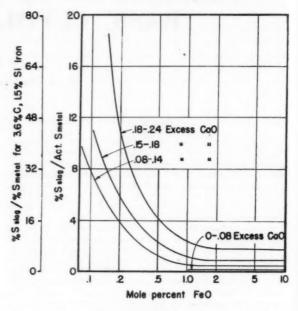


Fig. 1-Effect of Slag Analysis on Desulphurization. (After Rocca, Grant and Chipman²).

In the basic cupola both low iron oxide (below 2%) and high basicity may be obtained while in the acid cupola less basic conditions and sometimes more oxidizing conditions are encountered. The effect of iron oxide is shown clearly by Fig. 1 and the effect of varying %CaO vs %SiO₂ in the slag is indicated by the excess base calculation.

2. Desulphurization Under Operating Conditions—Slag and metal samples were taken simultaneously near the tap hole of a No. 3½ Whiting cupola with a magnesite lining during a number of heats. Analyses

$$b = \frac{CaO + MgO}{SiO_2 + Al_2O_3}$$
 (molar per cents)

^{*} The use of a ratio of this sort is simpler in evaluating slags since the relative weights of slag and metal are eliminated from consideration and may be substituted in later calculations. For example, assume that the average sulphur content of a scrap charge is 0.10 per cent and a final content of 0.02 per cent is desired in the metal. If equal weights of slag and metal were employed, a desulphurization ratio of 0.08/0.02 = 4 would be satisfactory if equilibrium were attained. If, for practical operating conditions, only 5 lb of slag to 100 lb of metal were to be used, a desulphurization ratio of 80:1 would be required under equilibrium conditions.

^{*} Iron oxide is computed from the chemical analysis as %FeO + 0.9% Fe₂O₃. Per cent Fe₃O₃ is adjusted by 0.9 to equalize the molar %O in Fe₂O₃ with that in FeO. Free base is computed by the following steps:

^{1.} Determine basicity ranking:

^{2.} When b less than 2.3,

Bases (CaO, MgO) are assumed to combine with acids (SiO₂ and Al₂O₃) on a 1:1 molar basis. After CaO and MgO are used in proportion to their percentage to neutralize acids, the remaining CaO is considered free lime.

^{3.} When b is greater than 2.3,

Bases (CaO, MgO) are assumed to combine with acids (SiO, and Al₂O₃) on a 2:1 molar basis. Free lime is then residual CaO as in (2).

TABLE 2-SLAG-METAL COUPLES FROM BASIC CUPOLA

Identi-				ag Analyses_					Ietal Analys	ies
fication	%CaO	%MgO	%Al ₂ O ₃	%SiO ₂	%FeO	% Fe ₂ O ₃	%S	S%	C%	%
5005-1	39.50	16.30	8.15	32.35	1.61	0.33	0.33	0.012	4.12	0.7
5007-1	50.60	9.15 18.60	8.75 5.15	28.30 22.60	0.60 1.10	0.16 0.16	0.69	0.014	3.68 3.87	1.0
-2	48.80						0.54		3.87	
5008-1	28.80 28.70	17.70 29.30	12.70 7.70	37.00 25.30	0.81 2.95	0.33 0.16	0.31 0.49	0.024 0.026	3.85	1.3 1.1 1.3
-2 -3	25.40	31.90	7.35	27.70	3.60	0.41	0.30	0.031	3.77	1.9
5009-1	38.90	17.00	9.40	31.60	0.80	0.08	0.49	0.018	3.64	1.1
-2 -3	41.10	19.45	8.50	28.60	0.88	0.16 0.08	0.64	0.017 0.018	3.71	1.1 1.0 1.1
	45.10	16.00	7.15	28.10	0.73		0.81		3.85	
5010-1	46.20 42.30	8.80 13.15	12.80 13.10	29.00 29.60	1.10 0.73	0.16 0.16	0.71 0.48	0.025 0.015	3.12 3.71	1.4 1.3 1.0
-2 -3	44.80	21.20	7.70	23.80	0.88	0.16	0.66	0.014	3.56	1.0
5011-2	32.50	12.05	22.70	29.70	1.38	0.48	0.27	0.023	3.47	1.3
-3	44.30 42.30	19.00 22.50	8.90 6.20	24.94 27.40	1.10 1.97	0.16 0.24	0.55	0.020 0.020	3.32 3.30	1.4
-4 -5	42.50	22.80	6.20	23.56	1.25	0.24	0.63 0.81	0.028	3.21	1.2 0.8
6012-1	37.50	25.65	6.20	27.20	1.45	0.32	0.72	0.024	3.68	1.4
-2	38.75	23.50	8.50	26.90	1.45	0.16	0.60	0.024	3.85	1.2
013-1	46.00	12.80	12.80	27.10	0.58	0.08	0.55	0.012	3.42	1.4
-2 -3	51.60	14.22	7.80	23.34	0.88	0.16	0.60	0.011	3.41	1.0
-3 -4	52.40 52.30	14.28 13.12	7.45 7.10	23.10 24.76	0.88 0.65	0.08	0.67 0.77	0.013 0.017	3.40 3.37	1.1
014-1	45.35	9.66	17.27	23.88	1.32	0.08	0.75	0.014	3.50	1.6
-2	48.82	11.30	14.61	23.22	0.59	0.24	0.90	0.014	3.46	1.6
015-1	48.00	10.32	15.06	23.64	1.03	0.08	0.67	0.019	3.60	1.56
-3	49.50	11.25	13.40	23.15	0.73	0.16	0.78	0.017	3.55	1.5
-4 -5	50.00 43.10	11.56 14.48	12.90 15.26	23.00 23.66	0.66 1.04	0.08	0.82	$0.016 \\ 0.015$	3.55 3.55	1.5
		12.30						0.024	3.56	1.6
016-1 -2	41.40 42.30	13.80	17.62 18.02	26.88 23.08	$0.37 \\ 0.51$	0.16 0.08	$0.67 \\ 0.75$	0.022	3.50	1.50
-2 -3	41.80	12.50	20.02	23.90	0.44	0.08	0.96	0.025	3.48	1.50
017-1	49.60	9.44	11.88	26.08	0.73	0.16	1.06	0.031	3.56	1.6
-2 -3	48.60	11.70 10.36	11.66	25.70	0.44	0.24	0.93	0.020	3.57 3.55	1.60
	46.00		15.42	25.84	0.58	0.49	0.68			
018-1	50.45 48.65	6.92 11.60	12.86 11.30	25.52 25.62	0.51	0.82 0.33	$0.95 \\ 0.85$	0.021 0.020	3.58 3.50	1.30
-3	46.45	8.40	17.00	25.76	0.59	0.16	0.63	0.023	3.40	1.50
019-1	52.10	12.60	8.64	23.94	0.80	0.24	0.64	0.023	3.58	1.42
-2 -3	46.35	17.04	8.34	24.66	0.95	0.65	0.64	0.024	3.47	1.45
	42.95	17.24	8.96	27.48	0.88	1.06	0.76	0.025	3.50	1.43
101-1	42.90 44.35	14.80 8.62	10.54 14.42	29.28 29.28	0.81 0.59	0.24 0.73	0.81	0.044 0.038	3.42 3.45	1.40
-2 -3	43.30	15.00	10.26	28.10	0.74	0.65	0.68	0.030	3.55	1.63
102-1	39.60	11.52	18.72	27.60	1.25	0.08	0.58	0.041	3.50	1.68
-2 -4	39.00	16.88	14.04	26.86	1.10	0.33	0.65	0.041	3.43	1.55
-4	40.00	15.80	12.60	27.08	0.74	0.41	0.71	0.034	3.44	1.55
104-1	41.00	17.60	10.54	28.90	0.66	0.25	0.67	0.019	3.70	1.78
-2 -4	44.15 43.95	11.88	14.66 11.16	27.50 27.18	0.52	0.25	0.76 0.91	0.019 0.025	3.67 3.70	1.90
-5	43.60	15.50 17.61	10.06	27.92	0.66 0.44	0.07 0.25	1.01	0.023	3.49	1.75
-6	43.30	16.90	10.90	27.66	0.59	0.07	0.84	0.022	3.64	1.50
-8	40.85	18.20	9.74	28.00	0.81	0.16	0.90	0.022	3.65	1.70
105-1	34.60	19.96	9.86	33.72	1.25	0.24	0.42	0.052	3.61	1.72
-2	35.60	22.70	8.22	31.80	0.88	0.16	0.71	0.049	3.44	1.56
106-1	38.20	18.80	10.02	30.92	1.39	0.57	0.78	0.037	3.66	1.42
108-1	36.45	19.25	14.08	27.60	1.55	0.08	0.47	0.026	3.67	1.81
-2 -3	32.30	28.30	7.88	27.96	1.62	0.33	0.68	0.034	3.47	1.69
-3	31.75	28.40	8.06	28.06	1.70	0.16	0.59	0.031	3.48	1.69
109-1	40.20	17.95	10.14	31.26	0.52	0.08	0.46	0.019	3.87	2.00
-2 -3	41.05 35.60	21.40 26.15	7.92 8.34	27.94 27.96	0.74 0.74	0.16 0.08	0.75 1.14	0.024	3.73	1.83
								0.028	3.65	1.70
10-1	48.30	13.52	7.90	28.20	0.81	0.16	0.37	0.019	3.90	1.87
-2 -3	51.35 49.90	15.20 16.20	5.36 5.96	24.70 24.76	1.04	0.65 0.24	0.54 0.66	0.022 0.021	3.75 3.67	1.69 1.64
-4	45.85	17.95	7.50	25.90	1.04	0.24	0.76	0.022	3.67	1.38
11-1	51.00	15.10	6.40	24.86	1.18	0.16	0.60	0.017	3.71	1.78
-2	49.55	17.40	5.70	24.30	1.33	0.08	0.64	0.018	3.73	1.65
-3	46.00	19.55	6.36 8.25	24.70 26.00	1.55 1.77	0.08	0.52 0.72	0.015	3.69	1.62

are given in Table 2 and details of the operation are discussed in Part II.

The results were plotted in a number of ways in the attempt to determine the principal factors affecting desulphurization in the cupola. It is recognized that these slag-metal couples are not in equilibrium but, on the other hand, the same forces encountered in the equilibrium studies, such as degree of basicity and oxidation of the slag, should affect the %S slag/as metal ratio.

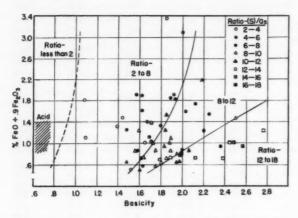
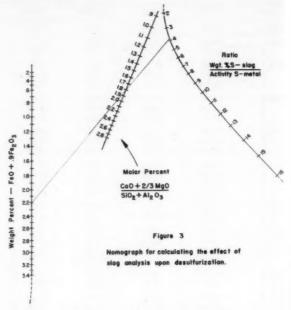


Fig. 2-Effect of Slag Analysis on Desulphurization under Cupola Conditions.

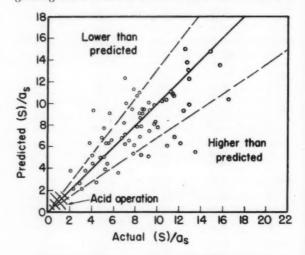
The most satisfactory method found is shown in Fig. 2 where the FeO and basicity are plotted for a given slag-metal couple and the degree of desulphurization is shown by different symbols. It is evident that the best results are obtained at low FeO and high basicity. By contrast, the field of acid slags shows slight desulphurization.

The basicity calculation is taken from Hatch and Chipman¹ dealing with blast furnace slags: Basicity $= \text{CaO} + 2/3 \text{ MgO/SiO}_2 + \text{Al}_2\text{O}_3$ (molar percentages). In the case of cupola slags, the data correlated better with this method than with the free base calculation previously referred to. This effect probably occurs because of the abrupt change in method of free base calculation in the middle of the cupola slag range

From this and similar plots a nomograph (Fig. 3) was developed by trial and error to fit the experimental data. If any two of the following variables are known, the third may be predicted: FeO slag, basicity slag, %S slag/as metal. For example, the nomograph indicates that at low basicity such as encountered in the acid cupola, i.e. below 0.9, the prospects of appreciable desulphurization are poor, even at the lowest FeO content found in the cupola (maximum possible ratio, % S slag/a₈, of 2). In contrast, the sample line on the nomograph indicates that even at a relatively high FeO content (2.2) in a moderately basic slag (1.8) the desulphurization ratio will be 3.9. Also, at FeO above 2.0 per cent, regardless of basicity, the ratio %S slag/a₈ metal will not exceed 8.5. This is borne out by basic open hearth practice where %S slag/S metal does not exceed 8 because of the FeO content of the slag.



The correlation of the nomograph with the 69 samples is considered to be fairly good (Fig. 4) when it is realized that the samples were taken from the beginning and end as well as the middle of the heats and



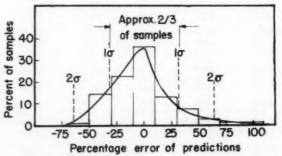


Fig. 4-Accuracy of Prediction of Desulphurization Using Nomograph.

that many experiments in operating conditions were conducted, such as charge make-up, fluxes, blast volume, type and size of coke, number of shutdowns, type of patching material and slag dam height. Temperature variations were also encountered. Although little effect on equilibrium conditions for desulphurization is obtained with temperature, the variation in fluidity of the slag may be important.

The correlation graphs, Fig. 4, indicate that with a given FeO and basicity the desulphurization ratio may be predicted to \pm 32 per cent (standard deviation, i.e. 2/3 of samples). The distribution of variations is

indicated in the lower graph of Fig. 4.

The equilibrium data of Rocca have been analyzed with the nomograph and the astonishing result that desulphurization is poorer than in the cupola is obtained. This does not mean that conditions in the cupola are really better than equilibrium, but merely that the slag-metal couples taken at the spout are not in equilibrium. It is believed that the %S slag is higher than called for by equilibrium data for the following reasons:

One source of sulphur in the slag is coke ash. These particles may fall into the slag and time may not permit an equilibrium between sulphur from this source

and sulphur in the metal.

Secondly, observations through the tuyeres show that pieces of calcium carbide work their way down through the bed. The falling globules of metal roll over this highly basic and reducing medium as well as over the incandescent coke surfaces. In this way the metal is subjected to a highly desulphurizing potential as it passes through the bed. The drops of metal finally pass through the slag layer, but again the path follows the most reduced portions along the coke surface. Throughout its course, therefore, the metal is in contact with better desulphurizing agents than the slag with which it issues from the tap hole. It should be expected then that the resulting sulphur in the metal would be lower than anticipated from the analysis of the slag.

In addition to pointing out that this nomograph does not represent equilibrium conditions, it should be apparent from the preceding discussion that the nomograph is characteristic of a particular basic cupola. For example, the use of a hot blast, a larger diameter cupola or other operating conditions will probably alter the relationship. The fundamental effect of FeO and basicity will be the same so that merely a correlation constant may be necessary to use the nomograph for another set of conditions.

B—Dephosphorization

1. Equilibrium Conditions—Phosphorus probably exists as Fe_3P in the metal and as P_2O_5 in the slag, in combination with basic oxides such as CaO and MgO. The problem then in phosphorus removal is to oxidize the P metal and to form a slag with high attraction for P_2O_5 . The end result of the reaction may be summarized⁵:

² Fe₃P metal + 5 FeO metal ⇒P₂O₅ slag + 11 Fe metal Combined as 4 CaO · P₂O₅ · At equilibrium the balance consists essentially of phosphorus as calcium phosphate tending to be reduced and leave the slag against phosphorus in the metal tending to be oxidized and enter the slag. Winkler and Chipman have derived the following equation to express these forces:

The phosphorus distribution between slag and metal, or in other words the dephosphorizing power of a given slag, is going to depend upon the ratio of:

Rearranging terms:

Dephosphorization is therefore highly dependent on oxidation and basicity. The iron oxide content and basicity of the slag largely determine the oxygen content of the metal⁶ while the basicity of the slag governs the free lime. (A certain amount of CaO, MgO is combined with silica and unavailable for reaction with phosphorus). These conditions are shown simply in Fig. 5 (from Winkler and Chipman).

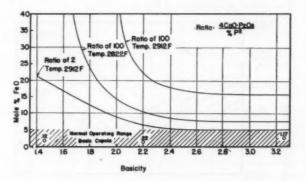


Fig. 5-Effect of Slag Analysis on Dephosphorization (From Winkler and Chipman⁵).

2. Operating Conditions—It is evident that dephosphorization is favored by high FeO in contrast to desulphurization. This condition also prevails under operating conditions as indicated by the slags obtained in the basic cupola under good desulphurizing conditions, Fig. 5. Dephosphorization can, however, be obtained in the cupola by using oxidizing conditions as shown by Carter. Low temperatures also favor this reaction as shown by the equilibrium graphs. In general, this method is not used because of the lower metal temperatures, only slight desulphurization, and poor carburization.

C-Carburization

Carburization is not a refining action in contrast to desulphurization and dephosphorization. It should be considered here, however, since it is related to the slag and is a marked advantage of the basic process. As in the cases previously discussed, equilibrium conditions and then operating data will be considered.

1. Equilibrium Conditions—The carbon content of liquid iron in contact with carbon is determined principally by the temperature and the silicon content, as shown in Fig. 6. High temperature and low silicon develop maximum carbon content. Slightly different solubility lines may be drawn for iron-graphite and iron-iron carbide equilibria, but the differences are too small to consider here.

2. Operating Conditions-If a cupola were operating under equilibrium conditions with metal coming from the well at 2800 F and 1.5 per cent silicon, the carbon content should be over 5 per cent. Furthermore, the carbon content of the charge should have no effect upon this equilibrium between liquid iron and coke at 2800 F. As an example of this tendency, low carbon (0.10 per cent C) steel punchings, can be heated with an excess of coke in an induction furnace to obtain practically any desired carbon content. A further example was encountered in melting 100 per cent steel charges in a 48-in. acid cupola-the carbon content of the first metal from the spout was 3.78 per cent. Furthermore, in blast furnace practice where carbon-free iron ore is charged, the pig iron will analyze 3.5 to 4.5 per cent carbon, depending upon temperature and silicon. The natural tendency of liquid iron in contact with coke is therefore to dissolve carbon to the percentages of Fig. 6.

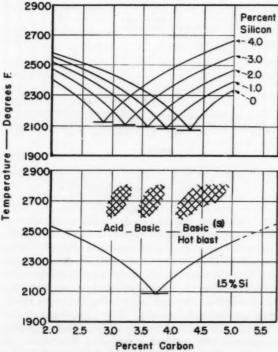


Fig. 6-Effect of Silicon and Operating Conditions on Carbon Solubility.¹⁰

In acid cupola practice the metal is prevented from reaching high carbon levels, principally because of the action of the slag. In the heat with 100 per cent steel just referred to, the carbon content fell to the range 2.50 to 2.75 per cent as soon as appreciable quantities of slag were encountered. It is believed that the action of the acid slag is twofold. First, it produces an adherent protective coating on the top and sides of the coke. Second, the slag contains appreciable amounts of iron oxide which oxidizes the carbon of the metal.

Under these conditions the metal does not reach equilibrium with the carbon of the coke and the carbon content of the ingoing metal becomes a dominating factor. In this case, the formula suggested by Levi⁸ is applicable.

$$(T.C. = 2.4 + \frac{1}{2} T.C. - \frac{1}{4} (Si + P))$$

out in out out

The use of graphite electrodes helps to increase carbon in this case because the graphite is not wet by the slag and a closer approach to equilibrium attained.

In basic practice, the slag does not adhere to the coke as with the acid lining. Furthermore, the iron oxide of the basic slag can be kept low and this results in little carbon loss. An important remaining source of oxidation is the zone near the tuyeres where the incandescent metal falls through an atmosphere containing free oxygen. This zone may be reduced in volume by preheating the blast. Carbon increase of 0.40 to 0.90 per cent is encountered with preheating of the blast to 400 to 900 F.9

II—Basic Cupola Operation

Slag control alone is not sufficient for good basic cupola operation. To obtain satisfactory results more careful supervision is necessary than in acid operation because success depends not only on melting but on refining liquid iron at a fast rate. To indicate some of the operating variables, the log sheet of a typical heat will be discussed.

The cupola used was a No. 3½ Whiting with a shell diameter of 51 in. The stack was lined to 30-in. diameter above the tuyeres. The outer lining was composed of 6-in. fireclay cupola block. The well was lined entirely with magnesite brick, and 4½-in. thick magnesite arch brick were used to provide a basic facing for 72 in. above the tuyeres. Recently a monolithic basic patch, installed with an air gun, has been used to replace burned out zones and for the breast and the spout to the metal notch.

A proprietary graphite clay mixture known as "Hellspot" has also been used for the breast. A front slagging spout with a 4-in. metal head above the top of the tap hole is used.

The cupola melts at about 3.5 tons per hr, and the metal is collected in a 2500-lb stationary forehearth which is tapped into ton ladles every 15 to 20 minutes.

The bed is lighted with wood and burned in with natural draft over a 4-hr period. The charge used during the heat was as follows:*

[•] In more recent practice 60 per cent steel, 40 per cent returns have been used.

Metal	%	Pounds
Steel	37.5	150
Pig	25.0	100
Returns	37.5	150
		400
% of Meta	Charge Weight	
Coke	16.3	65
Fluxes		
Limestone	5	20
Fluorspar	21/2	10
Calcium Carbide	11/2	6

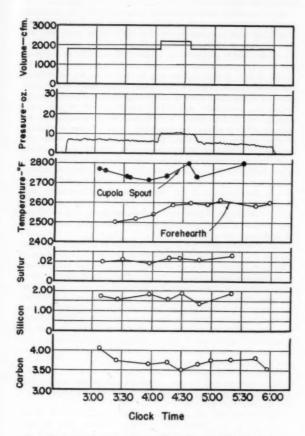


Fig. 7-Log Sheet of Basic Cupola Heat No. 5104.

A typical log sheet is shown in Fig. 7. Metal temperatures ranged between 2720 to 2800 F at the spout and 2500 to 2610 F from the forehearth. Blast volume was increased for a short period to determine the effect of higher temperature and pressure. A slight reduction in carbon was noted.

The principal problems encountered before attaining this practice were:

- 1. Refractory cost and maintenance: Burn out of magnesite brick was severe and cutting of the breast resulted in eventual shutdown. Excessive temperatures (over 2950 F), obtainable with the use of CaC₂ and high blast volume, exaggerated these difficulties.
- 2. Slag fluidity: Improper slag composition, oversize coke led to bridging.

Acknowledgment

The authors greatly appreciate the support and counsel of Mr. R. H. Schaefer, Vice President and Director of Research and Development, and the staff of the Metallurgical Department on these experiments. The assistance of Mr. R. A. Pinto, in recording the production data and helping with the cupola heats, has been especially helpful. The careful analytical work of Mr. B. E. Sockman and his staff, particularly in the determination of slag constituents, was the basis of a large portion of this paper, and therefore was invaluable.

Professor N. J. Grant, of the Massachusetts Institute of Technology, provided helpful data and suggested the importance of FeO content upon desulphurization.

We also wish to acknowledge the cooperation of the staff of the Brake Shoe and Castings Division, especially Mr. T. J. Wood, Chief Metallurgist, Mr. A. J. Derrick, Mahwah Plant Superintendent, Mr. W. T. Bourke, Metallurgist, and Mr. J. Miraglio, Cupola foreman, in this work.

Bibliography

 G. G. Hatch and J. Chipman, "Sulphur Equilibria between Iron Blast Furnace Slags and Metal," Trans. A.I.M.E., vol. 185, pp. 274-284 (1949).

2. R. Rocca, N. J. Grant and J. Chipman, "Distribution of Sulphur between Liquid Iron and Slags of Low Iron Oxide Concentrations," A.I.M.F. Symposium on Desulphurization, February, 1951.

3. Basic Open Hearth Steelmaking, A.I.M.E., pp. 504-549 (1944).

4. J. P. Morris and R. C. Buehl, "The Effect of Carbon Upon the Activity of Sulphur in Liquid Iron," *Trans.* A.I.M.E., vol. 188, February, 1950, p. 317.

5. T. B. Winkler and J. Chipman, "An Equilibrium Study of the Distribution of Phosphorus between Liquid Iron and Basic Slags," *Trans.* A.I.M.E., vol. 167 (1946).

6. K. L. Fetters, and J. Chipman, "Equilibria of Liquid Iron and Slags of the System CaO-MgO-FeO-SiO₂," *Trans. A.I.M.E.*, vol. 145, pp. 95–112 (1941).

 S. F. Carter, "Basic Lined Cupola for Iron Melting," A.F.S. Transactions, vol. 58, pp. 376–392 (1950).

W. W. Levi, "Operation of the Cupola," A.F.S. Transactions, vol. 58, pp. 1-19 (1950).

 D. J. Reese, Private communication based on operation of experimental hot blast cupola, International Nickel Company, Bayonne, New Jersey.

10. Alloys of Iron and Silicon, Greiner, Marsh and Stoughton, 1933, p. 175.

DISCUSSION

Chairman: G. Vennerholm, Ford Motor Co., Dearborn, Mich. Co-Chairman: M. Kuniansky, Lynchburg Foundry Co., Lynchburg, Va.

J. T. Mackenzie: 1 Dr. Flinn did not show the effect of manganese. Shortly after World War I, I got some pig iron containing 0.970 per cent sulphur which by melting with the addition of manganese dropped down to 0.431 on the first melt. By continuing the melting and running the manganese up to 1.17 we got the sulphur down to 0.127 in an acid cupola using a ratio of 8:1 of a 1 per cent sulphur coke. With pitch coke, repeated meltings brought it down to as low as 0.059. So that manganese in the metal is something you have to consider. I do not know whether it will be effective at these very low sulphur levels. But in the acid cupola at normal sulphur levels it is quite pronounced.

Mr. FLINN: Manganese can be quite an important desulphur-

¹ Technical Director, American Cast Iron Pipe Co., Birmingham, Ala.

izer, as Dr. MacKenzie pointed out. A common occurrence, of course, is in dealing with blast furnace metal as it cools down. A constant % Mn \times % S product is obtained depending upon temperature.

E. A. LORIA: ² Dr. Flinn has presented an excellent paper. Slag composition is, in part, a function of the temperature of formation in the cupola and the slag characteristic that probably governs the largest number of reactions in the cupola is the

degree of slag viscosity.

The most important aspect of this subject is the effect of slag composition on the quality of iron produced. Dr. Flinn has shown this relationship nicely for sulphur and phosphorus removal. I would like to consider the matter with respect to gases in cast iron, principally oxygen, which by vacuum fusion analysis varied from 0.001 to 0.025 per cent by weight; differences being attributed to raw materials, operating variables, and even ab-

sorption of considerable amounts after casting.

The information in the literature on the amount of oxygen that is in equilibrium with silicon and manganese in a popular grade of cast iron, say of 2.30 Si and 0.80 Mn, would be expected to contain only about 0.002 per cent at about 2800 F, which we have confirmed several times by vacuum-fusion analysis. Large increases in silicon content in the iron would lower the value only in the fourth place and the limit of accuracy of measurement of oxygen by vacuum-fusion analysis in this range is ± 0.0004 per cent. Carbon is not a factor since the equilibrium with oxygen in iron, as in steel, is strongly influenced by pressure which acts to the detriment of deoxidation with carbon and results in silicon becoming a more potent deoxidizer. Changes in acid slag composition to produce a higher SiO and a lower FeO content would have only little effect on the oxygen content of the iron since the latter is determined more by the silicon content of the metal than the slag in contact with it.

On the other hand, the viscosity of the slag when raised by higher SiO₂ and lower FeO produces another effect in that it provides less absorption of gases by the metal since the more viscous high SiOs slag is a poor transfer medium. This has been shown in the case of hydrogen in steel made by the acid open hearth melting process and the consequent freedom from flake susceptibility. A vigorous boil gets hydrogen to a low level and then the succeeding high SiO2 slag keeps it out. Actually, the same steel compositions made by basic practice with more fluid, basic slags do pick up hydrogen, and flake on processing. Also Fe₂O₂ is difficult to form in an acid slag because of SiO₂. In basic heats, FeO reacts to form Fe₂O₃ at the top of the slag bath while at the bottom of the slag layer the reaction Fe₂O₃ to FeO occurs. Thus, Fe₂O₃ acts as a carrying agent for oxygen in the basic open hearth slag. Of course, in steelmaking the metal composition is such that a greater degree of oxidation can occur therein compared to the composition of cast iron. However, a general relationship between gas absorption and cupola slag basicity is predicted by these observations.

When considering the oxygen content of the slag and its relation to the oxygen content of the iron, there are two systems to be considered. First, there is the partition factor for the oxygen distribution between the slag and the iron. Second, there is the deoxidation constant for the relation between the oxygen and silicon in the iron. When the statement is made that a lower FeO in the slag results in a lower O₂ in the iron, the first factor is being given the controlling position. This would be predominant if the iron contained no C, Si, Mn or other deoxidizers. When the iron does contain large amounts of active, deoxidizing elements, the second factor, the deoxidation constant, has the controlling influence, and the first factor, the partition factor between slag and metal, has minor influence.

When compared with heterogeneous reactions, homogeneous reactions proceed rapidly. For example, the homogeneous reaction between the dissolved oxygen in the iron and the dissolved silicon in the iron is rapid. Therefore, equilibrium is approached rapidly. The speed of the reaction, and the high silicon content of the iron tend to keep the oxygen content of the iron at a low value.

A heterogeneous reaction, such as that between the oxygen in the slag and the silicon in the iron, proceeds slowly. The interface between the slag and the iron forms an effective barrier to the rapid attainment of equilibrium. Some authorities on steelmaking practice feel that even with long times of contact between the slag and the metal, equilibrium is not attained in practice. It has been proven for basic electric reducing slags by the work of Marsh, Urban and Derge. This explained why the high silicon content of the iron effectively holds the oxygen content of the iron to a low value even while the iron is in contact with a slag high in oxygen.

At times the results with respect to V' ratio are not too conclusive. The V' ratio (i.e. CaO/SiO₂) is only an empirical factor which does not tell the complete story of slag behavior. A more definite relationship in special type slags would probably be the calculation of the mol fractions of all the basic oxides and acidic oxides and then setting up a ratio between them. Some assumptions as to the behavior of amphoteric Al₂O₂ in such slags would

have to be made

We have made many acid slag analyses and have remelted a constant amount of each slag on a fusion block in a furnace. The time and temperature were noted when each slag first showed signs of melting, when it began to come over the lip of the fusion block and when it reached the bottom of the block. the fluidity of each slag being indicated to a large extent by the length of time it took for the slag to flow down the block. Generally, the results confirm Dr. Flinn's work and follow the dictates set up by the CaO-SiO2-Al2O2 constitution diagram. In that connection, reference should be made to the region where cupola slags usually occur. This portion of the diagram has two eutectics, one at 1260 C and the other at 1165 C and the contour of the isotherms about these two points should be noted. Acid cupola slags tend to orient themselves around this particular analysis range and it can be seen that changes in basic CaO cut across isotherms sharply whereas changes in the acid constituent SiO₂ produce compositions which run parallel to the isotherms and in the direction of the second eutectic. Thus, some slight increase in SiO2 would still produce fluid, eutectic-type slag. Of course, the degree of superheat and the lowering of the FeO content should be taken into consideration in any comparative study

The clinker cement slags which contain lime, silica and alumina can also be studied by means of the CaO-SiO₂-Al₃O₃ diagram. The diagram indicates that alumina is an excellent flux for limesilica compositions especially in the highly basic region. If the silica content of these slags is sufficiently low, as is many times the case, they do not form sufficient dicalcium silicate to cause disintegration. Many clinker slags are carried slightly oxidizing (1 to 3 per cent CaO); however, this amount of iron oxide is usually sufficient to stabilize any dicalcium silicate formed. Slag composition made up of lime and highly aluminous materials can cover a wide range of melting temperatures. Very fluid slags

can be obtained in this region.

CHAIRMAN VENNERHOLM: The subject of nodular iron is of considerable interest to many foundrymen at the present time. The question may arise as to where the basic cupola fits into

the nodular iron picture.

MR. KRAFT: You do not mean nodular iron as-cast, I assume. We do not see any reason why it cannot be done. You can make an iron suitable for inoculation with magnesium and a good quality of gray iron in the same cupola. You would of course have to change the charges to make two different types, but it can be done.

Mr. Renshaw's experiences and ours are almost identical with respect to the fluxing situation. We have used fluorspar in all the heats to keep the slag fluid. We apparently had the same trouble he did at first. The carbide is used for the same purpose as Mr. Renshaw mentioned, to give low sulphur and increase the carbon. If you are running gray iron you would probably not need the carbide, but I suspect that the use of fluorspar would still be necessary.

CHAIRMAN VENNERHOLM: Another question that has been asked numerous times is, how can you justify the basic operation which involves a refractory which may cost as much as 6 or 7 times more than conventional acid refractory? Will this not offset all the advantages you may otherwise have?

I think it is of considerable interest that work done by both Mr. Renshaw and ourselves has shown that internal water jackets in the cupola may reduce the amount of refractory used to the extent where the refractory cost expressed in dollars will

² The Carborundum Co., Niagara Falls, N. Y.

^{*} See "Basic Cupola Melting and its Possibilities," by E. S. Renshaw, pp. 20-27 in this volume.

not increase, although the material itself may be several times more costly. We have operated a water-cooled cupola for a considerable period of time with an acid lining and found that the refractory erosion in the cupola is only about one-sixth of that of a normal cupola. We have every reason to believe that the condition will be the same with basic refractory. This, therefore, may offer one solution to this particular problem.

E. S. Renshaw: ⁸ In the nature of an experiment we operated a water-cooled cupola under basic slag conditions. Employing water jackets in the melting zone and lining the furnace well with basic refractory, it was found possible to obtain a slag of similar basicity to that produced in the basic-lined furnace. This opens up an interesting field of work and particular reference was made in this paper to the "metallurgical blast cupola" which does appear to function in the same manner. It is logical to assume that the saving in refractory cost should be appreciable.

W. W. Austin: 4 In view of the fact that the basic cupola produces low iron oxide content slags and may be operated under somewhat reducing conditions, I would like to ask if there is any hope for producing nodular iron by melting magnesium in the charge?

MR. FLINN: Dr. Austin asked whether you can somehow get the magnesium in the cupola rather than having the fireworks out in the shop as a result of adding it in the ladle. I do not know. I think we should work in that direction, however. The question of gas is a very important one. It does bear on this magnesium question as well. The solution of the magnesium question may be the control of cupola operation so we can keep the magnesium in solution without having some gas knock it out.

In the future we will have to pay more attention to these minor elements.

MEMBER: Could you use aluminum in place of magnesium? Could you use aluminum-silicon slags in the basic cupola?

Mr. FLINN: That is another possibility, perhaps.

MR. LORIA: It might be possible that aluminum additions could be tolerated in melting iron in the basic cupola in much the same manner as aluminum-ferrosilicon reducing slags are employed in basic electric steelmaking. There, sulphur removal is also achieved under clinker cement and lime-alumina type slags if they are basic, fluid and sufficiently reducing. The formation of alumina would be an excellent flux for lime-silica compositions, especially in the highly basic region. Aluminum is the most powerful of the common deoxidizers and contributes to the essential conditions for desulphurization, namely a strongly

basic slag and one very reducing in character. When added in the acid-lined cupola its tendency to produce pinholes, particularly on the cope side of thin section (plate) castings is well known, but it is not known if the same thing would occur if the addition were made to iron melted in contact with a basic slag.

CHAIRMAN VENNERHOLM: Prof. Piwowarsky, in Germany, was able to produce nodular iron by just melting and superheating

cast iron in a magnesia crucible.

S. F. CARTER: First of all, I would like to thank Dr. Flinn for getting us straight with the physical metallurgists. I was afraid some of the relationships we have been experiencing might be a little illegal. Have you firmly established to your satisfaction that in the denominator of the basicity ratio the Al₂O₈ is equally as effective an acid as SiO₂, and have you had sufficient variation of Al₂O₈ to establish that? Although calcium fluoride is present in minor proportions, did you figure your calcium fluoride as CaO, do you feel that it acts as CaO?

Mr. FLINN: We tried to indicate how much of our work has depended on the work of the physical chemist in steel and we have been guided largely by their practice. We have heats containing as much as 22 per cent Al₂O₃ and as little as 5.3 per cent Al₂O₃. We spent countless hours on this nomograph (Fig. 3) and after we were through with it there was no other way of treating it that would be better. We did take that from the recent work of Rocca, Grant and Chipman² where they treated it similarly.

Now as for calcium fluoride, again we copied them and felt it was sort of an inert thing and we did not pay too much attention to it. We have not been finding too much CaF₂ in our heats. I do not think we have a variation of calcium fluoride to show whether we are treating that right or not.

J. E. REHDER: 4 I would like to compliment Mr. Flinn on not only going seriously into the question of cupola slags, but on doing it in such an enlightening and thorough manner.

Further to Mr. Venerholm's note that nodular iron has been produced in Germany by melting cast iron in a magnesia crucible, it might be mentioned that Morrogh in England has reported the same type of thing. These facts combined with a study of the reactions involved show that magnesium and aluminum can be reduced from alumina- and magnesia-bearing slags over cast iron melts, but that temperatures over 3000 F and considerable time are necessary, so that this method of introducing magnesium into cast iron melts is costly and of little commercial interest at present.

³ Ford Motor Co., Ltd., Dagenham, England.

Senior Metallurgist, Southern Research Institute, Birmingham, Ala

⁵ Asst. Melting Supt., American Cast Iron Pipe Co., Birmingham, Ala.

⁶ Foundry Engineer, Bureau of Mines, Ottawa, Canada.

DEVELOPMENT OF FOUNDRY COURSES IN HIGH AND TRADE SCHOOLS

By

R. W. Schroeder*

What does industry expect from the trade and high schools? What do the schools expect from industry? Answers to these questions will greatly aid in development of foundry courses. The pattern shop should also be considered since the two are inseparable. Not knowing exactly what industry expects, the schools develop a "tide program"—one that rises and falls with the interest of the individual instructor, helped or hindered by the type of support received from industry. The following story will illustrate the point.

A foundryman told the writer that his son was to enter a certain school and that he took time off to visit the foundry teacher, wishing to see the equipment and the course of study. This plant owner was incensed over the fact that the instructor had other duties and did not have time to spend with him in the school foundry. He remarked, "Imagine a foundry instructor having to help coach the team." After questioning he informed the writer that that was his first visit to the school.

Interest in the Schools

What does industry expect of the schools and how can the instructor find out unless there is more of that cooperation which we all desire and still seem unable to attain. Had this man really been interested, he would have looked ahead to see what the school had to offer, and failing to find what he thought desirable, he would have endeavored to do something about it. How many of you have taken time out to visit your schools to see how your future workers are being trained, or to check into the subject material being used? Do you know if the teacher is interested in your problems? Has he visited or worked in one of the shops within the last year? Do you know or don't you care? Is it any wonder that our industry is facing a real man power problem today?

Creating interest in and training new blood for the

foundry or any trade is not as simple as the turning of a spigot. Industry calls for skilled help—the schools hear but cannot produce—young men have graduated and are now lost to other industries, very few drifting into the foundry. The Foundry Industry has done a very poor job of selling in the High Schools and Trade Schools. A salesman in their employ would soon be looking for another job if he used the same brand of salesmanship.

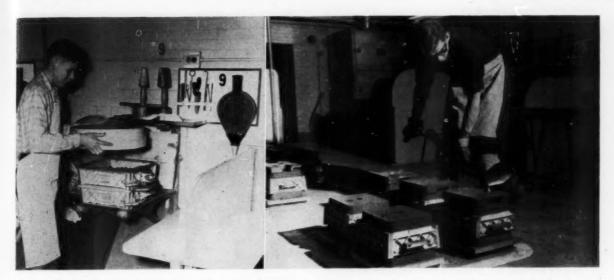
Tide Program

An outstanding example of a "tide program" is that which took place at the Hackley Manual Training School in Muskegon, Michigan. The development of the original foundry course in that school was something to behold. There were no students, a few rusty ladles and shanks, and some rusty flasks.

The foundry course when started years ago fit in with the foundries of the time and no doubt was run very efficiently. Gradually interest lagged and finally the shop was closed and dust was allowed to collect. Was this the fault of the school or industry, or a lack of cooperation?

A few years ago the foundries in the Muskegon area realized that they were not getting their share of the young people and decided to do something about it. Sparked by the leadership of Geo. W. Cannon, then chairman of the board of Campbell, Wyant and Cannon Foundry Co., a corner of the school was cleaned out, all the old "equipment" scrapped, and today they have a model foundry. Active cooperation with industry is as much a part of the program as are the foundry tools and equipment. A thumb nail sketch of that program is as follows:- A definite set of jobs (not exercises) based on progressive learning factors is required, every student must produce a good casting from each job started before he is allowed to start on the next pattern. Shop trips are included and reports are made on each trip by the student. Industry has set up a speaker's bureau and men from various plants are called upon to speak on pattern making, cupola practice, casting cleaning, etc. The foundry instructor at the school places his request for a speaker with the chairman of the Donors

Ass't. Prof., Foundry and Pattern Laboratory, Navy Pier Branch, University of Illinois, Chicago.



(Left)-Student making bench mold at Hackley Manual Training School, Muskegon, Mich. and (Right)-Student pouring off molds.

Committee. The chairman then contacts a man to present a talk on the particular subject. The same speaker would not be called more than once a year. Industry also subsidized the instructors salary, realizing that a man having the qualities they desired would not accept employment for what the school could offer.

The instructor or Foundry Manager as his title reads, is required to work for a short period each year in some one of the contributing plants. This keeps him in touch with industry and modern improvements. This is as far as the writer knows, the outstanding development of a foundry course at the High School level. Where industry is interested and has the real support of the school there can be no limit to what can be accomplished. What might be classed as a super vocational or Trade School program was recently instituted at Tennessee Agricultural and Industrial College. The students taking the two-year program constituting of 2000 clock-hours will have a training comparable to or exceeding any given anywhere in the States. The foundry under the capable guidance of Dr. Frank Crossley and assisted by Prof. Washington will have men who will be sought by foundries in all parts of the country. Dr. Benson L. Dutton, Head of the Engineering Department, is to be congratulated on the fine work he is doing.

I would like to present a brief outline of this course.

Unit I-General Foundry.

This unit covers a historical study of the development of the foundry. Care, use and maintenance of tools, materials and foundry equipment are studied. An elementary study of general foundry operations and practices is encompassed; shop rules, safety regulations and general accepted molding practices are studied. Clock hours, 320.

Unit II-Molds and Molding.

This is a detailed study and practice of the various molding and casting methods such as green sand, permanent molds, slush casting and centrifugal molding. Clock hours, 340.

Unit III-Cupola.

This is a study of the operation and maintenance of the cupola; fundamental methods and principles of cupola operations, charges, combustion and alloying principles. Clock hours, 340.

Unit IV-Metallurgy of Cast Iron.

This covers the manufacture of cast iron with relation to physical properties. It is a practical study of pouring, chipping, grinding, sandblasting and inspecting. Clock hours, 320.

Unit V-Core Making.

This is a practical applied study of core making including core box construction, core sands and binders. Applied practice in core making is also given. Clock hours. 340

Unit VI-Malleable Iron, Steel and Nonferrous Metals.

This is a detailed study of the physical properties of malleable iron, steel and nonferrous metals with relation to foundry procedure. Practical application of molding and casting various nonferrous metal is studied. Clock hours, 340.

This course is for specialized training only. Another course outline covering four years of work at the college level is more extensive. Worcester Polytechnic Institute at Worcester, Mass., in 1949 instituded a long range modernization program by which they expect to keep abreast with the industrial world. I am sure that some of the spirit that motivated the change in Muskegon, Michigan, would be appreciated in that area.

Milwaukee, Wisconsin, always noted for its splendid work in vocational education, has kept pace with this moving world by close cooperation with industry. The foundry course outline requires 2160 clock-hours of school instruction, plus the time spent in the shop as an apprentice. The school program includes related subjects such as: Foundry Arithmetic, Blue print Reading, Foundry Science, Patternmaking, Inspection and Testing, Job Relations and Economic Relations. Foundry Technology also comes in for its share of the time with subjects such as Foundry Fundamentals, Molding Methods, Coremaking Methods, Tools and Equipment, Materials and Supplies and Metallurgy for Molders. Plans for more modern shops are being considered. The Patternmaking Course is more extensive especially in Drawing Mathematics while the related shops include Foundry and Machine Shop.

What does industry expect from the schools?

1. Training for mechanics only.

Training for mechanics who will later become supervisors.

3. Training for supervisory positions.

4. Training to take advantage of the present F.E.F. program.

Development of any course should be influenced by requirements of industry within the area in which the schools are located. Thus the requirements in an area having production foundries working on automotive and plane castings would be somewhat different than in an area supplying repair parts for one of the heavy industries. However, regardless of how far from the great industrial centers, foundries may have been established, we will always find a few production jobs.

These shops, as well as all jobbing shops would be primarily interested in the training of young men to become molders, coremakers and patternmakers. The related subject matter should include elementary metallurgy, foundry mathematics, blueprint reading, sketching, sand control and if possible some basic work in patternmaking. Endeavoring to delve too deeply into this work could easily prove to be detrimental in that workers would all want to become foremen or supervisors. We do need men who are good mechanics and are contented in their work. We



Author explaining some technical aspects of the spiral fluidity test mold to some of his students.



Thomas Ross, Instructor at Rankin Trade School, St. Louis, Mo., about to discuss with students the wood patternmaking project in the 1949 A.F.S. Apprentice Contest.

know that all men do not make good leaders and also that book knowledge may spoil good workers. Mark Twain must have had this in mind when he said, "I never allowed schooling to interfere with my education."

Only by active cooperation in the particular area between school and industry can a curriculum be setup that would reflect the desires of the shops involved. To the fundamentals necessary for trade training in these areas we must add that which will be necessary to give a working knowledge of machine production. This calls for molding machines and necessary auxiliary equipment. Until industry shows real interest in foundry training in the High and Trade Schools, the necessary space, equipment and teachers will not be made available. Industry will create reserve funds, make depreciation allowances, invest in new modern equipment, but never seem to have enough left over for a good training program. It is similar to the old story of the farmer with a leaky roof on the barn-when the sun shines there is no need to fix it-when it rains he can't.

Dr. J. T. MacKenzie, Technical Director, American Cast Iron Pipe, made the statement a number of years ago, that he considered money spent in training as an investment, in that through this training his company had contributed something to the foundry industry and that the company had built up good will in so doing. The development then of any program depends on active support from industry and the desire to serve on the part of the schools.

Where funds have been made available and new schools are erected, they have shown a great deal of interest in purchasing new equipment and making space available for the school foundry. An industry committee formed along the lines as the one at Muskegon, Michigan, will open up wider avenues of man power supply, and do a great deal toward lifting

the foundry to a higher plane in the educational

It would possibly take years to read all the material written on proposed foundry and pattern courses supposed to be used in the High School, Technical School, Trade and Vocational Schools and various coop plans. Apparently what we need is a full time coordinator at this level. So the need is not in curriculum building but in sifting and throwing out the "fines." Both industry and the schools have the ball and like "Wrong-Way Corrigan" cannot or do not want to see the goal. Brought together, a gigantic reservoir of youth eager to enter the foundry field will

In closing I would like to refer again to the Hackley Manual Training Plan. It was the thought of Mr. Cannon at the time, to set up this model foundry, have blue prints made of its layout, list of equipment and duplicate sets of preliminary patterns so that they could be loaned for a reasonable length of time to any school or organization wishing to start a small training shop. Industry has overlooked a potential reserve of man power in not bringing the history and romance of the foundry industry to young people while they are still in the state of flux. Movies, shop trips, lectures to the pupils in the seventh, eighth and ninth grades and to the P.T.A.'s will do a great deal in breaking down the self-imposed barriers raised by our industry due to the lack of good housekeeping and apparent disregard for the health and comfort of their employees. There is an old saying that "foundrymen don't die they just dry up and blow away." Plants that have faced suits on silicosis will dispute that old saw.

We may have an ideal school shop as far as cleanliness and dust removal is concerned and inculcate the habits of safety. To what avail are our efforts when in his first visit to a plant he finds smoke, dust and perchance will see a man through a fog walking along the gangway slowly in hopes that he will reach his floor safely with that ladle of iron-or possibly a lot of core work is being poured and after a few minutes the sharp rancid fumes cause the nose to burn and the eves to tear.

What chance have we of diverting the youngsters into that field? Some foundrymen have stated that they did not care to be put on the visiting list because of the conditions mentioned. Let us have some real development in our industry on personnel work, health programs and improved conditions within the plant. Then we can develop courses that will create the desire to be workers in this basic industry.

n

n

e

g

DISCUSSION

Chairman: W. H. RUTEN, Polytechnic Institute of Brooklyn, Brooklyn, N. Y.

Co-Chairman: G. B. BARKER, University of Wisconsin, Madi-

C. J. FREUND: 1 How much time would a satisfactory high school foundry program require for introducing it into a curriculum which already is rather full?

MR. SCHROEDER: The foundry course would require a 2-hr

period a day. There is a desire on the part of the school officials to cut that down to 1 hr. Even within 11/2 hr it is difficult to conduct class effectively because a mold cannot be put away until the next meeting. It has to be inspected, and if possible, poured so the student will have full benefit of his work. We could accomplish a great deal within one semester, with the possibility of having those students who are really interested, repeating the course as an elective. That would give the student a good idea of whether he would want to stay in the foundry field.

I would like to see a program where we would add an extra year to the high school curriculum. If a student wanted to go to college he would have this practical training and within that year gain enough credits to enter college and carry on at that

CHAIRMAN RUTEN: Did you suggest 2 hr per week for two semesters followed up with 3 hr per week for the advanced course?

MR. SCHROEDER: The Smith-Hughes plan calls for 2 yr, I helieve

Mr. Freund: Is it likely that the authorities at the schools will accept a 2-hr a day proposition? My question does not refer to all communities, but only to those where the foundry industry is important.

MR. SCHROEDER: I think so. In fact the foundry training program at Muskegon, I think, is carried on for the full 4 vr. schools will not do it of their own accord and it is hard for an instructor in a school to get the authorities to do something. But if he has the backing of the industry he may get the time allotted. Lacking that support of the industry and the encouragement from the school itself, the instructor becomes discouraged and does a job of teaching for 6 hr per day and he is through. No instructor can do a job with these students working only 6 hr per day. Working together, you can accomplish a great deal.

MEMBER: We have one trade school in particular in Los Angeles that is training boys for the Foundry Industry. Some of these boys want to obtain work during the summer in the local foundries. We have a State law which prohibits boys under 18 yr of age from working, particularly around moving machinery. Do you have any suggestions whereby we can surmount that particular problem? How do you go about giving them experi-

Mr. Schroeder: That is a real problem.
W. J. Moore: ² There are no Federal laws that I know of that are applicable in the circumstances mentioned. There are State laws and they all vary. The one I am most familiar with is in Pennsylvania. I believe boys under 18 may not work at most jobs

MR. FREUND: I am quite certain that in Wisconsin any properly indentured foundry apprentice is excused from those laws which curtail the kinds of work that he can do under the age

CO-CHAIRMAN BARKER: I do know that if any industry presents a case to the Industrial Commission which has charge of the enforcement of this law, they can get a release from the law. Special permits are given and I assume that is how these boys would be working.

I have no doubt at all, if a boy was 17 or under 18 yr of age, if industry requested the Industrial Commission, they would give him a special permit. But the request would have to come from industry.

G. K. Dreher: 8 The Smith-Hughes Act has a tendency to continue and repeat the errors and practices of the past, among which are the things the author mentioned which keep young men from entering the Foundry Industry. The attitude of the men permitted to train those young men is that of the old school foundryman.

Mr. Schroeder: I wonder if that could not be taken care of by the Board of Education working with the industry. teacher would be required to take a certain amount of college work as he goes along. I am absolutely opposed to the idea that a man has to have a degree, as one educator once stated he would not allow his foundry school to be opened unless the instructor had a Ph. D. degree.

MR. DREHER: We are not talking about getting college men

¹ Dean, College of Engineering, University of Detroit, Detroit.

² Assistant Director, Dept. of Labor, Bureau of Apprenticeship, Harris-

⁸ Executive Director, Foundry Educational Foundation, Cleveland.

to teach a trade, but about getting men with foundry experience to teach and inspire them to enter this industry of ours. The men under the Smith-Hughes Act who are selected and obtained for that purpose are not, generally speaking, the kind of men who inspire a youngster of today to select this industry as a career.

I would rather take the college man and teach him the practical side and get fine character and the ability to act as a leader than take a man who has been a tradesman but who is incapable of further study and knowledge and have him try to lead these young men.

I think, by and large, the high school boys know more about arithmetic and general science than the man who is trying to

teach them foundry practice.

Mr. Schroeder: Some of that will have to be granted. But on the other hand, many men from industry are actually intercsted and I think you can note from Washburne Trade School in Chicago that we have many men who are tradesmen who have gone on and received their college degrees or received enough credits in college work to make them good teachers. The Milwaukee Vocational School also is an outstanding example of that. Practically every instructor there has a college degree. They were taken from industry.

It is a case of industry and the schools working together. Industry has not advised the schools just what it wants. The heavy industries are not represented on many of our boards of education. Until we find industry wanting to go in there and sacrificing some of their time we cannot expect to get those teachers.

F. C. Cech: 4 Is the high school foundry course optional or compulsory for the boys?

MR. SCHROEDER: In some schools it is required, in others it is optional.

MR. CECH: If it is compulsory you will find a few boys who will take the course, but if it is optional the boy will look for the easiest way out, or perhaps because he is inquisitive he takes the course and later drops out. We have started foundry courses with the junior schools in Cleveland. In Cleveland we ship sand in bags to the junior high school. We also ship them metal. We have designed a little flask for the junior high school to use. Occasionally we get some metal through the Board of Education and from industry.

Co-Chairman Barker: I think Dean Freund put his finger on it when he said, "How much work do you want the boys to take in school?" If industry is demanding a trained worker you may have to give him 2 or 3 hr a day every day in the week. If, however, industry is ready to take these men after they get out of high school, give them some in-plant training, then certainly you do not need to teach those boys 2 hr every day of the week.

It was mentioned that Milwaukee has a good training program. It is a 2100-hr program. If you will start adding that up, it amounts to 52 weeks, 8 hr a day, 5 days a week. Are you going

to keep that boy working 8 hr a day, 5 days a week, every day If you do, you are going to drive him out of the Foundry In dustry.

If they give the boy 4 hr a day for a 5-day week, that is 2 yr. I claim that may be altogether more than industry desires But, if industry desires it, let us give it to them. If they do not then why waste the boy's time?

We take boys who are a little over 18 when they come into our universities. We give them very good training. They know Foundry Industry pretty well after we give them one 2-hr session a week for 16 weeks. I do not claim they are finished molders. or patternmakers, but those boys know foundry practice when they leave school. I think it is up to industry to advise the schools how much training they desire in our graduates.

MR. SCHROEDER: I think I made it clear that industry does not make their needs known to the schools. Subsequently you have the "tide program" where the program in the schools depends on the interest of the instructor alone. It gets to the point where the instructor feels he put in a full day's work after 6 hr. He closes the book just as the molder puts his tools away, goes home and forgets about it till the next morning.

You are not going to develop foundrymen or foundry courses with an instructor working on a 6-hr day. Industry must show that instructor that industry is interested in what he is doing so that he will expend himself in developing his courses.

W. J. MacNeill: ⁵ I think industry must plead guilty to this lack of interest to some extent, but I think every foundry's problem is different as to what it wanted the schools to teach. I think the final determination is what your union contracts will let you do about trainces in your own shop.

The set-up at Muskegon has been ideal, but what is expected of industry when these boys have finished the course? Do they just give them a job shoveling in a wheelbarrow or give them

some training in the shop?

MR. SCHROEDER: I did not go back to Muskegon to find out what has been done with that program. I think the boys would become either first class mechanics or foundry technicians, or they would go beyond there and take up additional training. We want to create in the student the desire to go on within this field. So far we in the foundry industry have been satisfied with what has drifted into the foundry. We have to go out after these boys if we expect to get our just share of them, and I do not think we have been doing that.

MR. MACNEILL: I hoped that the industry in Muskegon did take the boys in and trained them further. I do not care what you teach them, if you do not do something about it afterwards the boys are going to become disgusted and discouraged.

MR. SCHROEDER: That brings us back the four points: (1) Training for mechanics only, (2) Training for mechanics who will become supervisors, (3) Training for supervisory positions. (4) Training to take advantage of F.E.F. program.

⁴ Head, Pattern Div., Cleveland Trade School, Cleveland.

⁵ Foundry Manager, Acme Aluminum Alloys, Inc., Dayton, Ohio.

INFLUENCE OF SILICON CONTENT ON MECHANICAL AND HIGH-TEMPERATURE PROPERTIES OF NODULAR CAST IRON

By

W. H. White, L. P. Rice, and A. R. Elsea*

GROWTH AND SCALING CHARACTERISTICS of cast iron have been studied by many investigators. In an earlier paper, White and Elsea1 presented a bibliography of technical articles on growth and scaling of cast iron and reviewed the theories explaining growth in cast iron. The results of these earlier investigations showed that irons containing approximately 6 per cent silicon and less than 2.5 per cent carbon are resistant to growth at temperatures as high as 1700 F. It has been shown that when irons of this type are alloyed with 1.25 per cent copper and 0.5 per cent chromium, these irons are also quite resistant to scaling. In spite of these good high-temperature properties, the useful applications of high-silicon cast iron have been limited because the material is brittle at room temperature.

Recently, methods have been developed² whereby the graphite in cast iron is made to form as a spheroid rather than the usual flake. This change in graphite structure is accompanied by a considerable change in mechanical properties. Not only is the tensile strength increased beyond those limits previously thought possible for cast iron, but the ductility or toughness is also greatly increased. This suggested that it might be possible to improve considerably the ductility of high-silicon cast irons by treating them so as to promote the formation of spheroidal graphite.

Experimental Work

Preparation of Test Materials—Present investigation, conducted under the sponsorship of the Jackson Iron and Steel Co., consisted of determining the effect of variations in silicon content on mechanical properties and growth and scaling characteristics of two series of cast irons of similar compositions, one containing flake and the other containing spheroidal graphite. The silicon content of these irons ranged from 2.6 to 6 per cent, with the carbon content being adjusted in each case to give an iron of approximately eutectic composition (approximately 4.3 per cent carbon equivalent).

The irons were made without alloys since the use of alloys would increase the number of variables that must be controlled. In addition to that, alloys might interfere with the formation of spheroidal graphite, and there was no intention of studying that lactor in the present investigation. Therefore, the proprietary magnesium addition agents could not be used to treat the irons because either they contained large quantities of unwanted alloys or else they contained so much silicon that it would be impractical to use them in making the lower silicon irons. Consequently, it was necessary to devise some other method of adding magnesium. Magnesium metal would be ideal except that its reaction with molten iron is too violent and recoveries are unpredictable. Compacts of magnesium and iron powder for use in deoxidizing cast iron and steel have been described in the literature.3 If such compacts were suitable for deoxidiation, they should afford a satisfactory means of adding magnesium to cast iron to produce spheroidal graphite.

In order to test the feasibility of adding magnesium in this way, several trials were made in which compacts of cast iron chips containing 10 and 20 per cent magnesium as chips were added to ladles of molten cast iron. Cast iron and magnesium in the form of chips were used, since they were readily available. This method of adding magnesium proved to be satisfactory. On the basis of these preliminary tests, pellets containing 20 per cent magnesium were selected for use in this investigation.

The test bars used in this investigation were prepared in the following manner: Five heats of the desired analyses, each weighing approximately 170 lb, were melted in an induction furnace. One-half of each melt was tapped into a ladle, treated with 0.3 per cent silicon added as ferrosilicon, and cast as a double-leg keel block. The remaining half of each melt was treated while still in the furnace with 0.55 per cent magnesium added as the magnesium-iron compacts. The metal was then tapped into the large ladle, treated with 0.3 per cent silicon added as ferrosilicon, and cast as a double-leg keel block. Figure

^{*}W. H. White, Metallurgist, Jackson Iron and Steel Co., Jackson, Ohio, L. P. Rice, Research Engineer and A. R. Elsea, Assistant Supervisor, Battelle Memorial Institute, Columbus, Ohio.

TABLE 1-DATA RELATIVE TO THE MELTING AND ANALYSES OF THE FIVE HEATS USED TO POUR THE KEEL BLOCK CASTINGS

Heat No. Code			Calcula	ated Compos	sition, Per C	Cent	Act	ual A	nalysi	s, Per	Cent	Carbon Equivalent	Magnesium , Added,	Treating Temp,	Pouring Temp,
	T.C.	Si	Mn	P	S	T.C.*	Si	Mn	P	S	Per Cent	Per Cent	F	F	
A-6384	1	3.50	2.50	0.70/0.90	0.15/0.18	< 0.05	3.42	2.61	0.73	0.18	0.037	4.29	0.0	-	2540
A-6384	1M	3.50	2.50	0.70/0.90	0.15/0.18	< 0.05	3.51	_		-	-	-	0.55	2600	2500
A-6113	2	3.25	3.25	0.70/0.90	0.15/0.18	< 0.05	3.36	3.22	0.74	0.19	0.028	4.43	0.0	_	2500
A-6113	2M	3.25	3.25	0.70/0.90	0.15/0.18	< 0.05	3.29	-	_	_	-	-	0.55	2600	2470
A-6381	3	3.00	4.00	0.70/0.90	0.15/0.18	< 0.05	3.16	4.01	0.76	0.17	0.028	4.50	0.0	_	2600
A-6381	3M	3.00	4.00	0.70/0.90	0.15/0.18	< 0.05	3.16	_	_	_	_	-	0.55	2600	2500
A-6382	4	2.65	5.00	0.70/0.90	0.15/0.18	< 0.05	2.89	4.95	0.74	0.16	0.028	4.54	0.0	_	2550
A-6382	4M	2.65	5.00	0.70/0.90	0.15/0.18	< 0.05	2.93	_	_	_	-	_	0.55	2630	2500
A-6383	5	2.30	6.00	0.70/0.90	0.15/0.18	< 0.05	2.45	5.94	0.74	0.15	0.033	4.43	0.0	-	2540
A-6383	5M	2.30	6.00	0.70/0.90	0.15/0.18	< 0.05	2.44	_	-	-	-	-	0.55	2640	2470

Carbon was determined on keel-block bars of both treated and untreated irons because experience has shown that carbon has a strong tendency to float out of magnesium-treated hypereutectic cast irons.



Fig. 1—Photo of keel block showing bottom side of casting. The two legs are 21 in. long and 11/4 in. square. The heavy top, which serves as a feeder for the legs, measures 21x51/8x13/4 in.

1 shows one of these keel-block castings. Table 1 contains data relative to the melting, treating, and casting of these five heats. It will be noted that the com-

position of each heat was calculated to give an iron of approximately eutectic composition, that is, with a carbon equivalent of about 4.3. The silicon contents of these heats were close to the calculated values; however, the carbon recoveries were a little higher than had been anticipated, with the result that carbon contents of some of the irons were high.

Microstructures and Mechanical Properties—Two test bars sawed from each keel block were 21 in. long and 1½ in. square. Each bar was marked with a code designation, as indicated in Table 1, and, since two bars were cut from each keel block, one was marked "a" and the other "b".

A specimen for metallographic examination was cut from one bar of each keel-block casting. Metallographic examination showed that all of the magnesium-treated irons contained spheroidal graphite, whereas the untreated irons contained coarse flake graphite. There was a general tendency for the average size of the spheroids in the magnesium-treated irons to decrease as the silicon content was increased. This may have been partly the result of the lower carbon content in the higher-silicon irons. Pearlite was present in both the flake and nodular irons with



Fig. 2-Structure of untreated iron containing 2.61 Si and 3.42 C (Bar 1). Picral etch. Mag. 100×.

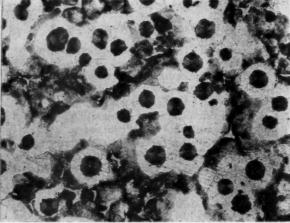


Fig. 3-Structure of magnesium-treated iron containing 2.61 Si and 3.51 C (Bar 1M). Picral etch. 100×.

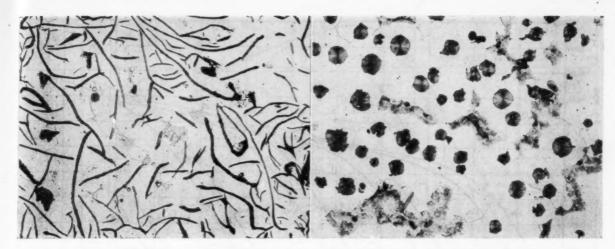


Fig. 4-Structure of untreated iron containing 4.01 Si and 3.16 C (Bar 3). Picral etch. Mag. 100×.

Fig. 5-Structure of magnesium-treated iron containing 4.01 Si and 3.16 C (Bar 3M). Picral etch. 100×.

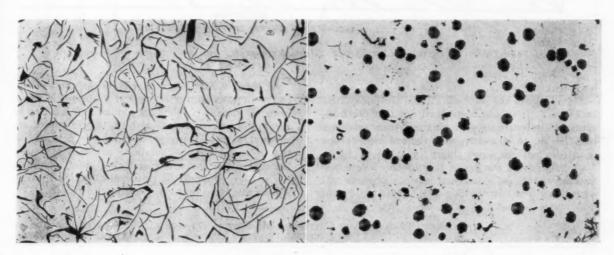


Fig. 6-Structure of untreated iron containing 5.94 Si and 2.45 C (Bar 5). Picral etch. Mag. 100×.

Fig. 7-Structure of magnesium-treated iron containing 5.94 Si and 2.44 C (Bar 5M). Picral etch. 100×.

the lowest silicon content. The amount of pearlite decreased with increasing silicon content, and, in the irons containing 5 per cent silicon or over, the matrix was fully ferritic. Figures 2 through 7 show typical structures observed in both the flake and nodular irons containing 2.61, 4.01, and 5.94 per cent silicon.

The square bars were turned full length to round bars 1.1 in. in diameter. A mark was placed on the end of each bar to show the side of the bar which had been in contact with the keel head. In the transverse and impact tests, this side of the bar was always placed in compression; thus, if any variation in structure occurred across the bar as a result of its orientation in the original castings, this variable would be minimized.

The turned bars were broken on 18-in. centers to determine their transverse properties. Figure 8 shows the load-deflection curves obtained for these bars. It will be noted that at the end of each curve there is a

short line which turns either down or up. If it turns down, it means that the bar was broken at the deflection indicated. If it turns up, it means that the limit of the deflection gage was reached (0.56 in.) before the bar was broken. In those cases, the deflection gage was removed and the bar broken without obtaining a complete load-deflection curve.

It is apparent from these curves that the magnesium-treated irons are considerably stronger and show more deflection than the untreated irons. In addition to this, the curves for the magnesium-treated irons have a straight-line portion, indicating that they deform elastically in the early stages of loading, whereas the curves for the untreated irons show no region of elastic deformation.

One-half of each broken transverse bar was broken in impact on 6-in. centers using an Amsler impact testing machine. Parts of the broken bars were used in making tensile, growth, scaling, and hardness test

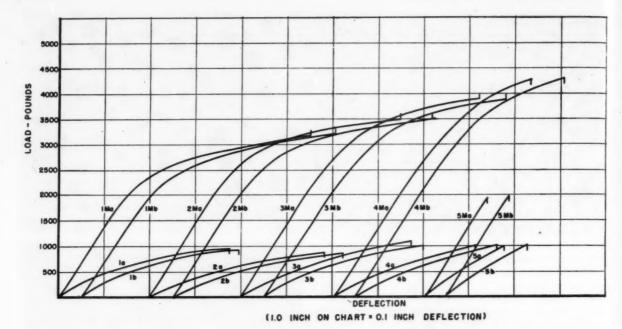


Fig. 8-Load-deflection curves for machined 1.1-in. diam arbitration bars tested on 18-in. centers.

specimens.

The results of the mechanical tests are summarized in Table 2. It should be pointed out that the transverse and impact data were obtained from machined bars 1.1 in. in diameter, and the values for these properties will be lower than for standard 1.2-in.diam bars having the as-cast skin intact. The tensile tests were made on button-end tensile specimens with a gage length of 2 in. and a gage diameter of 0.505 in.

In order to illustrate more clearly the effect of silicon content upon the mechanical properties of these irons, the data in Table 2 are plotted as a function of silicon content in Fig. 9 and 10. These curves are self-explanatory. However, it should be pointed out that spheroidal irons containing as much as 4 to 5 per cent silicon are strong and have sufficient impact strength to eliminate the principal objection to highsilicon cast irons, namely, excessive brittleness.

TABLE 2-RESULTS OF MECHANICAL TESTS ON BARS CAST FROM MAGNESIUM-TREATED AND UNTREATED PORTIONS OF FIVE CAST IRON HEATS

		e Properties in. Bar)		Tensile	Properties		
Code*	Transverse Strength, lb	Deflection,	Impact, Ft-Lb***	Ultimate Strength, psi	Per Cent Elongation in 2 In.	Hardness, Bhn	
la (3.42% C, 2.62% Si)	900	0.372	. 23	12,920	0.0	97	
lb Ditto	900	0.340	20	13,570	0.0		
IMa (3.51% C, 2.62% Si)	4170	***	>112	77,700	4.8	193	
1Mb Ditto	3935	-	>112	77,600	4.5		
2a (3.36% C, 3.22% Si)	870	0.390	19	12,440	0.0	104	
2b Ditto	865	0.380	19	12.690	0.0		
2Ma (3.29% C, 3.22% Si)	4350	-	111.5	73,150	3.2	197	
2Mb Ditto	4360	_	>112	74,500	3.5		
3a (3.16% C, 4.01% Si)	1060	0.375	20	13,420	0.0	108	
Bb Ditto	1010	0.360	19.5	13,880	0.0		
3Ma (3.16% C, 4.01% Si)	4675		70	75,750	1.0	208	
BMb Ditto	4595	-	69.5	77,300	1.2		
la (2.89% C, 4.95% Si)	1020	0.315	17.5	14,660	0.0	131	
b Ditto	1020	0.320	18.0	14,920	0.0		
Ma (2.93% C, 4.95% Si)	4340	0.435	57	90,450	0.5	248	
Mb Ditto	4320	0.402	39	90,450	0.7		
ia (2.45% C, 5.94% Si)	1170	0.180	16	13,680	0.0	171	
b Ditto	1020	0.180	18	12,430	0.0		
Ma (2.44% C, 5.94% Si)	1980	0.142	20	47,650	0.1	293	
Mb Ditto	2000	0.145	17	43,300	0.0		

<sup>The "M" indicates magnesium treatment. Duplicate bars (a and b) were tested.
Where no value is given, the deflection exceeded the limits of the gage (about 0.56 in).
The capacity of the Amsler machine was 112 ft-lb.</sup>

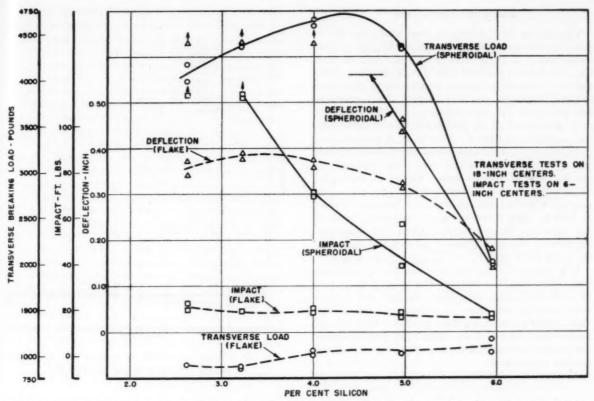


Fig. 9-Transverse and impact properties, as a function of silicon contents for five unalloyed cast irons containing flake or spheroidal graphite. The test bars were machined from keel-block castings.

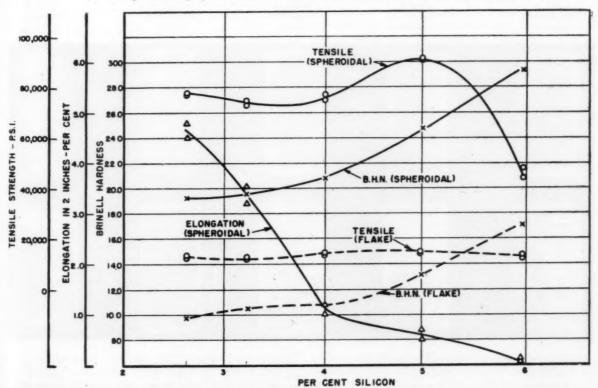


Fig. 10--Tensile properties and hardness as a function of silicon content for five unalloyed cast irons containing flake or spheroidal graphite.

Scaling and Growth Tests

The scaling rate of each of these irons was determined as follows: Scaling test specimens 1 in. in diameter and 2 in. long were machined from the test bars, the surfaces polished with 100-grit abrasive cloth, and the specimens accurately weighed on an analytical balance before the test was started. Each specimen was placed in an open fire-clay crucible which, in turn, was placed in an electrically heated muffle furnace operating at 1600 F. After 24 hr at temperature, the specimens were removed from the furnace, cooled, and reweighed, after which they were returned to the furnace for another cycle. Figure 11 shows the increase in weight calculated to grams per square centimeter of surface area plotted as a function of time at 1600 F. Each curve in Fig. 11 represents the average value for two specimens. Figure 12 shows the appearance of a representative specimen of each iron after 240 hr at 1600 F.

It is apparent from the curves in Fig. 11 that cast iron, which has been treated with magnesium to promote the formation of nodular graphite, has a much lower scaling rate than an untreated iron of the same analysis containing flake graphite. It is also apparent that, in the irons containing flake graphite, the scaling rate decreased progressively as the silicon content is increased.

It should be pointed out that the curves in Fig. 11

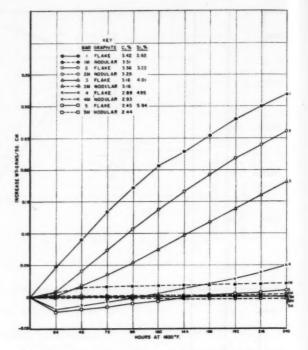


Fig. 11-Change in weight of scaling test specimens during exposure to air at 1600 F.

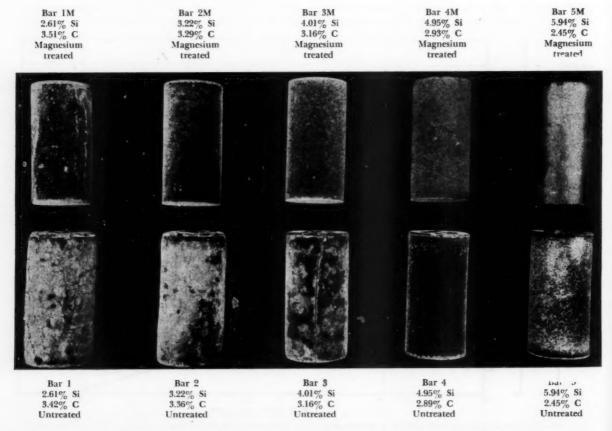


Fig. 12-Appearance of scaling test specimens after 240 hr at 1600 F.

represent the actual change in weight of the specimens calculated to grams per square centimeter of surface area. Thus, they represent a summation of the increase in weight resulting from oxidation and the decrease in weight resulting from decarburization. In this respect, it is of interest to note the shape of the curves for the flake irons containing 5 and 6 per cent silicon (Bars No. 4 and 5). The weight loss through decarburization in the first 24 hr was greater than the weight increase by oxidation. After the first 24-hr cycle, the specimens slowly gained weight. Apparently, most of the decarburization took place during the first 24 hr and, with further time at temperature, oxidation was the dominant reaction. The magnesium-treated iron containing 4 per cent silicon (Bar 3M) had a slight but consistent loss in weight throughout the test. In this case, the weight gained by scaling was slightly overshadowed by the weight loss from decarburization. The weights of the nodular iron specimens containing 5 and 6 per cent silicon remained almost constant throughout the test. After 240 hr at temperature, the 6 per cent silicon nodular iron had a thin film of scale, while the scale on the 5 per cent silicon iron was only slightly heavier. Thus it can be assumed that both the oxidation or scaling rate and decarburization rate for these two irons are

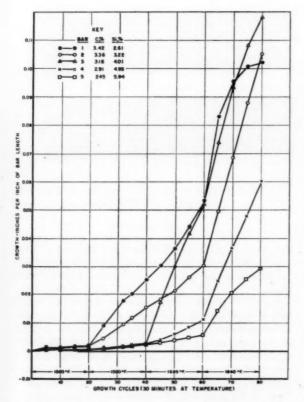


Fig. 13—Growth of cast iron containing flake graphite, subjected to cyclic heating. Each cycle consisted of heating to the indicated temperature, holding at temperature for 30 min, and air cooling. Initial length of test bars was 4 in. Each curve represents the average of two duplicate specimens.

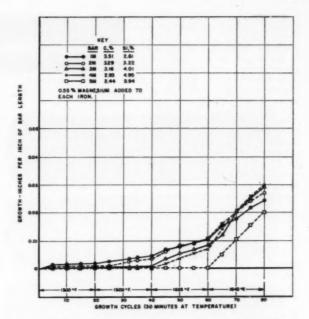


Fig. 14—Growth of cast iron containing spheroidal graphite subjected to cyclic heating. Each cycle consisted of heating to the indicated temperature, holding at temperature for 30 min, and air cooling. Initial length of test bars was 4 in. Each curve represents the average of two duplicate specimens.

quite low. In fact, the scaling resistance of the 6 per cent silicon iron was about equivalent to that of the stainless steel identification bands used on the growth test specimens described later in this paper.

Tests were conducted to determine the effect of variations in silicon content and changes in graphite structure upon the growth resistance of cast iron. A growth test specimen, I in. in diameter by 4 in. long, was machined from each of the arbitration bars. A stainless steel button was inserted in each end of each bar and all length measurements were made between these stainless steel buttons, so that accurate measurements could be made in spite of scale that had formed.

The growth tests consisted of heating the test bars to a predetermined temperature, holding at temperature for 30 min, and air cooling to between 300 and 400 F, after which the specimens were returned to the furnace for another cycle. After each five cycles, the bars were cooled to room temperature and the length measured to the nearest thousandth of an inch. A test specimen identical to the others, but one which was not subjected to cyclic heating, was used as a standard to calibrate the micrometer. Testing temperatures of 1300, 1500, 1625, and 1840 F were used, each bar being subjected to 20 cycles at each temperature in the order given. Figures 13 and 14 show graphically the length increase calculated to inches per inch of bar length plotted as a function of test cycles.

In an earlier investigation, it was shown that the growth of cast iron during cyclic heating is small,

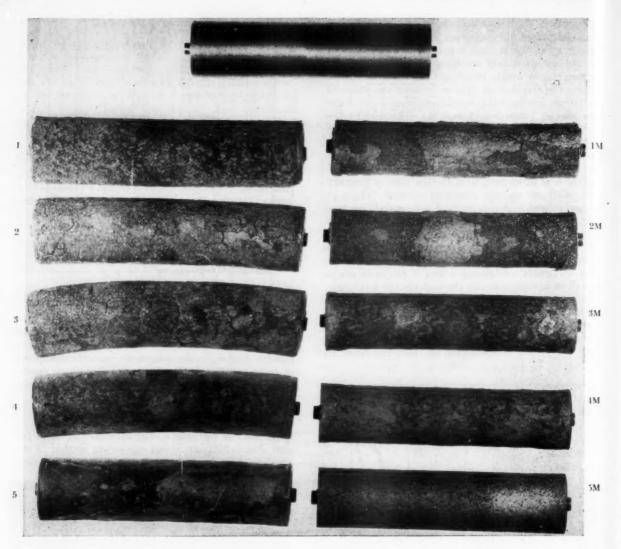


Fig. 15-Representative growth test specimens of each iron as they appeared at the end of the growth tests.

The top specimen is one that was not subjected to cyclic heating.

provided the critical temperature is not exceeded. However, if the iron is alternately heated and cooled through the critical temperature range, then growth proceeds at a rapid rate. Increasing the silicon content of the iron raises the critical temperature, and, therefore, raises the temperature above which rapid growth occurs.

The data presented in Fig. 13 substantiate these earlier observations. In these irons, containing flake graphite, the temperature above which rapid growth occurs is progressively raised as the silicon content is increased. A comparison of the curves in Fig. 13 and 14 show that nodular cast irons are more resistant to growth than are irons of identical composition containing flake graphite. The nodular irons are similar to the flake irons in that appreciable growth does not occur unless they are heated and cooled through the critical temperature range. The slight growth at 1300 F shown by both the flake and nodular irons with the

two lowest silicon contents was primary growth caused by graphitization of the pearlite. The irons with higher silicon contents were almost fully ferritic, and, consequently, no primary growth occurred in these irons.

Figure 15 shows a representative growth test specimen of each iron as it appeared at the end of the test. It is interesting to note the reduced diameter of most of the bars containing spheroidal graphite. The scale which formed on these bars at high temperatures flaked off when the bars were cooled, with the result that they decreased in diameter. On the other hand, most of the scale which formed on the iron containing flake graphite remained in place throughout the complete test. The 4 and 5 per cent silicon irons containing flake graphite bowed considerably during the growth test. This may have been caused by differences in growth rate between the two sides of the iron resulting from differences in graphite structure.

M

M

M

M

0

l,

it

e

t

e

e

It should be remembered that these test specimens were cut from the legs of keel-block castings, so that some variation in graphite structure occurred across the bar, the section next to the keel head containing the coarser graphite.

Summary

The addition of magnesium was effective in producing spheroidal graphite in cast irons containing from 2.6 to 6 per cent silicon. Untreated portions of these irons contained coarse flake graphite. The change from flake to spheroidal graphite was accompanied by a marked increase in transverse and impact properties of the irons containing as high as 5 per cent silicon. As the silicon content was increased to 6 per cent, these properties were lowered and approached the values obtained for irons of the same composition containing flake graphite. Additions of silicon progressively increased the hardness of these irons and the magnesium-treated irons were consistently 100 Brinell numbers harder than were the untreated irons. In general, the results of these mechanical tests have shown that spheroidal cast irons containing from 4 to 5 per cent silicon are much stronger than any of the other irons tested, and, at the same time, they have sufficient impact resistance to eliminate the principal objection to high-silicon irons, namely, their extreme brittleness.

The results of scaling and growth tests on these cast irons showed that those irons containing spheroidal graphite are considerably more resistant to growth and scaling than are the irons containing flake graphite. The 4 and 5 per cent silicon cast irons containing spheroidal graphite have a combination of mechanical properties and resistance to growth and scaling which should make them quite useful irons for many industrial applications. The 6 per cent silicon iron containing spheroidal graphite has even greater resistance to scaling and growth at testing temperatures up to 1625 F. The growth of this iron was almost unmeasurable and its scaling resistance at temperatures up to 1840 F was about equivalent to that of 18-8 stainless steel.

Acknowledgment

The authors wish to thank the Jackson Iron and Steel Co., sponsors of this investigation, for permission to publish these data. They also wish to express their appreciation to Dr. C. H. Lorig and Mr. G. K. Manning, respectively, Assistant Director and Supervisor of Metallurgical Engineering of Battelle Memorial Institute, under whose supervision this research was conducted.

Bibliography

1. W. H. White and A. R. Elsea, "Growth and Scaling Characteristics of High-Silicon Cast Iron," Transactions American Foundrymen's Society, vol. 57, pp. 459-474 (1949).

2. K. D. Millis, et al, U. S. Patent No. 2,485,760.

3. The Technology of Magnesium and Its Alloys, (Translation of a German book by Dr. Adolf Beck), p. 477.

DISCUSSION

Chairman-A. P. GAGNEBIN, International Nickel Co., New York.

Co-Chairmen-T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa., and W. B. McFerrin, Haynes Stellite Div., Kokomo, Ind.

F. B. Rote (Written Discussion): ¹ This is a very interesting paper in which is presented information of importance to potential or present users of ductile cast iron. Silicon is the most potent alloying element for ferrite, and as such can exercise considerable control on the properties of ferrous materials.

At the University of Michigan we have investigated, over a rather long period of time, the properties of ductile cast iron, in a research project sponsored by the International Nickel Co., Inc., and have developed considerable data on the influence of compositional variables on the mechanical properties. Attempts to compare the results of our investigations with those reported by the authors were only partially successful, in that we found higher levels of mechanical properties than did the authors in irons of essentially the same composition. The most significant differences were in reported ductilities, as measured by elongation in a tensile test.

Table 3 shows the comparative properties of the irons reported by the authors and irons of similar compositions made at the University of Michigan. These are plotted in Fig. 16. The two lower groups of irons were cast in 1-in. keel blocks and tested

TABLE 3-COMPARATIVE MECHANICAL PROPERTIES OF DUCTILE CAST IRONS

				Tensile Properties						
Chemical Composition % C % Si		Source & Heat Treatment		Tensile Strength, psi	0.2% Offset Yield, psi	Elon- gation				
3.42	2.62	None	193	77,650		4.6				
3.36	3.22		197	73,825		3.3				
3.16	4.01	White, Rice,	208	77,525		1.1				
2.93	4.95	& Elsea	248	90,450		0.6				
2.44	5.94		293	45,475		0.05				
3.72	1.21	None	233	103,000	53,000	7.5	5.0			
3.45	1.90		242	109,750	65,500	9.5	6.6			
3.42	2.27	U. of Mich.	208	95,000	59,000	9.0	6.2			
3.41	3.36		232	100,000	76,625	6.0	3.7			
2.75	4.43		235	92,250	82,500	4.5	2.3			
3.72	1.21	Annealed	114	49,500	32,500	27.5	23.7			
3.45	1.90		137	58,250	41,000	27.0	29.1			
3.42	2.27	U. of Mich.	142	62,800	46,000	23.0	20.0			
3.41	3.36		187	81,625	65,250	14.0	9.9			
2.75	4.43		222	96,750	77,000	11.5	8.2			

The first series of data are from White, Rice, and Elsea. The latter data are from University of Michigan.

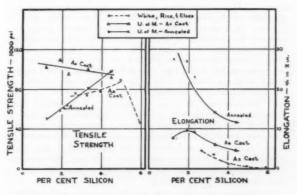


Fig. 16

¹ Asst. Prof. Dept. Metal Processing & Chem.-Met. Engr., University of Michigan, Ann Arbor, Mich.

either as-cast or as-annealed to a completely ferritic matrix. It will be noted that the most marked deviation in the two sets of data, those presented by the authors and those obtained at the University of Michigan, was in elongation. The highest elongation reported by the authors was 4.6 per cent in the 2.62 per cent silicon iron. The ductility decreased with increasing silicon content to essentially 0 at 5.94 per cent silicon. Data secured at the University showed a minimum of 4.5 per cent elongation at 4.43 per cent silicon, and elongation as high as 9.5 per cent at lower silicon contents.

Likewise, considerably higher tensile strengths were observed in irons produced and tested at the University of Michigan. The differences in tensile strengths in 1-in. keel castings from the two sources are about 25,000 lb at the 3 per cent silicon level and about 10,000 lb at the 4.5 per cent silicon level. No data are available at the University on ductile irons with over 5 per cent silicon, but it is difficult to reconcile the marked decrease in tensile strength with increasing silicon content over 5 per cent, such as is shown by the authors.

The microstructure in the castings produced at the University were very similar to those reported by the authors. The irons contained about 1.25 per cent nickel, which was secured from

the alloy used in making the magnesium addition.

Attempts to determine the influence of an alloy such as silicon on the mechanical properties of a cast iron encounter many complexities, unless a consistent matrix structure is produced in all castings, regardless of composition. Silicon has a two-fold effect on the structure of the irons, in that increasing silicon content decreases the quantity of pearlite in the matrix and at the same time increases the hardness and strength of the ferrite. Thus, complex changes in properties with variations in silicon content will be produced. Proper evaluation of the effect of silicon can be obtained, if a consistent ferritic matrix structure is produced. This is obvious from examination of the plotted data for the irons cast at the University of Michigan and fully annealed. It will be noted that the strength increased rapidly with increasing silicon content and decreasing carbon, and that the increase in strength was linear. At the same time, the elongation decreased with increasing silicon content from a high of 27 per cent to a low of 11.5 per cent.

Considering the data secured at the University, essentially the same conclusions as those presented by the authors would be reached, although the level of properties at the terminal points

in the silicon axis would be considerably higher.

C. C. REYNOLDS: What was the composition of Prof. Rote's base metal with respect to manganese and phosphorus?

MR. ROTE: The manganese content was 0.311 or 0.312 and

the phosphorus was 0.03 or 0.04 per cent.

MR. REYNOLDS: I would like to congratulate the authors on the excellent work they have done on scaling and growth. We have been doing work on ductile iron with a wide range of chemical analysis. We differ with Mr. White and also with Prof. Rote considerably in physical properties obtained. I think this is largely because our manganese and phosphorus levels were considerably different.

In the paper the manganese was very high (about 0.7 per cent). The phosphorus level was also high, which may be responsible for the low elongation. In Prof. Rote's work with 0.3 per cent Mn and lower phosphorus, we see that the properties improve. For example, at the 3½ per cent Si level, the authors of this paper had an elongation of less than 5 per cent. Prof. Rote increased that to about 13 per cent. We have been able to obtain as-cast data with 20 per cent elongation. It is not a simple matter of silicon content alone. One must take into account all the other elements in cast iron.

R. Schneidewind (Written Discussion): The authors have presented valuable data to the foundrymen on compositions of heat and scale-resistant irons with good mechanical properties. The data prove the excellent scale resistance beyond doubt.

The mechanical properties are on the low side due most probably to the unorthodox nodulizing procedure. The strengths and ductilities would be greater in each of the tests with a more common nodulizing alloy.

Nevertheless the results seem to bear out very well the effect of silicon on ferrite. In the accompanying graph (Fig. 17) are plotted the results of Yensen* in dotted lines. These values are for fully-annealed pure iron-silicon alloys, carbon-free. These figures show the solid solution effect of silicon with a serious discontinuity at 2.7 per cent Si which was noted to a less serious degree by others who used commercial materials. The strength and ductility act normally up to 2.7 per cent Si. Between 2.7 and 4.5 per cent Si the strength still increases but the ductility is lower. Above 4.5 per cent Si both strength and ductility drop drastically.

The data of the authors are plotted on the same graph. The values for modulus of rupture parallel very closely the curve of Yensen for tensile strength. The White data for tensile is less regular but these and the ductilities (plotted 10x actual) do

follow the Yensen data broadly.

Since these irons were designed to be used for high temperature service, it would have been wise to anneal them prior to testing for mechanical properties. However, due to the relatively high silicon contents, pearlite decomposition is practically complete in the case of higher silicon. The iron with lowest silicon has an appreciable amount of pearlite which could readily account for 10,000 psi greater tensile strength than if the structure were ferritic. Completely ferritic nodular irons containing about 2 per cent silicon will average 60,000 to 70,000 psi.

The ratio of tensile strength to Brinell hardness varies with silicon content starting with a ratio of 405 at 2.62 per cent silicon which is normal for nodular iron. The ratio then levels off to 365-372 up to 4.95 per cent and drops suddenly to 154 at

5.94 per cent silicon.

Mr. White (Reply to Dr. Schneidewind): The structures of our treated irons were all essentially nodular. We deliberately

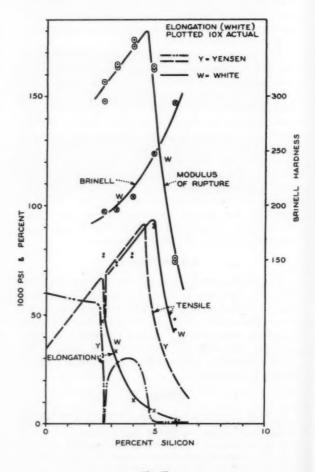


Fig. 17

² Massachusetts Institute of Technology, Cambridge, Mass. ³ Prof. Met. Engr., University of Michigan, Ann Arbor, Mich.

^{*}T. D. Yensen, "Magnetic and Other Properties of Iron-Silicon Alloys, Melted in Vacuo," Univ. of Illinois Experiment Station, Bulletin 83 (1915).

ľ

0

f

HARDNESS

BRINELL

avoided the use of nickel and other alloys in order that no other variable than silicon content should be present. If our method of nodulizing was unorthodox, it did accomplish what we intended. The difference in physical properties obtained by the University and ourselves could be due to the nickel but this we question. It is more likely that differences of other elements such as manganese and phosphorus may be responsible for the variation.

C. F. WALTON; 4 I wish to add my congratulations to the authors on a fine paper. The information presented appears to have considerable practical application. It is interesting to note the change in scaling characteristics between the samples with flake and spheroidal graphite. Do the authors feel that this was just a difference in the rate of scaling or was there also a change in the Pilling and Bedsworth ratio (ratio of densities of parent metal and scale)?

MR. WHITE: I wish to thank all who have commented on this paper. I think Mr. Reynolds answered almost identically what I would have said in reply to Prof. Rote. I think we must remember in this work we did not attempt to show the finest properties which can be obtained by the use of nodular or spheroidal structure. The one attempt made in this work was to evaluate the difference between nodular and plain cast iron at different silicon levels. We did not pretend that our phosphorus, manganese or any other element was the ideal for the greatest tensile strength, elongation or any other property. It is merely a question of the difference which you will obtain between the two structures in iron which would otherwise be a normal

foundry composition. I feel that it is quite possible that you can obtain much higher properties in many cases than we obtained. This procedure worked out very well for this comparison and, as I mentioned before, we were primarily interested in these properties, using silicon as the one variable.

MR. ELSEA (Written Reply): Mr. Walton asked for comments on the effects of structural changes on the Pilling and Bedworth ratio. The Pilling and Bedworth rule states that if the specific volume of the oxide is equal to or greater than the specific volume of the base metal then oxidation proceeds according to the relationship:

 $X^2 = Kt$

where X is the weight or thickness of metal which is oxidized, t is time, and K is the rate constant. The rule says nothing about structure, but one would expect that structure would have a marked influence on the constant K. For example, a white iron would not be expected to scale at the same rate as a graphitic iron of the same analysis.

In the case of gray iron, oxidation takes place at the surface and also along the interface between the matrix and the flake graphite. Actually the internal oxidation in a flake iron may be responsible for considerably more weight change than surface oxidation. In nodular iron this internal oxidation occurs much more slowly or not at all because there are no channels for rapid diffusion of oxygen. In addition to this, the differences between nodular and flake iron, with respect to both elasticity and thermal expansion, will have an effect upon the cracking tendencies of the scale. All other factors equal, the greater the tendency for scale to flake off and consequently the faster would be the scaling rate.

Asst. Prof., Case Institute of Technology, Cleveland.

U. S. NAVY NON-FERROUS DEVELOPMENT PROGRAM

By

Clyde L. Frear*

THE NAVY, especially during a national emergency, is an important purchaser and consumer of castings. Approximately three per cent of every combatant vessel is made up of castings, the greater proportion of which are procured from commercial sources. Although nearly every naval shipyard has a foundry and some of these foundries are comparatively large, their main purpose is to produce those castings which cannot be procured from private foundries in the time available to effect necessary repairs and overhauls.

Because of the peculiar operating conditions involved aboard ship, such as the necessity to withstand high pressures in small spaces, vibration from high speeds, shock from hits and near misses in combat, the necessity for holding weight to a minimum consistent with the loads to be carried, in other words the utmost in reliability per unit weight, specifications and inspection requirements are rather strict insofar as physical properties and freedom from defects are concerned.

Early Casting Development

Because of excessive failures experienced in steel castings shortly following World War I, and the difficulty of obtaining castings which would stand up in service, the Navy Department instituted a rather extensive and exhaustive program of development at the Naval Research Laboratory. As an outcome of this development, several papers by personnel of this laboratory and from one of the naval shipyards1,2,3,4,5 presented what was evidently the first really quantitative treatises on steel casting techniques. These papers did much to lay the groundwork for considerable subsequent work by private foundries and technical societies which tended to place the steel casting industry on a scientific basis and opened the way for a long program of development on this phase of the subject, as well as to indicate development procedures for other types of castings.

The use of radium for gamma radiography was also developed at the Naval Research Laboratory. For a number of years, this was the only nondestructive method of testing castings for the presence of internal defects as no X-ray equipment had been developed with sufficient power to penetrate heavy sections. The Navy has also been instrumental in the further development and standardization of the various other methods of nondestructive testing.

The Navy is a large user of non-ferrous castings, chief among which are the normal tin bronzes such as gun metal, hydraulic bronze, and ounce metal. These metals find their chief use in piping systems for salt water, also for fuel systems where corrosion resistance is required in conjunction with ductility to resist breakage under shock. Laboratory tests, also use in actual service over a long period of time, have shown that the tin bronzes possess better resistance to corrosion by sea water in rapid motion than do any other non-ferrous metals or alloys and, for this reason, are usually specified for this particular service.

As has been found by every other user of tin bronze castings, one of the difficulties which is experienced almost continuously with this class of material is seepage, weeping or leakage of pressure castings, such leakage occurring sometimes through unmachined walls. but more often through those areas which have been machined on one or both sides, and especially through bolt holes or through valve seats. If this leakage could be discovered before any machining operations were performed, rejections could be made at the foundry and costs kept to a minimum. Usually, however, such leakage does not show up until a considerable amount of machining has been accomplished, sometimes not until the part has been completely machined, and often not until it has been mounted in the final assembly. In such cases, it is not hard to realize that the losses due to leakage may amount to several times the first cost of the castings. If we could be assured of sound castings, we could often pay several times the usual purchase price and still save money on the completed job, not to mention a considerable saving in

During World War II, the Navy Department often

Senior Materials Engineer, Bureau of Ships, Department of the Navy, Washington, D. C.

The opinions expressed herein are those of the writer and do not necessarily express the views of the Department of the Navy or the United States Government.

had other considerations besides cost. In many cases, the extensive shipbuilding program was in serious danger of being delayed because of the non-receipt of parts containing bronze castings, the non-delivery usually being due to castings which showed the usual leakage after being machined. A considerable amount of the writer's time was spent at various foundries in an attempt to obtain a quick correction of the foundry difficulties in order to produce castings of acceptable soundness. That he was able to succeed in this mission in numerous cases was due mostly to good fortune as a study of the foundry literature gave little help in the particular cases involved. True, there had been a large number of papers published, but much of the information was contradictory, and little of it was of mathematical precision. In plain and simple words, founding, especially of bronzes, is still mostly an art, little science, with a large amount of luck, often bad, thrown in.

In order to obtain usable castings in many instances, it became necessary to resort to impregnation. It is true that impregnation will often aid in producing acceptable castings, provided the leakage is due to small amounts of scattered shrinkage, but it appears to the writer that it is a rather poor substitute for soundness, provided it is possible to obtain satisfactory absence from defects. Usually this is possible, sometimes at a slight increase in cost, perhaps at no extra cost. In one case in the writer's experience, the foundry procedure required to produce a perfectly sound casting was actually about half the previous cost of making a poor casting.

Non-Ferrous Development

In an attempt to improve the general quality of bronze castings, the Bureau of Ships has initiated a rather comprehensive series of development projects aimed, first at learning the real cause of defects, especially the mechanism by which the interdendrite shrinkage forms; secondly, to develop methods of eliminating such shrinkage; and finally, to standardize, insofar as practicable, the foundry techniques necessary to produce castings with the desired absence from defects. Except in a very few cases where the information derived from this development might affect the national security, the results will be published in the technical press in order that industry may have the benefit of the findings. The savings resulting from procurement of castings having improved soundness should amount to many times the cost of the development.

As stated above, interdendritic shrinkage is probably the most common type of defect found in tin bronzes. It is sufficiently common that it is considered to be an inherent defect in this type of non-ferrous alloy. If this is the case, however, how can we account for heats which every foundry obtains now and then, which shrink with a deep pipe in the risers and contain nothing but gross shrinkage, much as occurs in insufficiently fed steel and brass?

The writer has noted a tendency to blame interdendritic shrinkage upon the wide solidification range as shown in the copper-tin equilibrium diagram, but

here again, this does not account for those few melts having the same composition and made by the same foundry techniques, insofar as we can determine, but which have shrinkage tendencies like the more normal shrinking alloys, and show no traces of interdendritic shrinkage. It seems evident that something has happened in these heats which has changed the mechanism of the dendritic growth, or, if not the shape of the dendrite, at least the speed of individual dendritic growth. It appears possible that we may find some element or combination of elements which, when present in small amounts, can change entirely the mechanism of dendritic growth from apparently rapidly growing branches to a more solid growth which can permit feeding of the whole casting, provided, of course, that we provide this necessary feeding. Hesse⁶ found certain indications pointing to this possibility when he added nickel to gun metal. The chief disadvantage of using nickel, however, is that if we use enough to correct the shrinkage characteristics entirely, the tin content will be reduced below the percentage required to give the desired resistance to salt water corrosion. This statement is based upon the usual replacement of a definite percentage of tin by an equal percentage of nickel.

Development Projects

The following development projects, initiated by the Bureau of Ships, are now in effect at various naval shipyards:

 Determination of rates of solidification of bronze castings and methods of their control,

2. Study of convection currents in molds,

 Study of various methods of melting, molding and pouring to eliminate defects,

4. Application of radiography to bronze castings. This series of projects was initiated in an attempt to determine the mechanics of solidification, chiefly concerning the method of dendritic growth; also to determine the possibility and the practicability of altering the method of growth, in order to provide improvement in feeding between the dendritic branches; finally to study the cooling of metal in the molds and to derive methods of controlling the directional solidification. It has been the experience of many foundrymen that certain castings designed to provide correct directional solidification, and molded to provide complete feeding, still contain shrinkage after the castings have solidified. This may be due to the failure of the metal to follow the expected path through the mold cavity, thus upsetting the thermal gradient which had been carefully calculated and applied. The possibility of this is shown by moving pictures on gating produced by Johnson and Baker at the Naval Research Laboratory.7,8 Step gates have been used for years, supposedly to provide hot metal at the top of a casting and still utilize certain advantages inherent in bottom pouring. Their use was predicated upon the belief that, when the metal reached the next higher ingate, it would then enter the mold cavity through this higher gate and cease to flow through the lower one. The moving pictures mentioned showed that, contrary to this assumption, the

metal would continue to flow through the lower ingate, returning to the down gate through the upper gates. This work was one of the first to show positively what happens when metal is flowing into a mold. Considerably more positive information is needed before we can calculate with any degree of certainty just what will happen to the metal within the mold cavity, and before we can lay out the correct system and sizes of gates, sprues and risers for any specific design.

The third project mentioned above is being conducted in regular production work and consists in determining the effects of the different variables upon the properties and the soundness of the various castings, as determined by regular physical tests, by radiography, by microradiography, by density measurements and by hydrostatic tests, in the as-cast condition and after removing specific thicknesses from one or both surfaces. In this connection, low voltage x-ray examination of cross-sections (1/8 to 1/4 in. thick), microradiography and pressure tests after machining appear to be the most effective methods yet devised for determining the presence of internal shrinkage and its distribution within the casting.

These projects are being continued, the rate of progress depending upon the work load at the particular activities and also to a great extent upon the funds available. One outstanding paper, in the writer's opinion, which resulted directly from one of these projects was by Ames and Kahn⁹ which presented proof of the narrow pouring range for the tin bronzes if we are to obtain castings of even comparative soundness. These findings were essentially further proof of opinions which had been published previously.

Temperature of Molten Bronze

Perhaps the most important thing accomplished by this work so far is the realization of the fact that we have no practical equipment for accurately measuring the temperature of the molten bronze. True, we have immersion pyrometers which are more or less portable but which can be used but a comparatively few seconds at a time, for determining metal temperature in the ladle only, and usually only before starting to pour. We have no equipment which is satisfactory for determining temperatures of the metal while it is in the furnace being melted or heated, or during the actual pouring operation. It is intended to initiate a project to develop equipment for the accurate and continuous measurement of the metal in the furnace, and to record this temperature during the melting cycle; also to measure the temperature in the ladle during the pouring in order that pouring can be stopped if the temperature falls below the satisfactory range for the particular casting being poured.

It has been fairly well established that good metal is necessary if we are to obtain good castings, even though it does not necessarily follow that good metal will always produce good castings. Because of the necessity for good metal, it appears highly desirable to develop a test to determine the quality of the metal while it is in the furnace, and before it has been poured into the molds. To be practical, such a test would have to be low in cost and be rapidly com-

pleted, yet give a positive indication as to whether the metal in the furnace is capable of producing a good casting, provided of course, that all other conditions are correct. These requirements for speed and economy would naturally preclude the use of tensile test bars for this purpose, but would probably permit the use of hydrostatic or other pressure tests and might possibly permit the use of x-rays.

Improvement in Feeding

As many of the difficulties experienced with bronze castings could be traced to the lack of complete feeding, the Bureau of Ships, during World War II initiated at the Naval Research Laboratory, a project to develop methods of keeping bronze risers open to the atmosphere, thereby permitting feeding as long as the casting itself contained any liquid metal. The direct result of this project was the development of the gypsum riser sleeve reported by Taylor and Wick.10 Although this sleeve gave excellent feeding in most cases, it could not be used with high-nickel bronzes or with monel metal, or with alloys requiring high pouring temperatures. There are two other disadvantages to the use of these gypsum sleeves, first, the foundry using them has to set aside a part of the foundry or the core room in which to make them, and an oven in which to bake them. This would be a distinct handicap in some of the smaller foundries where space is at a premium. The chief disadvantage, however, lies in the extreme fragility of these sleeves and the consequent ease with which they revert to a powder, making it almost impossible to keep the gypsum out of the sand system. The presence of much of this material in the sand would have a very deleterious effect upon most molten metals coming into contact with it, especially after it had had a chance to become moist in the sand heap.

With the disadvantage of the gypsum sleeve in mind, the various naval shipyards were requested to attempt to find another material which would be suitable for the purpose, but which would be free from the inherent disadvantages mentioned. The Puget Sound Naval Shipyard¹¹ reported suitable mixtures of infusorial earth, bentonite and plastic as having good insulating properties, suitable strength, high temperature refractoriness, and containing nothing which would combine chemically with the molten metal. Doubtless there are other materials which would be suitable for the purpose, which could be produced commercially and shipped in volume and which could, therefore, be purchased ready made in suitable sizes and shapes, thus eliminating the need of preparing them in the foundry.

te

to

be

in

Sic

the

me

Molding Materials

It is probably true that moisture in the mold may have considerable effect upon the soundness of a bronze casting poured into that mold. Tests have further indicated that the presence of water of hydration and perhaps even of hydrocarbons or other organic compounds containing hydrogen remaining in baked cores and dry sand molds may cause porosity. If such is the case, it appears that something should be done d

15

18

e

11

e

0

e

15

of

d

g

1

g

e

d

5-

e

d

v.

n

is

IS

t

n

0

1-

n

1

25

8

1-

h

E

d

g

r

n

d

h

to develop a molding material, or a mold and core binder which would contain no water, either free or combined, or other hydrogen compounds at the time the metal is poured into the mold cavity. The so-called "C" process, which originated in Germany, but which is being further developed in this country may be a partial answer to this problem, even though plastics used as bonding materials contain hydrogen compounds. These may not be harmful, perhaps because they may not decompose to form nascent hydrogen, or because such decomposition may lag behind the solidification of the alloy poured into contact with them. The Navy's part in the development of this process has been partially reported by Ames, Donner and Kahn.¹²

Radiographic Examination

Radiographic examination, by x-rays and gammarays has proved itself to be a very useful inspection tool to determine the presence of internal defects in steel castings. Considerable development work is being conducted at present to determine the applicability of radiography to the inspection of bronze castings. It is known that it is capable of showing the presence of massive defects such as gas holes and large shrinkage cavities. Because the rays fail to indicate heterogeneities or discontinuities smaller than approximately two per cent of the wall thickness of the part being tested, they will probably fail to indicate the individual microscopic shrinkage cavities which make up most of the interdendritic shrinkage. By careful work and accurate interpretation of results, it has been found possible to detect the presence and show the extent of such shrinkage porosity which shows as a general darkening of the affected areas on the radiographic negative. Considerably more development will be required before radiography can become a completely suitable tool for the inspection of tin bronzes.

One very successful use, however, is the application of radiography to determine the relative soundness of a bronze casting which leaks or seeps under pressure tests and which it is desired to impregnate. Pressure tests by themselves prove only whether a casting leaks or is apparently sound. It gives no indication as to whether the leakage is due to microscopic porosity, to gross shrinkage cavities, to cracks or to other defects. The old procedure of permitting impregnation of "weepers" and "seepers," but rejecting outright the so-called squirters has been found to be an unsuitable criterion. Numerous castings which show but slight tendency to seep under high pressure have been shown to consist of two chilled skins with but a slight lattice of metal in the remainder of the wall. Other castings, which leaked fairly large streams were found to contain very slight shrinkage porosity or a single microscopic channel through the wall. Radiography has been found to give an accurate indication of the actual internal continuity, upon which indication, permission to impregnate can be safely based.

Conclusions

The above statements have been offered to express the writer's thoughts and opinions concerning development which is considered necessary or desirable at the present time to effect immediate improvement in the general quality of bronze castings. The projects listed or described do not include all desirable projects, nor is the Bureau of Ships at present conducting actual work on all of them. Furthermore, the listing of these desirable projects should not be construed as a desire on the part of the government to be the sole agency to conduct research and development on these specific phases. The Bureau of Ships will be pleased to obtain suggestions for additional development projects which will aid in the general improvement in quality and procurement of bronze castings.

References

1. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction in Steel Castings, II, Free and Hindered Contraction of Cast Carbon Steel," TRANSACTIONS, American Foundrymen's Association, vol. 42, pp. 440-470 (1935).

2. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction in Steel Castings, IV, The Free and Hindered Contraction of Alloy Steels," Transactions, American Foundrymen's Association, vol. 43, pp. 1-32 (1936).

3. C. W. Briggs, R. A. Gezelius and A. R. Donaldson, "Steel Casting Design for the Engineer and the Foundryman," Transactions, American Foundrymen's Association, vol. 46, pp. 605-696 (1939).

4. S. W. Brinson and J. A. Duma, "Application of Controlled Directional Solidification to Large Steel Castings," Transactions, American Foundrymen's Association, vol. 48, pp. 225-282 (1941).

 S. W. Brinson and J. A. Duma, "Studies on Centerline Shrinkage in Steel Castings," Transactions, American Foundrymen's Association, vol. 50, pp. 657-765 (1943).

6. A. H. Hesse, "Some Causes for Porosity and Leakage in Non-Ferrous Castings," Transactions, American Foundrymen's Association, vol. 49, pp. 331-360 (1941).

7. W. H. Johnson and W. O. Baker, "Gating Systems for Metal Castings," Transactions, American Foundrymen's Society, vol. 56, pp. 389-397 (1948).

8. W. H. Johnson, W. O. Baker and W. S. Pellini, "Principles of Gating Design, Factors Influencing Molten Metal Flow from Finger Gating Systems," Transactions, American Foundrymen's Society, vol. 58, pp. 661-668 (1950).

9. B. N. Ames and N. A. Kahn, "Effect of Superheating and Casting Temperatures on Physical Properties and Solidification Characteristics of Tin Bronzes," Transactions, American Foundrymen's Society, vol. 58, pp. 229-260 (1950).

 H. F. Taylor and W. C. Wick, "Insulating Pads and Riser Sleeves for Bronze Castings," Transactions, American Foundrymen's Association, vol. 54, pp. 262-282 (1946).

11. E. D. Boyle, "Promoting Riser Fluidity," AMERICAN FOUND-RYMAN, vol. 13, no. 2, p. 45, February 1949.

12. B. N. Ames, S. B. Donner and N. A. Kahn, "Plastic Bonded Shell Molds," *The Foundry*, vol. 78, no. 8, p. 92, August 1950.

DISCUSSION

- Chairman: B. N. Ames, New York Naval Ship Yard, Brooklyn.
 Co-Chairman: J. W. BOLTON, The Lunkenheimer Co., Cincipacti
- P. C. BUTLER: ¹ Has the Navy considered using molds at temperatures up to 1200 to 1400 F with plaster and silica as molding agent such as we use in precision casting?

MR. FREAR: How heavy a casting can you pour in a plaster mold heated to that temperature?

Mr. Butler: I have seen some that weighed several hundred pounds.

Mr. FREAR: In that case, we should consider it. I might mention that Navy castings run from a fraction of an ounce to a hundred tons.

G. P. HALLIWELL: ² Have you investigated the material that Battelle Memorial Institute has been using in their clay molds which baked at 1600 F for a number of hours?

² Director of Research, H. Kramer & Co., Chicago.

¹ Chief Metallurgist, Arwood Precision Casting Corp., Brooklyn.

Mr. Frear: We have not.

Mr. HALLIWELL: According to Battelle there is no moisture. Have you used the fracture test that is being worked on under the sponsorship of the A.F.S. Brass and Bronze Division Research

Committee at the University of Michigan?

MR. FREAR: The work being done at the University of Michigan through this Committee's sponsorship is a wonderful job. I believe the fracture test to be one of the best methods for determining the quality of metal while it is still in the furnace. I am inclined to believe that not many of us have seen fractures of bronze castings which are free from gas. I have seen but few myself. We should have some standards which will aid the foundryman in determining with surety whether he has gas-free metal, or the approximate amount of gas present.

J. W. Bolton: The author infers that if a steel casting is radiographically sound, it is a good and acceptable casting. Carrying that inference further, if you had one radiographed and acceptable according to that standard, it would be a very good casting. We in the steel casting business know that, while radiography is a good and useful tool it tells only a small part of the story. Taking your bronze casting, if radiography is over-

emphasized, a great dis-service might be done.

MR. FREAR: We have to consider the sensitivity limits of our equipment. X-rays (or gamma rays) which will penetrate 1 in. of bronze, will not show a discontinuity smaller than approximately 1 per cent of the wall thickness. Actually we are fortunate if we get 1 per cent sensitivity in the 1-in. wall. This means that if the defect is smaller than 0.01 in. we will not see it. As individual discontinuities forming interdendritic shrinkage are smaller than that, we will not see them as separate defects. They will appear as an overall darkening of the film extending as far as the porosity extends. We are not sure of our interpretation yet.

MR. BOLTON: You seem to beg the issue on the small internal cracks not discerned in radiography. Certain conditions of centerline shrinkage are there. The castings are unsound and there is a plane of weakness there. How are you going to get around

that?

MR. FREAR: A large number of defects in bronze are large enough to see by radiography. There is no difficulty in interpreting these as definite discontinuities. Even in steel castings we get defects fine enough that we cannot see them on any radiograph. Radiography is not the entire answer to inspection but has to be tied in with other methods. As far as I am concerned, the Navy will try out every method of non-destructive inspection

^a Director of Research, The Lunkenheimer Co., Cincinnati.

that comes to our notice. Every one which we find satisfactory, and useful, we will probably require in our testing where we need the highest quality of casting.

S. W. CHAPPELL: 4 What program has the Navy undertaken to substitute aluminum bronzes for manganese bronzes?

MR. FREAR: The Navy, at least the Bureau of Ships, has instituted no program to substitute aluminum bronzes in place of manganese bronzes. While it does not bear on this question, I want to repeat that we will do anything practicable to cut down on the use of tin, keeping in mind, however, the necessary resistance to corrosion in moving salt water.

Although there is no program for substitution of aluminum bronzes for manganese bronzes, there is a tendency and a desire to go to higher strength materials, such as using aluminummanganese bronze for the present manganese bronze.

manganese bronze for the present manganese bronze.

MR. CHAPPELL: The reason I asked is because that has been done in most outboard structural bronze. I heard that during an overhaul some cracked castings were found. They knew not the source of the castings. I do not know whether they determined whether the castings were cracked originally or whether they cracked under stress. We ourselves had not found such failures in manganese bronze.

MR. FREAR: I believe the failures to which you refer may be those in aluminum-manganese bronze where the zinc content is on the high side, with the result that an excessive amount of beta solid solution is formed. This excessive beta tends to cause stress corrosion failures. It has been found necessary to keep a careful balance between the aluminum and zinc content in order that the beta constituent will be kept below certain maximum limits.

MR. CHAPPELL: I question whether or not the cracks in these castings were not shrinkage cracks originally in the castings.

J. J. MAYER: ⁵ Aluminum bronze withstands alkaline action. Aluminum dissolves out in salt water whereas manganese bronze withstands salt water corrosion. Perhaps there could be a substitution. We are replacing many tin bronzes with manganese bronze which is a solid solution and we do not have this interdendritic condition to deal with.

MR. FREAR: There are many plans where substitutions can be made. We tried aluminum bronzes, silicon bronzes, in fact just about every copper-base alloy that exists. As far as I have been able to determine, nothing resists corrosion from sea water, moving at high speed, as does a copper-tin alloy with at least 5 per cent tin; hence the continuing use of gun metal and similar alloys.

g

th co te 90 te F

pe

th to In to tw

bui

⁴ Supt. of Foundries, Electric Boat Co., Groton, Conn.

Vice President, Lumen Bearing Co., Buffalo.

STRUCTURE AND MECHANICAL PROPERTIES OF A MO - NI - CR CAST IRON

By Edward A. Loria*

ABSTRACT

Information is given on the microstructure and mechanical properties of 1.2-in. diameter bars of a gray iron containing 2.77 per cent total carbon, 1.40 per cent silicon, 1.40 per cent molybdenum, 1.00 per cent nickel, 0.50 per cent chromium ascast and after tempering at 400 F, 600 F and 700 F. The matrix is a nixed structure consisting mainly of acicular ferrite and a dark-etching network pattern of low temperature transformation products (lower bainite and possibly martensite) arising from dendritic segregation of alloying elements. An ascast tensile strength of 72,500 psi with a transverse strength of 5600 lb and accompanying deflection of 0.48 in. was obtained. The tensile strength was raised to 98,300 psi upon tempering. The mechanical test data show that this acicular iron has an as-cast and reheat strength that is comparable with the best of the acicular irons now in use.

Introduction

CAST IRONS WITH AN ACICULAR STRUCTURE in the matrix and, of course, fine randomly distributed graphite, have the highest strength of any irons commercially available at the present. Molybdenumnickel acicular irons are the most common type, the alloy content being adjusted in relation to the cooling rate or size of the casting so that the austenite transforms to an acicular product at about 900 to 500 F rather than forming pearlite at higher temperatures or martensite at lower temperatures. Flinn and Reese,1 working with nickel-molybdenum alloyed iron with 2.50 per cent total carbon and 2.50 per cent silicon, have shown that high mechanical properties may be obtained if the matrix structure of the iron is acicular. They also pointed out that the combined carbon in the matrix must be eutectoid composition to form a truly acicular structure. In order to produce this eutectoid composition (0.70 to 0.90 per cent carbon range), a proper balance between the total carbon and silicon must be maintained.

The lower strength irons with tensile strengths from about 45,000 to 55,000 psi are being widely

produced today. Most of these irons are made with a relatively low carbon content (between 2.7 and 3.2 per cent), which is low enough for good strength but not so low as to cause casting difficulties. The silicon content is usually between 1.25 and 2.75 per cent (with the higher amounts corresponding to the lower carbon contents and smaller sections), which is sufficient to prevent chilled spots without an excessive reduction in the combined carbon.

Inoculation is generally employed to insure a random distribution of the graphite flakes, while suitable alloys may be added to give a fine lamellar pearlite, with little if any free ferrite or cementite, and to increase the strength to the required amount. Notwithstanding, mold cooling is so slow that it is practically impossible to form an acicular structure in unalloyed gray iron. The addition of molybdenum and nickel delay the initiation of the pearlite reaction on cooling and allow the retention of the austenite down to the 500 to 900 F range where, if the mold cooling rate is satisfactory, the acicular structure will be formed².

In order to devise a composition for high strength iron it is obvious that a low carbon equivalent is required. For the sake of castability and minimum shrinkage it is desirable to choose the higher carbon and lower silicon. Of the alloying elements, molybdenum and chromium appear to be the most potent strengtheners. Much work has been done on the alloying of cast iron but the results have not been well systemized so that the foundryman who uses alloys does so as a result of his own experience. The effect of a given quantity of a given alloy is not yet too well known. In many cases, alloys are mixed so that, for example, the chill promoting tendency of chromium is counteracted by a chill reducing tendency of nickel and the two alloys, combined, affect the strength of the resultant matrix. Table 1 lists data available in the literature on the chemical composition and mechanical properties of acicular iron as-cast comparable to those to be given in this paper.

The present work was undertaken in order to examine the structure and properties of a triple alloy iron when cast in the customary 1.2-in. diameter bars

M

0

of 1

n

n-

g

y or d

a

ıl

n

^{*} Formerly with Mellon Institute of Industrial Research, Pittsburgh, presently in the Research Dept., Carborundum Co., Niagara Falls, N. Y.

TABLE 1-CHEMICAL COMPOSITION AND MECHANICAL PROPERTIES OF SOME ACICULAR CAST IRONS

		Ana	alyses, %						A	s-Cast							
							Tr	ansverse			Tensile						
TC	CC	Si	Мо	Ni	Cr		Load, lb		efl. n.	Brinell, BHN	Strength, psi	Ref					
2.30	0.75	2.31	1.36	1.03			4700	0.	33	321	71800	(1)					
2.70	0.85	1.70	0.90	0.45	0.55		4500			260	65000	(3)					
2.49	0.66	2.71	1.03	1.01			5440	0	57		80500	(4)					
2.60	0100	2.35	1.10	1.10						300	65000	(3)					
2.65		2.48	1.15	1.21			5900	0.	19	288	72700	(3)					
2.75		2.75	0.75	0.75						269	65000	(3)					
2.75		1.50	1.00							286	56500	(3)					
2.34	0.65	2.64	0.97	2.93	0.25					375	78250	(5)					
2.34	0.65	2.64	0.97	2.93	0.25					418	71500	(5)					
2.65	0.00	2.60	1.12 1.2-in. diamete	1.12	0.15	tosts	7300 0.80-in	dismeter	tensile	305	67500	(6)					

and after a low temperature reheating as it has been shown that the as-cast strength of acicular iron is increased by tempering at 500 to 700 F for 5 hr or more¹⁻². The iron was melted in a cupola with 60 per cent steel in the mixture and the necessary alloying done in the cupola and ladle to produce the following composition:

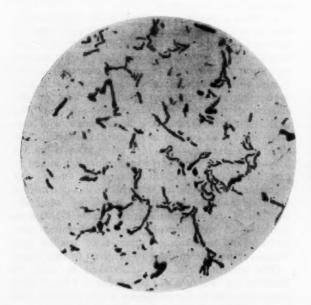
TC .	CC	GC	Si	Mo	Ni	Cr
2.77	0.70	2.07	1.40	1.40	1.00	0.50

In order to avoid reduced strength due to low pouring temperatures,⁷ the test bars were cast at 2750 to 2800 F. The transverse loading tests were made on the 1.2-in. diameter bars broken on the 18-in. centers and the tensile tests were machined therefrom to a 0.80-in. diameter, both tests being made according to ASTM specifications. Brinell hardness measurements and microstructural examination were made on the cross-sections. Tempering of the bars at 400 F, 600 F and 700 F for 5 hr was carried out in a controlled atmosphere furnace.

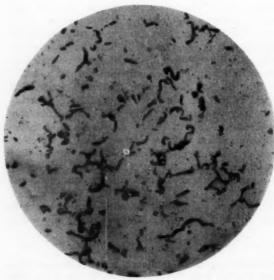
Structural Observations

Representative photomicrographs of the as-cast structure are shown in Fig. 1 to 3. It is evident in Fig. 1 that the graphite size is small and the distribution of the interdendritic type, whether it be examined in an area containing a small amount of primary carbide or in one containing a large amount. Figure 2 portrays the etched structure from the same two areas. The presence of primary carbide particles and a dendritic network pattern is apparent. These outlines of cross-sections of dendrite arms occur owing to the fact that segregation of alloying elements during freezing is much more pronounced with respect to dendrite core and interarm areas than it it with respect to centers and boundaries of entire grains**.

^{**} There is a further response of the primary dendritic structure to cooling rate. This is the effect of the direction of heat flow on the orientation of dendritic growth. The extent of directional growth depends, of course, in the intensity of the thermal gradient developed. In a round bar cast in sand the dendrites near the surface will be more or less radial in arrangement whereas toward the center the distribution becomes random.







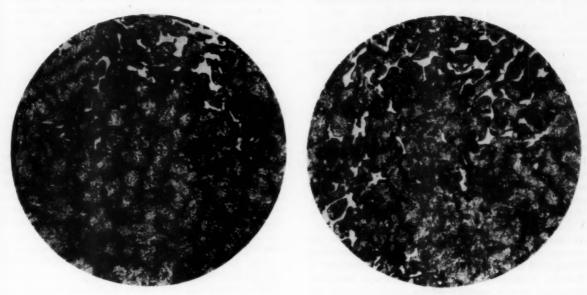
B. Area with large amount of primary carbide.

Fig. 1-Graphite structures in Mo-Ni-Cr cast iron. Unetched. x100

Figure 3 shows the matrix structure at higher magnification. The predominantly upper bainitic (acicular ferrite) structure within the network and also the finer, darker and less resolvable areas of lower temperature products (lower bainite) arising from alloy enrichment at the cross-sections of dendrite arms can be seen in each photomicrograph. In effect, there is a gradation from the truly acicular ferrite to an almost martensitic structure around the carbide particles. Eagan⁸ reported a similar mixed structure for an alloyed gray iron of different composition and believed that the "pseudo-dendritic pattern" was an indication of combined carbon segregation; that is,

areas higher in combined carbon etch darker than those of lower carbon content, thus revealing the original dendritic pattern of the metal. According to Eagan, when carbon was diffused more evenly throughout the matrix, no such pattern can be seen.

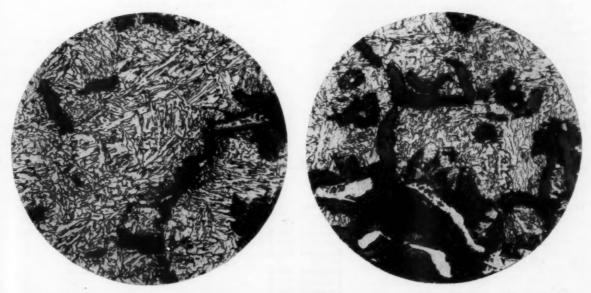
Evidence of dendritic segregation in this acicular iron is shown by the dark cell-like outlines. Segregation always occurs during the freezing of an alloy, the segregation is more severe in some cases than others. In other words, some elements are more prone to segregate than others and consequently the resulting structure in areas where segregation occurs will be different. Further, some elements are in-



A. Same area as Fig. 1A. Note dendritic network.

B. Same area as Fig. 1B.

Fig. 2-Matrix structures in Mo-Ni-Cr cast iron. Picral etch. x100



A. Area with acicular ferrite and lower bainite.

B. Large amount of primary carbide and lower bainite.

Fig. 3-Matrix structures in Mo-Ni-Cr cast iron. Picral etch. x500

herently slower to diffuse through iron than are others. Typical are elements with atoms similar in. size or larger than iron atoms such as manganese, nickel and molybdenum. Small atoms such as carbon diffuse thousands of times faster; thus, for example, plain carbon cast steels are usually less segregated by the time they cool to room temperature than are alloy cast steels. This has been illustrated in the work of Hawkes and Brown,9 Loria, 10 Timmons11 and Wilks, Avery and Cook.12

Because of the known higher hardness of lower bainite over that of the higher temperature transformation products, a microhardness survey was made as an aid in identifying the observed microstructures. A Tukon microhardness tester equipped with a Vickers pyramidal diamond indenter operating on a 25-gram load gave a range of 275 to 305 VPN (average 285) for the acicular ferrite regions and a range of 375 to 465 VPN (average 410) for the darketching, lower bainite network.

Mechanical Properties

The tensile, transverse and hardness data obtained for the triple alloy iron bars whose microstructure has just been discussed are contained in Table 2. Each value listed therein is an average of three or more tests. The data reported by Flinn¹ on an iron which is quite similar as far as carbon equivalence and alloy content (exclusive of chromium) are concerned are included therein for comparison purposes. The as-cast mechanical properties of the molybdenumnickel-chromium iron are noteworthy in that an average tensile strength of 72,500 psi with a transverse strength of 5,600 lb and accompanying deflection of 0.48 in. is obtained. Tempering produces markedly improved tensile properties, the strength being in creased 19.0 per cent by a 5-hr draw at 400 F, 35.6 per cent by a 5-hr draw at 600 F, and 35.2 per cent by a 5-hr draw at 700 F.

These data are compared to those obtained by Flinn² on a molybdenum-nickel iron with an as-cast tensile strength of 71,860 psi which was increased b 17.0 per cent, 36.5 per cent and 35.4 per cent after similar tempering treatments. The ground mass of an acicular iron always has some retained austenite which is test removed by tempering after cooling to room temperature. It is believed that the improvement in mechanical properties is a consequent result. Possibly, the tempering of any lower bainite or

martensitic structures such as were seen around the primary carbide particles may have contributed also to the observed increase. Metallographic examination of the tempered test bars revealed no significant changes in the structures depicted in Fig. 1 to 3. Only high temperature, long time homogenizing treatments would break down the carbide particles and broaden (dilute) the network. Consequently this would be impractical under the circumstances.

Summary

The microstructure and mechanical properties of 1.2 in. diameter arbitration bars of a particular gray iron containing 2.77 per cent total carbon, 1.40 per cent silicon, 1.40 per cent molybdenum, 1.00 per cent nickel, 0.50 per cent chromium as-cast and after tempering at 400 F, 600 F and 700 F have been presented and analyzed. The matrix reveals a mixed structure of mainly acicular ferrite and a dark-etching network pattern of low temperature transformation products (lower bainite and possibly martensite) arising from dendritic segregation of alloying elements. During freezing all substances dissolved in the iron tend to be rejected to the grain boundaries or dendritic fillings to form a high concentration on a microscopic scale. The rate of diffusion of most alloy constituents is too slow to permit dispersal of this segregation. Thus, the iron is made up of low alloy and high alloy portions. The low alloy parts transform rather rapidly, but as the alloy content increases, the transformation is more sluggish until the highest concentrations are still austenitic when casting is comp.ete. An as-cast tensile strength of 72,500 psi with a transverse strength of 5,600 lb and an accompanying deflection of 0.48 in. was obtained. This strength was raised from 86,300 psi to 98,000 psi upon tempering from 400 to 700 F. Probably the improvement was brought about by the removal of any retained austenite and the tempering of the lower bainite (martensite) structures.

References

1. R. A. Flinn and D. J. Reese, "The Development and Control of Engineering Gray Cast Irons," TRANSACTIONS, American

Foundrymen's Association, vol. 49, p. 559-607 (1911). 2. R. A. Flinn, M. Cchen and J. Chipman, "The Acicular Structure in Nickel-Molybdenum Cast Irons," *Transactions*, American Society for Metals, vol. 30, p. 1255-1283 (1942).

3. R. S. Archer, J. Z. Briggs and C. M. Loeb, "Molybdenum in steels, Irons and Alloys," Table 74, p. 284-303 (1948).
4. G. A. Timmons, V. A. Crosby and A. J. Herzig, "Produces

TABLE 2-MECHANICAL PROPERTIES OF ACICULAR IRON AS-CAST AND AFTER TEMPERING TREATMENT

Analysis, %							Trans	Transverse		Tensile	Inc.	
тс	CC	Si	Мо	Ni	Cr	Condition		Load,	Defl., in.	Hardness, BHN	Strength, psi	Tens.
2.77	0.70	1.40	1.40	1.00	0.50	As-Cast		5600	0.48	325	72500	
						Tempered	400 F			368	86300	19.0
						Tempered	600 F			375	98300	35.6
						Tempered	700 F			363	98000	35.2
2.30	0.75	2.31	1.36	1.03		As-Cast		4700	0.33	321	71800	
						Tempered	400 F			364	84000	17.0
						Tempered				340	98000	36.5
							700 F			364	97200	35.4
5-hr T	emper					P					-,	2012

1.2-in. bars, 18-in. span 0.80 in. tensile specimens

so

011

ni

3.

ng

es

ly

of

ay

er

nt

er

e-9

ed

h-

a.

e)

e-

in

es

n

St

of

W

ts

nt

til

en

of

nd

d.

Si

n-

19

er

n-

an

ar

m

es

High Strength Iron," *The Foundry*, vol. 66-67 (1938-1939). 5. R. A. Flinn and H. J. Chapin, "Ductility and Elasticity of White and Gray Irons," TRANSACTIONS, American Foundrymen's Association, vol. 54, p. 141-153 (1946).

6. Alloy Cast Irons Handbook, American Foundrymen's As-

sociation, p. 240-266 (1944).
7. G. A. Timmons and V. A. Crosby, "Effect of Pouring Temperature on the Strength and Microstructure of Gray Cast TRANSACTIONS, American Foundrymen's Association, vol. 49, p. 225-254 (1941).

8. T. E. Eagan, "Effect of Heat Treatment on the Endurance Limit of Alloyed Gray Cast Iron," TRANSACTIONS, American Foundrymen's Association, vol. 54, p. 230-240 (1946).

9. M. F. Hawkes and B. F. Brown, "As-Cast Structures in Cast Steels," The Iron Age, vol. 163, no. 41, p. 138-147, October (1948).

10. E. A. Loria, "Detection of As-Cast Austenite Grain Size in Heat Treated Cast Alloy Steels," *Transactions*, American Society for Metals, vol. 40, p. 677-702 (1948).

11. G. A. Timmons, "Factors Affecting the Ductility of Cast Steel," Transactions, American Foundrymen's Association, vol. 51, p. 417-480 (1944).

12. C. R. Wilks, H. S. Avery and E. Cook, "Relation of Quenching Rate and Hardenability to the Mechanical Properties of Several Heat Treated Cast Alloy Steels," Transactions, American Society for Metals, vol. 38, p. 437-470 (1947).

13. Alexander Finlayson, "Diesel Engine Crankshafts Cast in Gray Iron," Foundry, vol. 78, no. 8 and 9, Aug. and Sept.

1950

14. F. B. Rote and W. P. Wood, "Segregation of Molybdenum in Phosphorus-Bearing Alloyed Gray Cast Iron," Transactions, ASM, vol. 35, pp. 402-434 (1944). 15. W. W. Braidwood, "Theoretical Considerations and Pro-

duction Details for Acicular Cast Iron," Foundry Trade Journal, vol. 87, pp. 685-689 (1949).

DISCUSSION

Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit. Co-Chairman: R. SCHNEIDEWIND, University of Michigan, Ann Arbor, Mich.

J. S. VANICK (Written Discussion)1: The term "segregation" in connection with metals and alloys has been too frequently associated with undesirable characteristics and with the concentration of undesirable elements. In this case, cast iron is known to freeze with a eutectic carbide structure which assembles in a cell-like pattern and, as such, the structure is not regarded as "segregated." It is obvious that when this occurs in a cast iron in which the carbide component may not have time to diffuse after partially or completely breaking down, some areas in its structure will retain these residues. The microstructure in the neighborhood of these residues will be correspondingly enriched in carbon. The micrographs shown seem to bear out this feature, and Eagan's experiments which are referred to indicate that the carbon enrichment can be diffused if such diffusion is desirable or necessary.

In this connection, Eagan seems to have proven a point which the author chooses to disregard and to label the condition with his own unfortunate terminology, lending confusion to the situation and creating an impression among the inexpert that an undesirable structure (segregation) occurs in the

material. I would favor a revision of the text in which the normal crystallization of cast iron into the carbide eutectic structure is recognized, and Eagan's point concerning its diffusion emphasized unless the author has other evidence of his own proving that the "segregation" is actually a concentration of undesirable elements or an enrichment of desirable elements remaining as a residue of slow diffusion.

T. E. EAGAN (Written Discussion): 2 Gray irons with the acicular structure have been used by several companies so that we now have a considerable amount of practical information about them. The iron offered by the author should be reviewed in the light of the information now available. The available knowledge has been magnificently presented by Finlayson.18

In the first place, the author presents the results obtained in 1.2-in. arbitration bars and has not apparently given consideration to the fact that the acicular irons are all section sensitive. This could lead practical foundrymen and others who are not acquainted with this type of iron into trouble.

In the second place, any alloy addition to gray iron should be economical. Flinn and Reese 1 have shown that results as good as reported here can be obtained by 0.75-1.00 per cent molybdenum with varying quantities of nickel depending on the section thickness of the casting. The author presents metal with 1.36 per cent molybdenum in one case and 1.40 per cent molybdenum plus 0.50 chromium in another which from an economical standpoint makes the results obtained of academic interest. Furthermore, Brinell hardnesses of 363 to 375 are almost out of the practical machinable range especially if there are any hard particles in the matrix. From a practical standpoint we have found that 325 Brinell in the acicular irons is about the upper economical machinability limit.

The use of chromium in acicular gray irons is not recommended. We have found in our experimental work that chromium over 0.20 per cent will produce a pearlitic structure in crankshafts, but that the 1.2-in. arbitration would be acicular.

The identification of constituents in acicular gray irons is semetimes difficult, therefore, the use of the word dendritic segregation is sometimes misleading. The author seems to believe that the peculiar dendritic pattern shown is due to alloy segregation while this discussion, using an entirely different analysis of iron, called it combined carbon segregation. Mr. Loria seemed to believe that the carbon would be able to move through the space lattice without hindrance, but that the alloy would have difficulty doing so. This may be true; however, it is also true that the transformations in gray iron are very sluggish, especially when large quantities of alloys are involved. Hence, it is possible to trap the carbon before equilibrium conditions are attained; this would give the pseudo-dendritic structure shown. This was true in the case of the irons examined by the discussor because further holding in the austenitic range caused these pseudo-dendritic forms to disappear. The irons in question were pearlitic and not acicular.

Rote and Wood14 have shown that the phosphorus in high

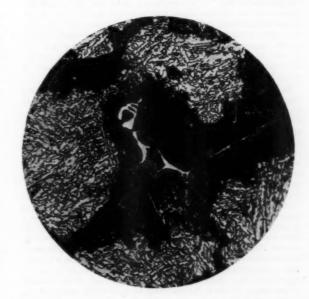


Fig. 4-Photomicrograph showing effect of molybdenum-phosphorus eutectic on the structure of a highnickel-molybdenum iron having the following analysis: T.C. 2.80; G.C. 2.21; C.C. 0.69; Si, 1.47; Mn, 0.97; P, 0.059; S, 0.074; Ni, 1.94; Cr, 0.08; Mo, 1.11. Etch-Picral. Mag.-500x.

¹ Research and Development Div., the International Nickel Co., Inc.,

² Chief Metallurgist, Cooper-Bessemer Corp., Grove City. Pa.

Fig. 5-Steadite distribution and structure in gray cast iron. A-Area showing distribution. Picral etch. Mag. -100x. B-Area showing steadite structure. Picral etch. Mag.-500x. C-Area where eutectic envelopes entire cells. Picral etch. Mag.-500x.

molybdenum containing gray iron will form a complex molybdenum phosphide eutectic, which varies in composition, but in general had a composition as follows: C, 2.10; Si, 1.10; Ni, 1.65; P. 4.40; Mo, 5.85. Now the phosphorus being the last to freeze is always segregated in the original dendritic network and where it robs the surrounding areas of molybdenum, which then transforms at a different temperature forming what the author calls lower bainite. Rote and Wood claimed that they did not find this eutectic in their tests where the phosphorus was below 0.12 per cent. However, the discussor has found it to occur with phosphorus contents as low as 0.08 per cent. It all depends on how fast the test bar or casting cools. Figure 4 illustrates the type of thing we are considering. The white constituent in this case has been identified as the phosphide-molybdenum eutectic. It is recommended that this structure be compared with Fig. 3B in the author's paper. It would be interesting to know the amount of phosphorus present in the material reported by him.

F. B. ROTE: ³ I would like to comment on the identification of the microconstituents presented by Dr. Loria. He identified the light-etching constituent as upper bainite, and the dark etching constituent as lower bainite. The former identification agrees with the published literature; however, I feel that the dark etching material is pearlite, rather than lower bainite. Similar structures were shown in a paper which I presented with Prof. W. P. Wood ³⁴ several years ago, and by others.

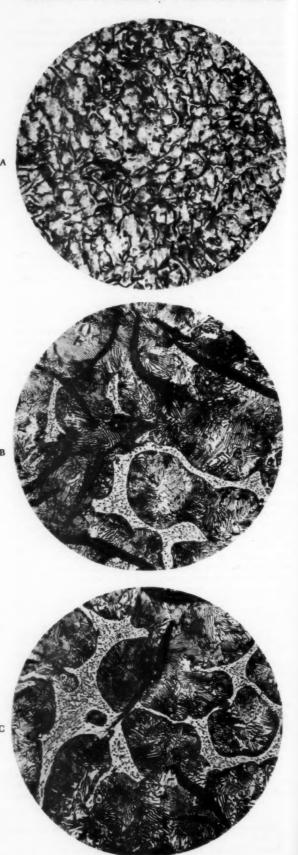
Hardness tests can not be used with certainty in the identification of complex structures, particularly those which contain bainite, since the hardness of upper bainite might be less than, and the hardness of lower bainite greater than the hardness of pearlite. Thus, complete reliance on a hardness test to identify complex structures may lead to wrong conclusions.

E. A. LORIA (Author's Closure): The comments of Messrs. Vanick and Eagan are appreciated. First of all I would like to say that I am not promoting the use of this particular composition of iron. It just happened to be one that I had the opportunity to examine about five years ago and in view of the dearth of published information on acicular irons I thought that the results would be of general interest. The iron composition, as Mr. Eagan surmises, was not designed for crankshafts as can be seen from a comparison of the analyses, particularly chromium content, and the mechanical properties for such irons as given in the Finlayson article quoted by him.

In reply to Mr. Vanick, the normal crystallization of cast iron into a eutectic structure and Eagan's point of combined carbon segregation are recognized and not discounted in the paper. Proof of dendritic segregation is shown metallographically by the appearance of low temperature transformation products in the cell boundaries and the greater microhardness of these structures in relation to the acicular ferrite in the center of the celllike outlines. In pure metals, where the atoms are all the same, the last portions of liquid to freeze in the spaces between dendrite arms are exactly like the initial nucleus and the primary dendrite axes. On the other hand, in this iron, there are many different kinds of atoms and they do not freeze out in the same ratio as they exist in the liquid metal. The nuclei have a greater ratio of iron atoms than the liquid, but as the dendrites grow the proportion of atoms other than iron freezing out increases. Finally, the last metal to freeze (which fills in between the branches of the dendrite skeletons) has dissolved in it a fairly high concentration of atoms other than iron. Thus it can be appreciated that there is a difference in the type of distribution of structures to which the austenite in these areas transforms on cooling to room temperature. The low alloy parts transform rapidly, but as the alloy content increases, the transformation is more sluggish until the highest concentrations are still austenitic when casting is complete.

cooling to room temperature. The low alloy parts transform rapidly, but as the alloy content increases, the transformation is more sluggish until the highest concentrations are still austenitic when casting is complete.

*Formerly with University of Michigan, now with Albior Malleable Iron Co., Albion, Mich.



Mr. Eagan brings out the fact that the mechanical test results were obtained on 1.2-in. diam bars and that acicular irons are all section sensitive. This is correct which means that if the composition were designed for a 6 to 8-in. section it would be overalloyed for a 1 in. section with transformation taking place at a lower than optimum temperature with consequent greater hardness and strength. Thus, for example, with molybdenum maintained in the range 0.8-1.0 per cent, nickel can vary from 1.0 to 4.5 per cent in casting from 0.5-in. to 10-in. sections while chromium will be used up to 0.25 per cent for sections from 3 to 6 in. and up to 0.50 per cent for sections 6 in. and over. At the same time, silicon will be lowered from 2.0 to 1.6 for sections 3 in, and over. These figures are confirmed by the work of Braidwood.15 Probably no hard and fast rules can be made on the basis of section thickness. As is being done for pearlitic irons, cutting up a casting and comparing structure and mechanical properties with the arbitration bar is the best way to determine the relationship.

Mr. Eagan points to the early work of Flinn and Reese and states that results as good as shown in Table 2 have been reported. That may be true. Actually, for comparison purposes, the second iron listed in Table 2 was taken from the Flinn and Reese paper, being the one iron that was closest in analysis other than chromium) to the one being considered and for which there were mechanical property data, both as-cast and after tempering. Of course, it should be realized that the Brinell hardness values reported in Table 2 are for the 1.2-in. diam bar cross-section. This iron, when poured into a 6-in. section, would have a Brinell hardness ranging from 247 to 277 which would be quite machinable. Overalloying in the test bar section and tempering produced the high values reported.

Mr. Eagan states that chromium additions over 0.20 per cent will produce pearlite in crankshafts while the 1-2-in. arbitration bar will have an acicular structure. This may be analogous to the results published for chromium steels and straight carbon hyper-eutectoid steels which show that an increase in either of these elements accelerates the pearlite reaction. However, chromium is being used in the amounts specified previously for heavy section castings such as hot-working dies and camshafts. It will increase the hardness, strength and wear resistance of heavy sections to the same extent as molybdenum.

Finally, Mr. Eagan raises an important question as to the nature of the white constituent associated with the dark-etching structure shown in Figs. 2 and 3. He may be correct in identifying the particles as the phosphide-molybdenum eutectic. The phosphorus content of the iron was 0.12 per cent and no doubt a certain amount of the white constituent is the phosphide-molybdenum eutectic. I considered the particles as mostly carbide, their formation being enhanced by the presence of chromium in the iron, the fast cooling rate in the test bar and owing to the fact that the author did not observe the dot-like particles usually found in the steadite phase of ordinary low phosphorus, pearlitic irons. This is vividly shown in Fig. 5 taken from the centers of 1.2-in. arbitration bars of an iron whose analysis and mechanical properties were:

TC	CC	GC	Si	Mn	P	S
3.41	0.48	2.93	1.51	0.54	0.42	0.16
Transver	rse Load 325 lb	Deflect 0.23		Tensile St 32250		Brinell 207

EFFECTS OF CERTAIN ELEMENTS ON GRAIN SIZE OF CAST COPPER-BASE ALLOYS

By

R. A. Colton and M. Margolis*

Introduction

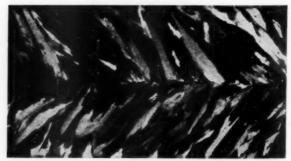
IMPROVED PERFORMANCE of copper-base casting alloys can come about only with better understanding of the factors controlling physical properties, casting behavior, machinability and other characteristics of engineering importance. Among the least understood of all these factors is the role of grain size of cast brass and bronze in determining behavior of the alloys. Miscellaneous observations on grain size have been reported occasionally but correlated data on the relationship of grain size and properties are not available.

Lack of useful information on the effects of grain size may be in part due to the difficulty of making grain size of cast alloys an independent variable in experimental work. The cast grain size of brasses and bronzes depends largely on the cooling rate, the cooling direction, and the amount of turbulence during freezing. It can be further complicated by other effects to be discussed later. Consequently, it is often difficult to decide which of the various factors involved is primarily responsible for measured variations in physical properties.

It has been reported, more from observation than from investigation, that fine grain size of cast copperbase alloys is desirable because of improved tensile properties. There is precedent for this belief from the knowledge available of wrought ferrous and non ferrous materials where demonstrable improvements in tensile properties accompany grain refinement. However, the available evidence that a parallel condition exists in cast alloys is decidedly sketchy. Moreover, in attempting to follow the thesis that fine grain size per se is responsible for optimum properties, one is confronted by exceptions such as that illustrated in Fig. 1. This photograph represents a macroetched section of 88-10-0-2 as produced by continuous casting. Large radial columnar grains are normally obtained by this method of casting; at the same time, the

tensile properties are usually equal to or better than those obtained by casting this alloy in any other fashion. Thus it is obvious that grain size differences can at best be only partially responsible for the changes of properties that occur when the casting conditions are varied.

There is good reason to believe that the usual observation about low tensile properties being associated with coarse grain size is due to lack of understanding of other factors operative during the solidification process. Grain size of cast materials is almost always a function of cooling rate. When the grains are coarse, a slow cooling rate is indicated and this may be the result of a high pouring temperature. Use of high pouring temperatures can lead to greater-thannormal shrinkage in castings, or to excessive quantities of dissolved gases in the metal, and it is known that both shrinkage porosity and gas porosity can seriously impair tensile properties of cast copper-base alloys. That coarse grain size is observed when low physical properties are obtained may be only a coincidence, or at best, an indication of the fact that the metal had been poured at a high temperature.



Nominal Composition
Cu 88, Sn 10, Pb 0, Zn 2

Tensile Properties
Tensile Strength 51,500 psi
Yield Strength 33,700 psi
Elongation 12.0%
Impact 25.5 ft-lb

Fig. 1—Continuous cast 80-10-0-2 bronze. Macroetch shows coarse columnar grains.

^{*}American Smelting & Refining Co., Research Department, Barber, N. J.

d

g

There are adequate explanations of low properties in such cases by considering the effects of shrinkage and gas porosity; coarse grain size may be only a secondary effect, unrelated to the observed properties. Fine, equiaxed grains are frequently observed in porous areas of castings; tensile properties are always low in these cases.

The lack of clear-cut evidence of the effect of grain size on properties is directly related to the lack of data on the factors affecting grain size. The influence of cooling rate is understood, and it is true that this may be the dominant factor in controlling grain size. Other possibilities exist, however, for determining grain size of cast alloys and these should be explored to increase understanding of the problem. One important aspect, the influence of composition has been studied by Northcott1 who measured the effects of small additions of a large number of second elements on the columnar grain length of copper solidified under controlled conditions. He found that marked changes in columnar grain length of cast copper are found with additions of small amounts of some elements, such as lead, thallium, tin, and platinum, while other elements, in equal amounts, have negligible influence on columnar grain length. Among such addition elements were bismuth, zinc, aluminum and chrom:um.

Of greater interest, perhaps, to the foundry industry are the factors operative in affecting grain size of the alloys used by foundries. The principal copper-base alloy in use today is 85-5-5 containing nominally 5.0 per cent tin, 5.0 per cent lead, 5.0 per cent zinc, balance copper. In studying effect of composition on grain size, it appears practical to use an alloy of this type. Such an investigation should cover effects of composition, both major constituents and impurities, on grain size. The normal impurities found in commercial alloys of this type are iron, nickel, antimony, phosphorus and sulphur. It was decided to study first the effects of nickel on the grain size of a pure metal 85-5-5-5 alloy. It has been reported by Vanick2 that nickel is a grain refiner of cast 85-5-5; other desirable effects such as reducing shrinkage, improving lead distribution, and improving "castability" have also been attributed to nickel in 85-5-5.5. To obtain systematic data on the effects of nickel in 85-5-5-5, and to study effects of composition in general, on grain size, this investigation was undertaken.

Experimental Procedure

In designing experiments to study effects of composition on grain size it is necessary to eliminate cooling rate as a variable. Experimental conditions must be such that the only effects observed are those due to composition. Furthermore, to determine the effect of nickel on grain size it is essential that the effects of tin, lead, and zinc on grain size be understood, with and without nickel present. To fulfill these conditions a group of experiments were conducted as described in the following.

Eliminating cooling rate as a variable in experiments with cast alloys requires that the materials to be stud'ed or compared have cooling rates that are exactly alike. This condition makes pouring of castings impractical because of the difficulty of keeping all experimental conditions constant. Instead, the melts involved must be cooled simultaneously, at the same rate. To achieve this, melting was carried out in graphite crucibles 1 in. I.D., 8 in. long with 3/32 in. wall thickness, using a small Ajax-Northrup high frequency induction furnace. With the lining diameter of the Ajax furnace approximately 6 in. it was possible to group nine of the small graphite crucibles around the periphery, standing on a circular block of graphite. Two more such crucibles could be placed in the center of the furnace so that if desired, 11 separate melts could be made simultaneously, brought to the same temperature, and when the power was turned off, allowed to cool simultaneously. For all practical purposes, the cooling rate of the 11 melts in the furnace could be considered identical. A cover was usually placed on the furnace to reduce the heat lost by radiation from the top, and also, to slow down the rate of cooling. With this arrangement, solidification apparently starts at the bottom of the crucible and proceeds toward the top for at least half of the freezing interval.

On three separate occasions cooling curves were made for melts solidifying in the furnace using a platinum-platinum+10% rhodium thermocouple with a Leeds & Northrup potentiometer. In each case the cooling rate was found to be approximately 35 F per minute. A more complete discussion of this phase of the work is made later in this report. From these determinations it is fair to assume that the cooling rate of the furnace itself is known, and that it is reasonably constant, so that comparisons can be made between melts made separately as well as among melts made at the same time.

Separate experiments were run to determine the effect of nickel on the grain size of copper plus 5 per cent lead, copper plus 5 per cent tin, copper plus 5 per cent zinc, and 85-5-5-5 alloy made with pure metals. By studying each of the major alloying constituents individually, it was hoped to determine the magnitude of effects exerted on grain size by each element of composition of the alloy.

Materials used consisted of the following: electrolytic copper, high purity tin, lead and zinc, and high purity, cobalt-free nickel shot.

The "castings" obtained in these experiments were cylindrical billets 1 in. in diam, 5 in. long, weighing about 475 grams. Each casting was faced off on a lathe at one end, with the other end (bottom) cut off with a parting tool. The transverse section obtained was polished, etched and observed at low magnification. The remaining part of the casting was faced off in a lathe so that a longitudinal section was available. The machine chips were collected and used for chemical analysis. The longitudinal section

¹L. Northcott, Journal Inst. of Metals, vol. 62, p. 101 (1938).

²J. S. Vanick, The Foundry, October 1950, vol. 78, p. 88.

TABLE 1-CHEMICAL COMPOSITIONS AND GRAIN COUNT DATA

							Grain	Cou	nt
					1	Distan			ton
Bar		Compo	sition,	%		of bar, in.			
No.	Cu	Sn	Pb	Zn	Ni	1/2	1	11/2	2
L-1	99.9+					6	5		
L-2	Rem		. 5.77			19	22	18	20
L-3	Rem		5.84		0.08	22	22	20	23
L-4	Rem		5.38		0.12	21	20	22	21
L-5	Rem		5.57		0.24	15	16	17	16
L-6	Rem		5.45		0.48	17	19	18	16
L-7	Rem		5.01		0.71	17	18	17	17
L-8	Rem		5.47		0.87	17	17	15	18
L-9	Rem		5.38		1.82	20	17	16	19
T-1	99.9+					3	3	2	4
T-2	Rem		4.7			11	12	6	11
T-3	Rem		4.6		0.05	14	18	13	7
T-4	Rem		4.7		0.10	14	13	15	9
T-5	Rem		4.8		0.17	12	11	11	9
T-6	Rem		4.6		0.44	14	12	16	15
T-7	Rem		4.6		0.68	12	16	16	12
T-8	Rem		4.6		0.99	9	10	8	4
T-9	Rem		4.7		1.80	16	12	15	10
Z-1	99.9+		***			3	3	2	2
Z-2	Rem			5.8		9	3	3	
2-3	Rem			5.8	0.06	10	5	6	6
Z-4	Rem			5.8	0.14	3	3	1	
2-5	Rem			5.8	0.24	4	3	4	3
Z-6	Rem			5.8	0.49	11	13	6	5
2-7	Rem			5.8	0.73	6	7	4	
Z-8	Rem			5.8	0.94	4	6	7	3
Z-9	Rem			5.8	1.92	7	4		
N-1*	84.13	4.69	4.77	5.52 (d)		20	21	18	
N-2	83.0	5.2	5.3	6.2	< 0.01	21	20	21	18
V-3	83.0	5.2	5.3	6.2	0.05	22	22	19	20
N-4	83.0	5.2	5.3	6.2	0.09	27	23	23	17
N-5	83.0	5.2	5.3	6.2	0.25	18	18	17	17
N-6	83.0	5.2	5.3	6.2	0.45	18	15	17	14
N-7	83.0	5.2	5.3	6.2	0.79	20	15	17	14
V-8	83.0	5.2	5.3	6.2	0.79	19	18	18	14
							14	12	13
V-9	93.0 .14, Sb 0.5	5.2	5.3	6.2	1.98	16	1.4	14	13

was polished to remove tool marks, macroteched, and studied for grain size. Available then, were complete longitudinal sections of each casting as well as transverse specimens taken from each sample at the same position.

Samples were polished through 3/0 paper, and etched by immersion in saturated ammonium persulfate solution to reveal grain contrast.

Each experimental run will be described separately along with observations made.

Copper + 5% Lead

For this experiment nine graphite crucibles were used around the periphery of the furnace, and two crucibles in the center. The center crucibles were present principally to prevent excessive heat loss by radiation on the center side of the crucibles placed around the wall. Without crucibles in the center of the furnace higher rate of heat loss toward the axis of the furnace could be expected. Melts made in the center crucibles were usually of similar alloys but not necessarily materials used as part of the experiment.

Of the nine melts made simultaneously, the nomi-

nal compositions were as follows:

L-1	Copper						
L-2	Copper	+	5%	Lead			
L-3	Copper	+	5%	Lead	+	.05%	Nickel
L-4	Copper	+	5%	Lead	+	.10%	Nickel
L-5	Copper	+	5%	Lead	+	.25%	Nickel
L-6	Copper						
L-7	Copper	+	5%	Lead	+	.75%	Nickel
L-8	Copper	+	5%	Lead	+	1.00%	Nickel
L-9	Copper						

The procedure used was to melt down copper first, when molten, lead was added to all but one crucible, and after 10 min alloying time, nickel additions as reported were made and another 10 min allowed for alloying. Good mixing is obtained in high frequency electric furnaces. When the temperature was approximately 2150 F, the furnace was turned off and solidification allowed to progress in the furnace, with a cover in place.

The "castings" so produced were marked and prepared as previously described. The chemical analysis of each of the castings is reported in Table 1. The polished and etched samples were photographed and are shown in Fig. 2.

It is desirable to have some method of comparing grain size other than visual observation. Numerous techniques have been used to measure size of grains in various materials but after consideration of the specific problems involved in this work, the intercept method was adopted. The procedure involved photographing the polished and etched longitudinal section of each specimen at $1\frac{1}{2}$ x magnification, drawing parallel lines on the prints at intervals of $\frac{1}{2}$ in., 1 in., $\frac{1}{2}$ in. and 2 in. from the bottom of the casting, and counting the number of grains cut by the intercept line. The data from these measurements are listed in Table 1. Measurements were not converted to grain diameter because of no apparent advantage.

Examination of the etched castings of Fig. 2 and the data in Table 1, indicates that 5 per cent lead has a strong refining effect on the grain of copper. The differences between samples L-1 and L-2 are easily seen. The effect of nickel on the grain size of copper containing 5 per cent lead is much smaller, and relatively speaking, almost neglible. Some grain refinement can be found in Sample L-3 with 0.08 per cent nickel and L-4 with 0.12 per cent nickel, but all other castings apparently are of comparable grain size. Whatever effect nickel may have appears with small additions and disappears with larger nickel contents.

The rather unusual combination of grain sizes found in the castings of Fig. 2, can be explained by comparison with structures found in copper wire bars, brass ingot, etc. Apparently the major part of the heat extraction takes place from the bottom and sides of the crucible, creating small equiaxed grains near the bottom of the casting. Some chilling must occur at the top of the casting, while the coarse grains must be the result of a few nuclei forming and growing slowly because of heat extracted about uniformly from all of the upper portion of the casting. For purposes of these experiments only the bottom half of the castings was used for grain size comparisons;

d

Cu Pb Ni

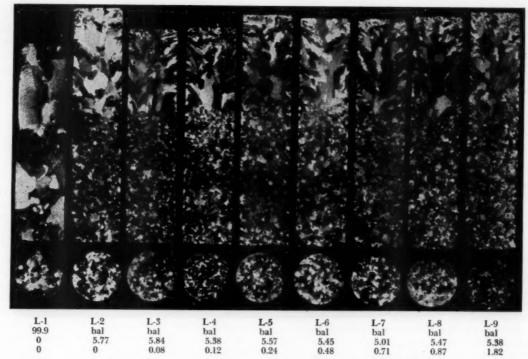


Fig. 2-Photo of macroetched sections of copper plus 5 Pb series.

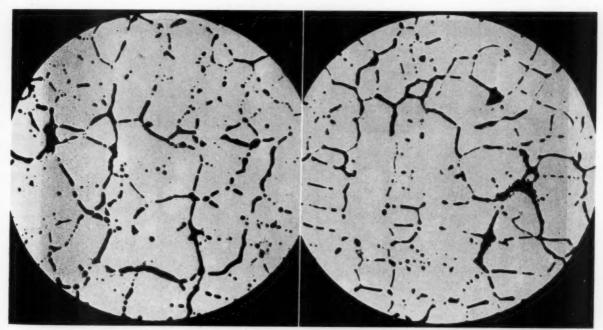


Fig. 3-Photomicrographs showing effect of 2% Ni on lead distribution of 5% Pb-Cu alloy. Unetched. Mag. 50×.

since castings produced under similar conditions are compared, this procedure seems reasonable.

No internal or external shrinkage was noted in the copper-lead or the copper-lead-nickel castings. Some piping was found as expected in the copper casting. From all indications the presence of nickel had no

effect whatsoever on shrinkage of the alloys.

It has frequently been reported in foundry discussions that nickel is beneficial with regard to distribution of lead in copper alloys. To test this theory metallographic examination was made of two copperlead alloys, one without nickel, the other containing Cu Sn Ni

Cu Zn Ni

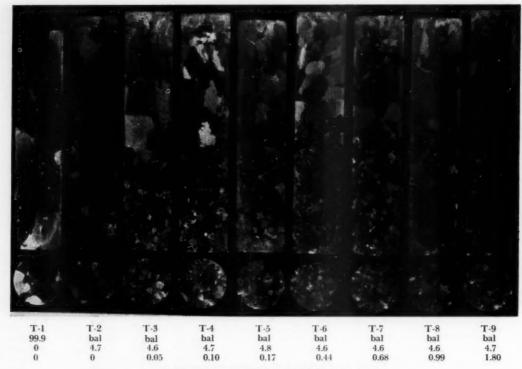


Fig. 4-Macrographs of Cu-Sn series.

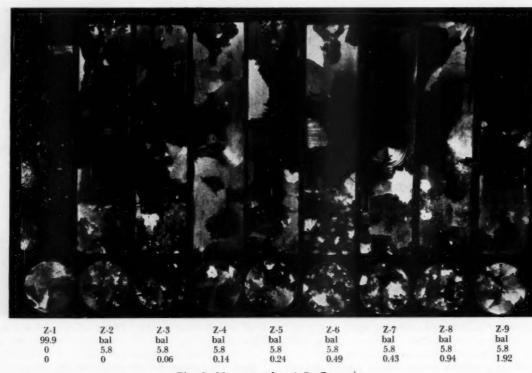


Fig. 5-Macrographs of Cu-Zn series.

2 per cent nickel. Photographs of the polished and unetched castings are shown in Fig. 3. No differences were found in either microstructure or lead distribution. It is possible, of course, that with other cooling conditions nickel might influence distribution of lead; no such effect was found in this work.

Copper + 5% Tin

Following procedures previously described, a determination of the effect of nickel in copper plus 5 per cent tin alloys was made, using alloys of the following compositions:

T-1 Copper
T-2 Copper + 5% tin
T-3 Copper + 5% tin + 0.05% nickel
T-4 Copper + 5% tin + 0.10% nickel
T-5 Copper + 5% tin + 0.25% nickel
T-6 Copper + 5% tin + 0.50% nickel
T-7 Copper + 5% tin + 0.75% nickel
T-8 Copper + 5% tin + 1.00% nickel
T-9 Copper + 5% tin + 2.00% nickel

Polished and etched samples are shown in Fig. 4; chemical analyses and grain count data are listed in Table 1. Study of the results indicates that 5 per cent tin refines the grain size of copper somewhat, and that further refinement is brought about by the presence of nickel. Once again, the first increment of nickel, 0.05 per cent, produced finer grains than in 5 per cent tin alloy alone, but further additions of nickel up to 2 per cent do not seem to exert any appreciable influence on grain size. The coarser grains of sample T-8 probably have no significance other than representing an unfavorable axial location longitud nally. The same cut through a different diameter probably would have been in line with the other castings.

Slight internal porosity (probably interdendritic) was found in all the 5 per cent tin alloys in an area near the bottom of the samples. The nickel content seems to have negligible influence on this porosity. Severe piping occurred in the copper casting.

Copper + 5% Zinc

In studying the copper plus 5 per cent zinc series it was necessary to modify the procedure somewhat to insure having uniform compositions of all 5 per cent zinc alloys. It is quite difficult to add zinc to small crucibles uniformly. For this experiment a master alloy of 95 per cent copper, 5 per cent zinc, was made up in larger quantities from high grade materials and cast into pencil molds to furnish remelting stock. The master alloy was then directly remelted in 475-gram lots and nickel additions made to each crucible as before.

The following melts were made:

Z-1 Copper
Z-2 Copper + 5% zine
Z-3 Copper + 5% zinc + 0.05% nickel
Z-4 Copper + 5% zinc + 0.10% nickel
Z-5 Copper + 5% zinc + 0.25% nickel
Z-6 Copper + 5% zinc + 0.25% nickel
Z-7 Copper + 5% zinc + 0.75% nickel
Z-8 Copper + 5% zinc + 1.00% nickel
Z-9 Copper + 5% zinc + 2.00% nickel
Z-9 Copper + 5% zinc + 2.00% nickel

The castings were prepared as previously described. Figure 5 is a photograph of the polished and etched sections, and Table 1 lists chemical analyses and grain count data. Examination of the castings and data indicate that zinc has little, if any, effect on grain size of copper. It is difficult to evaluate the effect of nickel but it may be that some slight refining is

present in the nickel-containing alloys. When the grains are as coarse as those indicated, location of the longitudinal section becomes more significant. Since grains are three-dimensional and the etched surfaces reveal only two dimensions, there is room for considerable variation in intercept numbers. As the number of grains increases, grain shape becomes less significant. This being the case it is difficult to determine whether nickel has had any effect at all.

In this series of castings the addition of 5 per cent zinc to copper has markedly changed the shrinkage characteristics. Severe external shrinkage was found in all castings containing 5 per cent zinc. A representative casting is shown in Fig. 6. Some internal shrinkage porosity was also found in the lower half of the casting with the zinc containing alloys. From all indications, nickel has had no observable effect on shrinkage of the copper-zinc alloys.

Copper; +5% Tin + 5% Lead + 5% Zinc

In addition to knowing the effects of nickel on the binary alloys containing the elements used in making 85-5-5-5 alloy, it is desirable to know the effect of nickel in the quaternary alloy. For purposes of such an experiment a pure 85-5-5-5 alloy was made up in a 25-lb heat in the Ajax-Northrup induction furnace. The melt was poured into 1 in. diam cylindrical molds. The remelting stock was then machined to remove the outside skin and cut into pieces for further use.

The procedure followed in this series was that already described. The castings made had nickel contents of 0.05 per cent, 0.10 per cent, 0.25 per cent, 0.50 per cent, 0.75 per cent, 1.0 per cent, 2.0 per cent. In addition, one casting was made of high purity 85-5-5-5 with no nickel, and another run using commercial grade 85-5-5-5 alloy. One other sample was run, a duplicate of the high purity 85-5-5-5, was used for making a cooling curve as mentioned before in this report.

The castings from this experiment were prepared as described previously. A photograph of the polished and etched sections is shown in Fig. 7. In Table 1 are listed the chemical analyses and grain count data. Examination of the results indicates that nickel has had some small grain refining effect up to 0.10 per cent. From 0.1 per cent nickel to 1.98 per cent nickel the grain size has increased progressively indicating a slight coarsening effect. The commercial 85-5-5-5 sample, N-1, with 0.44 per cent nickel (and other impurities) has a grain size comparable to the high purity 85-5-5 alloy so that there have been no unusual effects from combined impurities.

External and internal shrinkage porosity was absent in all cases so that nickel has exerted no observable influence on this characteristic.

The cooling curve produced in this experiment is shown in Fig. 8. The rate is about 35°F/min. The arrest indicates that freezing begins in a high purity 85-5-5-5 alloy at about 1835 F.

Since the cooling rate used in these experiments was moderately slow it was thought worthwhile to determine whether the effects of nickel as noted with



Fig. 6-Photo showing external shrinkage of 5% Zn-Cu alloy.

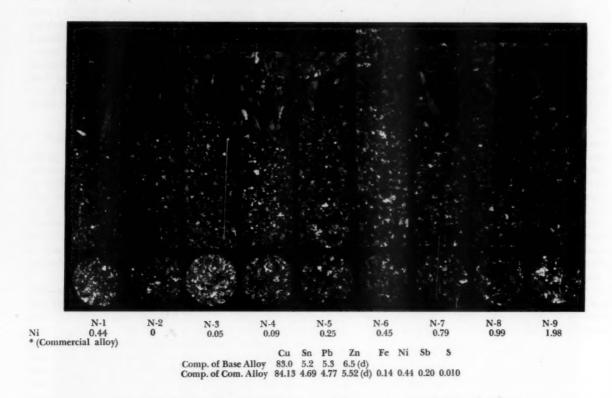


Fig. 7-Macrographs of high purity 85 Cu, 5 Sn, 5 Pb, 5 Zn alloy series.

a furnace cool would change with a faster cooling rate. For this experiment two 500-gram melts of high purity 85-5-5-5 alloy were made side by side, in clay-graphite crucibles, melted in the Ajax high-frequency furnace. When the metal was molten, 0.50 per cent nickel shot was added to one of the melts. When the metal temperature reached 2280 F, the furnace was turned off and the two crucibles poured simultaneously into two graphite crucibles of the type used in the other experiments. It was hoped to obtain a high rate of chill by pouring into graphite. For all practical purposes the two castings can be assumed to have been poured at the same temperature and to have solidified at the same rate.

The castings were machined, polished, and etched. They are shown in Fig. 9. It can be seen that the grain size of the castings is much finer than any obtained by furnace cooling. The grain size of the casting with nickel is substantially the same as the alloy without nickel, although there may be some slight effect from nickel. The effect of cooling rate,

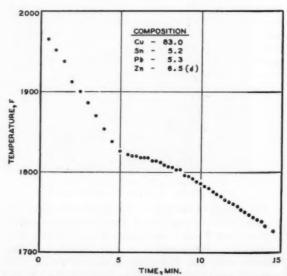
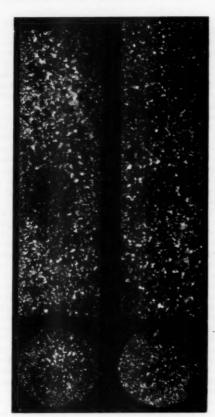


Fig. 8-Cooling curves of high purity 85-5-5-5 alloy.



Nickel		L-11 0	L-12 0.52%	
		Alloy Com	position	
	Cu	84.25	Pb	4.97
	Sn	5.07	Zn	5.71 (d)

Fig. 9-Macrographs showing effect of nickel on grain size of rapidly cooled high purity 85-5-5-5 alloy.

however, is great and apparently, of far more significance than minor changes in composition.

Discussion

Consideration of the results of these experiments indicates that grain size of cast copper-base alloys is influenced by chemical composition. Composition, however, is apparently less important than cooling rate in determining size of grain. The effects of major alloy constituents on the grain size of copper would seem to depend on the metallurgical nature of the alloy. Lead is only slightly soluble in solid copper and, therefore, during solidification, lead will remain molten long after the copper matrix freezes and thereby restrict grain growth. As a result, the grains remain small, and are rounded at the edges since each grain must be in contact with lead over a considerable portion of its surface during solidification.

At the other extreme, zinc is completely soluble in copper in a 5 per cent zinc alloy. The alpha solid solution formed would behave very much like copper itself; as a result, the grains resemble those of copper cooled under the same conditions.

Tin is also soluble in copper in a 5 per cent tin alloy. There is some possibility of inhomogeneity in a slowly cooled copper-tin alloy, as well as the possibility of a second phase forming. The fact that the solid solution formed will behave less like copper itself than does the alpha phase of the copper-zinc system would offer the possibility of grain size variation. The results indicate some refining action from the tin, less than that from 5 per cent lead, but much more than that found with zinc alone.

The grain size of the high purity 85-5-5-5 is the same as the 5 per cent lead-copper alloy. Therefore, it is reasonable to assume that the presence of lead in the alloy has controlled the grain size and the zinc and tin have had negligible effect. It would appear that in any copper-base alloy containing lead over some minimum content, the grain size, independent of cooling rate, will be influenced strongly by the lead. If compounds are formed, or other insoluble elements are present, the effect of lead may be overshadowed, but in solid solution alloys, lead would be a grain-refiner.

The effect of nickel on grain size has ranged from negligible to relatively small in the alloys studied. Compared to the effect of either lead or tin the effect of nickel as a grain refiner is minor. With the 5 per cent tin alloy some refining has occurred with low nickel contents. This may be due to nickel-tin compound formation. Some refining action was also noted in the high purity 85-5-5 alloy with very low nickel additions. Again, this may be due to nickeltin compound forming. In every case where nickel shows any grain refining tendencies the effect seems to be strongest with the first increment of nickel-up to 0.10 per cent. In some cases the effect disappears with increasing nickel content. Since commercial 85-5-5-5 specifications permit up to 0.75 per cent nickel content, and alloys usually contain around 0.50 per cent, it would seem as if any beneficial effects of nickel on grain size would be assured.

With more chilling and higher cooling rates there seems to be no significant difference in the effect of nickel. The grain size of a chill cast (in graphite) high purity alloy with 0.50 per cent nickel is somewhat finer than the same alloy without nickel. The difference, however, is slight and the effect is probably inconsequential compared to the effect of small variations in cooling rate. Further exploration of the effects of nickel with fast cooling rates might prove of interest.

With regard to other claims frequently made for nickel additions to copper-base alloys, the results of this work indicate that the effects of nickel on shrinkage have been negligible. No shrinkage was found in any of the copper-lead alloys or the high purity 85-5-5-5 alloys, with and without nickel. Some internal shrinkage was found in the copper-tin alloys and both severe external shrinkage and small internal shrinkage was found in copper-zinc alloys. Wherever shrinkage was observed it seemed to be independent of nickel content. In the experiment where a high cooling rate was used similar shrinkage areas were found in both the casting containing 0.50 per cent nickel and the one containing no nickel. It would appear that

nickel has no measurable effect on shrinkage with both slow and fast cooling rates.

The results of microscopic examination of copperlead alloys with and without nickel indicate that there has been no change in distribution or size of the lead constituent with the presence of nickel. Nickel may influence lead distribution more with faster cooling rates but it is difficult to understand metallurgically why nickel should have any effect on lead in copper alloys since nickel is completely soluble in copper and the solid solution must behave much as does copper itself.

Summary and Conclusions

A group of experiments have been conducted to determine the effect of nickel on the grain size of copper plus 5 per cent lead, copper plus 5 per cent tin, copper plus 5 per cent zinc and high purity 85-5-5-5 alloy. By eliminating cooling rate as a variable it has been possible to measure effects of nickel content up to 2 per cent on grain size of castings cooled at a rate of 35 F/min. From the results of these experiments, it is possible to draw the following conclusions:—

1. Composition of copper-base alloys influences grain size but the effect is far less significant than cooling rate. A high rate of cooling is apparently the best method for obtaining fine grain in a cast alloy.

2. Five per cent lead refines the grain of copper markedly. Lower lead contents may contribute to grain refinement of commercial copper-base casting allows

3. Nickel up to 2 per cent in a 5 per cent Pb-Cu alloy has negligible effect on grain size.

4. Five per cent tin refines the grain of copper but less than does lead.

5. Nickel contents up to 0.10 per cent have some grain refining effect on Cu-5 per cent Sn alloys. Higher nickel contents do not seem to have any further effects.

6. Five per cent zinc has negligible effect on the grain size of copper.

7. Nickel contents up to 2 per cent have no measurable effects on the grain size of Cu-5 per cent Zn alloys.

8. Nickel contents up to 0.10 per cent refines the grain of a high purity 85 Cu, 5 Sn, 5 Pb, 5 Zn alloy slightly, but the effect disappears and some coarsening occurs with increasing nickel content.

9. With a fast cooling rate (chill casting) high purity 85-5-5-5 alloy with 0.50 per cent nickel has very slightly finer grain size than does the same alloy without nickel. The difference, however, is extremely small.

10. Nickel contents up to 2.0 per cent have no observable influence on shrinkage characteristics of any alloys studied with both slow and fast cooling rates.

11. Nickel contents up to 2.0 per cent have no measurable effect on distribution and size of lead constituent in a 5 per cent Pb-Cu Alloy.

12. From a cooling rate curve at 35 deg/min, high purity 85-5-5-5 alloy begins to freeze at 1835 F.

DISCUSSION

Chairman: F. L. RIDDELL, H. Kramer & Co., Chicago. Co-Chairman: H. G. Schwab, Bunting Brass & Bronze Co., Toledo.

A. K. Higgins: ¹ I found this paper most interesting. The authors did a fine job. I have certain questions and suggestions not in connection with castings but in connection with measurement of grain size in alloys.

We have investigated methods of determining grain size with perhaps six of the larger wrought alloy companies. We have identical photomicrographs out for examination and I think they have been read for grain size by perhaps 50 different men, presumably experts in the field. The lack of agreement on the grain size of specific photographs has been astounding.

We find that for equi-axed grains (either wrought or cast) the use of ASTM standard photographs offers the most reliable means of judgment.

The extension or change of the method to apply to cast alloys is easily done although it does not apply to columnar grains.

What deoxidation was used, if any, on the lead series of alloys? I would have a hard time judging the furnace conditions, because normally an induction furnace gives highly oxidized heat, but with graphite crucible melting the atmosphere may be different.

I noted in the tin series, the next to highest nickel content was apparently out of line in grain size as compared to the rest of the group. Is there any explanation for that, or can it be accounted for by the random variations in the experiment?

MR. COLTON: In reply to his comment on measuring grain size, I think Mr. Higgins probably well appreciates the difficulty of trying to do it intelligently on a small sample. We found the best way was to compare samples side by side. We knew our method was not rigid, but it was simple. We could never trust one person to do it, and had to have about six people make grain counts independently.

With regard to the question on the deoxidizer used, the furnace itself is not oxidizing under these conditions because the graphite reacts to produce enough reducing gases to keep the surface clean at all times. We did not think it necessary to use a deoxidizer, nor would we have dared to, because we would have confounded the results by introducing another element into the system.

Mr. Higgins' question about the apparent anomaly in our tin series may be explained by considering what may happen in sectioning pieces like this with relatively coarse grains. Any particular longitudinal section may be inconsistent because we are looking at only two dimensions. You must always remember that you are looking only at two dimensions when in reality there are three.

CHAIRMAN RIDDELL: In regard to the ASTM method of grain size determination suggested by Mr. Higgins, the large grain size present in the copper-zinc alloy might possibly show only one grain at 75 diameter magnification and thus give doubtful results.

J. S. Vanick: ² Mr. Colton has given us a thought-provoking paper. We have been through this work as he has indicated and I thought you ought to know what we did at the time we did it.

The procedure for grain size determinations that has been described in this paper, is an excellent description for what you would do if you were studying ingots. Men who handle that type of metal like to go along with the notion that nickel does not do much for them in the matter of grain size. You have a very slow cooling rate in an ingot which Mr. Colton's procedure provided. It would be interesting to know what thickness the casting would need to have to provide the slow cooling rate. As near as I could guess, it appeared to be 31/2 to 4 in. thick, based upon a 35-deg cooling rate per minute, for a 100 to 150-deg temperature drop.

The Colton experiments, confined to small specimens produced by the slow solidification in the crucible, of I-lb heats from an assumed constant temperature, provide a theoretical yardstick of a single cooling rate which may have been ideal for the production of a coarse grain. The disturbances common to the melting, pouring and solidification of metal cast into a mold,

¹ Supt., Allis-Chalmers Mfg. Co., Milwaukee, Wis.

² Development & Research Div., The International Nickel Co., Inc., New York.

0.

he

ns

re-

th

ve

ev

in

le

YS

18

at.

be

SI

in

ty

ie

11

ce

16

16

ld

o

in

in

17

ve

er

iv

ly

12

d

n

ıt

a

re

16

d

g

over a range of cooling rates, should be much more decisive in staging the conditions that define grain size in commercial castings.

In our tests, we felt that it was desirable to simulate foundry practice and pour castings into sand molds in such a way as to obtain the chilling effect of a mold surface with its attendant influence upon the nucleation of grains and the accompanying segregation or flow and movement of the solid-liquid interface. Segregation in these alloys at liquid temperatures is lightning fast for precipitating solids, and even faster for precipitating gases. These precipitates move ahead of the colder solid front and lead to the center-line structures commonly called shrinks, which may merely consist of fine-grained material or represent areas contaminated with impurities or gas residues. In a casting, these solidification phenomena, moving from a mold surface, produce a fine grain which may later be dissipated into coarser grained interiors if the internal cooling rates are sufficiently slow or if inhibitors to grain growth are not involved in maintaining a fine grain. The temperature gradient that is produced in the mold becomes a necessary factor in establishing casting grain size. This feature was not reproduced in the Colton tests because the metal was allowed to solidify slowly in the crucible. It is doubtful whether such turbulence could develop in a 1-lb heat as compared to the commotion attending the pouring of a mold even in such simple shapes as 11/2-in. diam or 9-in. long bars. Specimens of this size were used in our tests.

In addition to cooling rate as a factor, there are a number of chemical and mechanical disturbances to the solidification of a liquid metal that would encourage the formation of a fine grain. Among these are evolution of gas, the turbulence in casting, the inclusion of impurities, and the precipitation of finely dispersed solids. In the bronzes chosen for this work, a lead bronze, particularly at 5 per cent lead, would produce a considerable shower of inclusions which would be soft and possibly liquid, but sufficiently obstructive to interfere with normal grain growth. These factors were known to us in our studies of grain size. They also conform to the general laws governing crystallization. At this point, the methods used to study grain size come in for careful consideration.

We counted grains by the ASTM method in the area of the field, and therein another factor enters that has not been mentioned up to now. You can have, in some of these coarse looking grains, several small grains that will not be seen in the single setting. You have to rotate the microscope field to find that there are three or four grains within a single appearing grain boundary because the three or four grains have the same orientation. In the published macrograph photos, you see only the apparent coarse grain outline.

Our tests also standardized on a measurement of grain size across a section 1½ in. up from the base of the 9-in. cylinder. The grains were carefully counted individually in the field. The area was rotated. Frequently the grain size might appear to be of mixed large and small grains, but an adequate count would establish the prevailing average.

The pouring temperatures were carried through four temperature levels. This introduced corresponding differences in

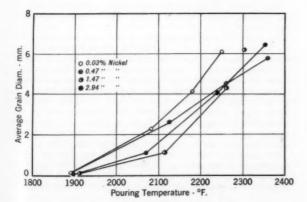
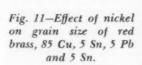
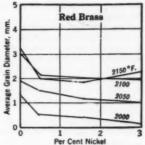


Fig. 10-Grain size for nickel-alloyed bronze, 89 Cu, 11 Sn, Ni as shown (Kihlgren).

cooling rate. These variations were considered necessary and desirable. They might be simulated by using a wedge-shaped test bar or step bars or step cylinders to provide differences in cooling rates. With these changes in cooling rate, it becomes possible to project the change in grain growth that occurs particularly with high pouring temperatures which would correspond to slow cooling rates. The results showed that it is possible to pour a nickel-free composition at a lower temperature and achieve a grain size as small as that obtainable with nickel additions poured at higher temperatures. The diagrams (Fig. 10 and 11) for the various bronzes also showed that the difference in grain size were smaller for the leaded alloys. In fact, the greater





the quantity of lead, the smaller the difference in grain size for different pouring temperatures. This observation confirms the general belief that insoluble ingredients get in the way of growing crystals and, by so doing, produce small grains or interfere with grain growth.

By these methods of test, the grain size for metal poured under foundry conditions at varying cooling rates, demonstrated that a desirable cooling rate might be obtained in which grain growth was negligible. If, along with a negligible grain growth, ingredients were present which interfered with grain growth, then the function of addition alloys might be negligible. The practical requirements in the foundry are such as to make foundrymen use sufficient superheat to run a casting under whatever mold conditions are presented and, in doing so, the cooling rates vary widely from the ideal. It is for applications such as this that some measure of the influence of different cooling rates, as obtained by the use of successively higher pouring temperatures, can become instructive in demonstrating harmful or useful effects upon grain size.

In conclusion, I would be interested in any explanation Mr. Colton could give us regarding the relation between his cooling rate and a corresponding thickness of a casting poured into sand; also, an explanation for the apparent coarse grain shown in the top side of these small ingots.

Mr. COLTON: I will answer first the question of slow rate of cooling applying to ingots only. In an investigation of this type it is essential to have uniform cooling rate between samples, and the absolute cooling rate is less important than having any selected rate the same for all samples so that they can be compared. When we had finished the work, we decided we did not know whether our observations would be similar with a fast cooling rate. Consequently we ran the test described in the paper with the high cooling rate. This technique did give us a much finer grain size. There was very little grain size difference however in the samples with and without nickel. These experiments may not cover the whole range of possible grain sizes, but a fast cooling rate and a slow cooling rate do simulate conditions found in sand castings.

Regarding the experiments described by Mr. Vanick, I do not believe it is possible to control conditions accurately enough in pouring castings to obtain identical cooling rates. We do not think experiments of that type are reliable in the measurement of factors affecting grain size. Since cooling rate itself is probably the most effective factor in determining grain size of cast metals, it seems reasonable to use techniques that minimize variation in cooling rate: pouring castings would seem to offer considerable chance for such variations.

We took exception to Mr. Vanick's data when originally published because we do not believe the differences in grain size

attributed to the presence of nickel are so great that they have any significance. That is not to say, that nickel does not have some influence, but from the published data it is difficult to draw

any definite conclusions on the effect of nickel.

In regards to whether the temperature in our crucibles was always uniform, we ran experiments which may or may not prove it. We ran cooling rate curves on different heats, at various times. We placed a thermocouple in one of the nine crucibles then measured the temperature at regular intervals as the metal cooled. For each such experiment we would place the thermocouple in a different crucible as far as position in the furnace was concerned. The 35-deg per min cooling rate, we found, was a function of the particular furnace we used. This rate did not change during the various experiments so we felt safe in assuming uniform cooling rate for all crucibles in any one experiment.

The coarse grains at the top of our "castings" were the source of considerable discussion in our laboratory. It so happens in the Research Dept, we have done a great deal of work on copper ingots and quite often exactly the same condition is found in a wire bar. Near the top surface of a wire bar, you will find an area of coarse columnar grains while the balance of the wire bar will have fine equiaxed crystals. The best explanation we have to date is that this effect is a function of the method of

heat extraction.

We decided that the unusual structure did not have great significance for the purposes of our work. We naturally could not use the top half of the bar. The bottom half, however, apparently cooled at a uniform rate, and was suitable for use.

We felt that as long as we compared like things, we were on safe ground. When we ran 11 samples under the same conditions, and used only the bottom 2 in. of the bars produced, we felt justified in making comparisons between specimens.

After due consideration we felt safe in comparing samples one from one experiment with another as long as we used only the bottom portions of the bars. This followed our knowledge that the cooling characteristics of the furnace were constant.

G. K. EGGLESTON: ⁸ Several foundrymen have mentioned the effect of nickel additions when nickel was first being promoted for use in red brasses. Tests were made that seemed to indicate that you got most of the benefit of nickel (assuming there was a benefit) at about 0.75 per cent; no additional benefits were noticed until you went to 2 per cent. The illustrations in the

paper seemed to indicate this same general tendency.

I am interested in the approach the authors used in this investigation. Some years ago, we ran a series of investigations trying to determine what the effect of silicon was on lead brasses. We took the same approach making additions to copper-lead, copperzinc, and copper-tin. Then we ran three-element combinations such as copper-lead-tim. We found interesting results using this approach. We found that silicon had very little if any detrimental effect in many of the combinations, but on several combinations we really got in trouble.

MEMBER: What was the purpose of this study? Were you interested in the physical properties or density or soundness of structure that would be obtained with these grain sizes? I find that often on the larger grain size in red brasses we obtain higher tensile strengths, often accompanied by unsoundness

Mr. COLTON: Our purpose in this work was to conduct a scientific investigation for the sole purpose of obtaining knowledge. What application of that knowledge there will be, I cannot say. In order to better understand the significance of grain size in copper-base alloys, it is necessary to understand the factors that determine the grain size.

MEMBER: Did you do this from a control standpoint?

Mr. Colton: We did not begin our investigation with control of grain size as a primary object although it would necessarily

be one of our ultimate goals.

In scientific investigations you normally have to start at the beginning. We thought this was a satisfactory place to start our work. We found no records of similar work being done. In this work we have studied the effects of composition. The next step will be the study of other variables. Perhaps through investigations of this nature we can design better alloys or minimize some of the difficulties foundrymen have. We are all striving to accomplish that.

Mr. COLTON: I feel certain that the nickel was in solution. The materials used included the highest grade cobalt-free nickel shot available. We weighed it out to the closest 0.1 of a gram. With induction furnaces it is no problem to dissolve small nickel shot in copper or copper-base alloys in 5 min. I am certain it dissolved completely. We sampled from various parts of the bar to make certain we had an average sample.

MEMBER: Were those castings made in carbon molds?

If any heat was transferred from the crucible would that make any change in the grain size too?

MR. COLTON: It must have.

MEMBER: If you used carbon or silicate carbide crucibles?

MR. COLTON: It probably would.

G. P. HALLIWELL: A In connection with the large grains on top of the solidified metal, I would like to interject here, the possibility of a critical grain size, due to some impurity, which might be a gas. There are many examples of this phenomenon, the most classic of which is the original work on tungsten lamp filaments, where thoria was introduced to control grain size. However, a critical amount produced exorbitant grains. Perhaps some critical amount of impurity, gas for instance, was present during the latter stage of solidification.

MR. COLTON: That is a possibility we did not consider, but it could be so. For our purposes, however, I can go back to the statement I made before about comparing like things. I think we are on safe ground as to why the coarse grains formed. As long as we used the bottom part of a bar the coarse grains did not enter into our considerations. We felt that the cooling rate must have been such that at that particular location in the bar

the grains were coarse.

G. M. Thrasher: ⁵ I would like to question the matter of zinc addition to samples. The small amount of zinc (5 per cent) would not act the same way that the tin would act. It would take about 15 per cent to approach the same degree of alloy as the tin samples and the explanation of that shrinkage may be due to some volatilization of the zinc in the sample.

MR. COLTON: In regard to the effect of 5 per cent zinc as compared to 15 per cent zinc, I do not think there would be any great difference in range of grain size. Zinc forms a solid solution with copper, while lead is not in solution. Tin, up to a certain percentage, is in solution. It would appear that the effect of an element on grain size depends on how it combines with copper. We are interested only in these alloys. What happens in others we do not know. It is quite possible that a separate investigation would be needed for every alloy under consideration.

We do not believe from the evidence here that any higher zinc content up to 15 per cent would make any great difference. For the zinc alloy studies we melted down a master alloy of 5 per cent zinc and copper instead of trying to add zinc to copper in our little crucibles. Our final results showed 5.8 per cent zinc and we probably did not add more than 6 per cent originally. Our zinc losses were obviously low.

Mr. Thrasher: How much loss did you have in melting down

the final samples?

Mr. COLTON: In the original master alloy samples we added 6 per cent zinc which allowed for losses on remelting. Our final

zinc content was 5.8 per cent.

Member: We have had occasion to cast some very large (100-lb) bearings with 8½ per cent tin, ½ per cent nickel and the balance copper. On machining these bearings the grain structure became evident. The grain size seems to vary and we cannot find the reason for it except for the pouring temperature. When we pour the castings at low temperature we get very fine grain size and with that we get some of the defects that you would expect when you pour cold. Yet, when we pour in the recommended range we invariably get this large grain size. The castings appear to be sound. At one time we even got three or four grain sizes, some very small and some very large.

Mr. COLTON: Your observation of grain size showing up in machining is a common one. In polishing yellow brass hardware castings it frequently occurs, too. It is not unexpected. Your observation that pouring temperature is a most critical factor in

MEMBER: Are you sure the nickel you added was in solution in those small heats?

⁴ Director of Research, H. Kramer & Co., Chicago.

³ Engineer, Barnes Mfg. Co., Mansfield, Ohio.

⁵ R. Lavin & Sons, Inc., Elmira, N. Y.

OYS

n in

The

shot

dis-

r to

ake

on

the

ich

on,

mp

ize.

aps

ent

but

the

ink

As

did

ate

bar

inc

nt)

uld

as

be

lid

to

he

nes

p

n

ner

er-

OV

to

er

wn

6

al

m.

IC-

we

re.

ne

DU

he

he

or

in

re

in

the determination of grain size is one that many others have made, too. I believe pouring temperature to be among the most significant of all the factors.

WM. ROMANOFF: 6 As long as there has been such a big difference of opinion as to the value of coarse or fine grains, it has also been agreed that this paper has not proven anything in a practical way. I would hate to see this discussion discontinued thinking there might be anything detrimental about nickel. Anybody making brass castings today is using ingot because he could not get anything else if he wanted to. As long as nothing has been proven practically, nickel should not be eliminated from specifications. Otherwise, there would be no mate-

rial available in ingot form.

M. J. Davison: T I would propose that the difference in grain size observed between the top and bottom portions of any one of these vertically sectioned ingots is due to the existence of a critical cooling rate.

The cooling rates are such that at the top the rate of grain growth exceeds the rate of formation of nuclei, whereas at the bottom the rate of formation of nuclei is greater than the rate of grain growth.

Has the author any indication as to what is the minimum level of lead at which it begins to exert an observable effect on the grain size?

Mr. Colton: The first thing we want to investigate in the next phase of this work is how little lead must be present to affect grain size. Your explanation of the coarse grains is a good one. There is a sharp line of demarcation so that your idea sounds most reasonable. If there is a critical rate how can it be found?

In reference to Mr. Romanoff's remarks, we do not suggest removing nickel from copper-base alloys. Most commercial bronze specifications permit 0.75 per cent nickel, and most commercial alloys have about 0.5 per cent nickel present. Since from this study there is apparently adequate nickel already present to give the desired beneficial effects you might conclude there is no reason to add any more nickel. That does not mean that more nickel cannot be added, but we cannot see any great advantage to further additions because there seems to be enough nickel in most commercial alloys to give whatever grain refining effect can be expected.

J. C. PENDLETON: 8 The melting and pouring temperature is most important. We bring the melt up to 2250 or 2300 F.

MR. COLTON: It is certainly true with many alloys that if you do not thoroughly melt and mix the molten metal segregation is a strong possibility.

MR. PENDLETON: It is difficult to alloy manganese bronze unless you heat it up to 1950 F.

MR. COLTON: There is no doubt about the necessity for having adequate temperature of the bath to thoroughly alloy manganese bronze and aluminum bronzes.

CHAIRMAN RIDDELL: I would like to ask Mr. Colton a few questions on his remarks about shrinkage with the nickel additions. He mentioned that there was no apparent affect on shrinkage. I think possibly that this might be tied up with the slightly reducing conditions which would tend to have an effect on the shrinkage characteristics of the various alloys.

Another thing, I think some work was done showing that an increase in the nickel content changes the solidification range of an alloy. The work was done on a tin bronze. To some degree do you not think that the change in the solidification range will have an effect on shrinkage?

MR. COLTON: There is the possibility of gas in the metal. You are entirely correct, the metal may have contained gas. All melts, however, would have had the same amount of gas, but different amounts of nickel. Presumably under those conditions, the gas had little or no effect. We had reason to believe, however, there was no gas in these alloys.

I have never been able to understand exactly how I per cent nickel can change the solidification range of these alloys appreciably. I cannot understand this any more than I can the reasoning that phosphorus changes the solidification range of tin bronzes. I have never been able to measure these alleged changes. It is difficult to see metallurgically how they can happen.

It would seem that a small amount of nickel must have a minimum effect. In the course of making castings, think of all the other things we know have large effect, and think of the small effect this must have.

CHAIRMAN RIDDELL: I think they went up to 9 per cent nickel and the liquidus line went up gradually and the solidus lines went up quite steeply which would make the solidification range much shorter. At 9 per cent it was quite narrow compared to that at no nickel.

If I remember the curve right, the points on the liquidus and solidus lines were determined at 1 per cent increments. The solidus line was determined metallographically and the liquidus line by cooling curves.

W. H. JACOBSON: 9 I am interested in the remarks the author made with reference to wire bars. When the oxygen content of wire bar is approximately 0.035 per cent you get a coarse grain structure. As the oxygen content is increased say to approximately 0.05 per cent, we note the area of coarse grain structure is decreased with a corresponding increased fine structure area. This coarse grain structure, we believe, is due to the absorption of reducing gases which naturally increases with the decreasing of the oxygen content. I am wondering if the coarse grain area in the author's investigation might also be due to reducing gas absorption.

Mr. Colton: I do not really know but we considered our metal more or less gas-free.

Mr. Jacobson: Is it not possible that you would get a reducing

gas absorption due to the graphite in your small crucible?

MR. COLTON: One answer to that, I believe, is that the continuous casting process is carried out in all graphite system with no apparent ill effects.

Mr. Jacobson: However, the molten metal in your continuous cast solidifies at a faster rate and does not remain molten at high temperatures for as long a period as your procedure indicates.

MR. COLTON: The cooling rate varies somewhat but the product usually has close to theoretical density. This is in a reducing system. I agree, however, that your suggestion may be a possible explanation.

CO-CHAIRMAN SCHWAB: We had occasion to try to add coppernickel shot as ladle addition to alloys melted in a cupola where the metal temperature is much higher than was used here and we found it extremely difficult to get the shot into solution. In fact, there were times when we would pick the shot out of the casting if we were not careful. That probably would bear investigation.

MR. COLTON: The mixing characteristics of the induction furnace are excellent. In order to make exact additions it was necessary to use very small pieces of nickel shot and they would dissolve at the surface of the melt. We tried to do our alloying at a temperature of 2150 F but in the melting process we may have raised the temperature higher. Close temperature control is difficult to obtain with high frequency furnaces.

CO-CHAIRMAN SCHWAB: I thought you made your addition somewhere slightly above 2000 F.

Mr. Colton: At least 5 min was allowed for mixing. We found no nickel shot in our castings and all our analyses were uniform.

CHAIRMAN RIDDELL: Mr. Colton, on melting in an induction furnace in 11 crucibles, does each crucible act like a separate furnace? I admit that in a single cavity in a high-frequency furnace you get good mixing, but will 11 of them act like individual crucibles with regard to mixing action?

MR. COLTON: Yes, each crucible apparently acts like a separate furnace with regard to stirring action of the metal.

Mr. VANICK: I am still unsatisfied with the 5 per cent zinc addition. Why did that shrinkage occur?

Mr. Colton: Apparently all of these alloys had an opportunity to feed. Since the coarse grains were at the top of the "casting" the hottest metal must have been at the top. We found no pipe in any of the bars, so that any shrinkage had to be inter-nal or external below the top. In the zinc series the wall would collapse. It is possible that 5 per cent zinc changes the amount and nature of the solidification shrinkage.

⁶ Vice President and Technical Supt., H. Kramer & Co., Chicago.
⁷ Metallurgist, The Canada Metal Co., Toronto, Ont., Canada.

⁸ Supt. of Foundry & Pattern Shop, Newport News Shipbuilding and Dry Dock Co., Newport News, Va.

Plant Process Control, Nassau Smelting & Refining Co., Tottenville, Staten Island, N. Y.

"SPEC" SHEETS AND THEIR VARIOUS USES

By John Taylor*

ABSTRACT

In this paper the author discusses the need for the use of specification sheets in the foundry for profitable operation. A specification sheet gives you something specific with which to back up quotations on casting orders. With the details of flask size, type of molding machine used or floor used, whether or not facing sand was used, etc., it is easier to justify a cost with this kind of information than it is with an approximate estimate.

Specification sheets for foundry work and blueprints in a machine shop are practically synonymous in purpose. Blueprints have been used in machine shop work for many years because it was found a long time ago that machine operators had to have specific details on the length of the cut, the diameter of the hole, the tolerances, etc. Before the use of blueprints, sketches were usually made by the mechanical engineer or the designer and the machine work was carried out accordingly, but as our production machine began to grow, it was found necessary to keep the original drawings or sketches in the engineering department and pass the information along to the machine shop in the form of blueprints.

In the meantime, the assembling of machine parts was still left to the ingenuity of the supervision and workmen involved in this work. However, in recent years it has been found advisable to use sub-assembly or assembly drawings, and in many cases specification sheets so that sub-assembly and assembly work would be done according to the original design or, in the words of the home craftsman, "complete the assembly without having any parts left over."

Today it is almost impossible to walk into any machine shop or assembly plant, particularly any of our high production companies and find them doing work without the aid of blueprints or specification sheets. This information not only is a specific guide to the workman but it eliminates the vest pocket notebooks and the practice of the foreman trying unsuccessfully to carry all of the necessary information in his head.

Comparing the progress made in supplying sufficient and proper details to plant supervision and

operators in the metal trades field to the same fundamental practices in the foundry industry, a large gap is found to exist in comparable progress, particularly in jobbing and semi-jobbing foundries. Until the past 15 years, the use of specification sheets or cards was practically an unknown practice. New jobs would come into a foundry and, after a huddle between the superintendent and foreman and sometimes the molder, a decision would be reached as to the best method of making the mold. In many instances this would include the size of the risers or heads, the number of gates, the depth of the cope and drag, etc. The foreman would usually mark down this information in a notebook or some mark would be made on a pattern with chalk whereupon the job was ready for the molder. Assuming that the decisions of the superintendent and the foreman resulted in the molder being able to produce a good mold, the balance of the order would be run and shipped.

A Guide for Repeat Orders

When the same castings were again required by the customer, the same procedure would be repeated because no records were kept as to how the mold was made, whether or not the cope was vented or even what size flask was used. As the result of this kind of practice many foundries have lost thousands of valuable hours of supervisory time and in many cases produced unnecessary scrap simply because sufficient time was not spent at first to record the details of the job.

Although some foundries have realized the value of specification sheets or cards and are now using them to good advantage, there is still a large percentage of foundries who have no records on how any particular casting is made. It has often been said by some foundrymen that the keeping of specification sheets would require a lot of filing space and yet these same foundries maintain large pattern storage facilities without hesitation.

Some foundrymen may reason that a good molder needs no specification sheet on how to make his mold but careful analysis will prove that this is not true. The old time completely trained molder can make a good mold without a specification sheet but in doing so "feels" his way. If this same job repeats itself after

[•] Vice President, Lester B. Knight & Associates, Inc., Chicago.

d

e

d

1

a lapse of two or three months or more, he must again "feel" his way because his memory could not serve him well enough to remember all of the little details as well as the big ones. Another point in favor of the use of specification sheets is that foundries are faced with the problem today of having fewer and fewer skilled molders who can produce a mold satisfactorily. By the same token, a large percentage of today's molders as well as coremakers are not apprenticed workers and therefore do not have the job knowledge of the molders and coremakers of 20 or 30 years ago.

SPECIFICATION CARD FOR MOLDING

FRONT

MOLDING PROCEDURE AND STANDARD DATA					
Metal	Pattern Number				
Pouring Temperature	Pattern Type				
Ladle Additions	Pieces Per Mold				
Sketch	Sprue				
	Runner				
	Risers				
	Flask				
	Location				

BACK

Elements	No.	Std.	Allow.	Elements	No.	Std.	Allow.
Constant				Cut Sprue Ream			
Bands				Blow Cope			
Dust Parting Spray				Peen (Inches)			
Chaplets				Draw Cope			
Heap Riddle Facing				Draw Pattern			
Tuck				Set Cores			
Nails							
Chills							
Vent							
				Minutes/Mold			

Specification sheets can and should be used, particularly on molding and coremaking in jobbing, semi-jobbing and production foundries because it is in this type of work where the greatest losses occur today through the lack of proper information. This recorded information can be established simply and effectively at the time the foreman gives the necessary instructions to the worker. After it has once been established it will save the foreman's time and tend toward better accuracy because the written record is always better than the foreman's memory.

The simplest form of specification can be kept on a card indexed according to customer and/or pattern number with a rough sketch of the casting showing the position of the gates, runners and risers and the corresponding sizes. Adjacent to the sketch should be noted the other important details, such as length and width of flask and the depth of the cope and the drag. In this same location on the card should be

noted whether or not the cope must be vented and if possible the pouring temperature and analysis of the metal. All of this information must be determined before a job can be run and it will only require a few additional minutes to record this information on an appropriate card. In the small jobbing foundry these cards can be kept on file in the molding foreman's office so that they will be convenient at all times. In the larger jobbing, semi-jobbing or production foundries this information is usually kept on file in the Methods or Standards Department where it is available to the foreman at all times. In many cases these data can be made a part of the molding order so that the foreman gets this information when he gets the order to make the castings.

Specification and Standard Data

In more recent years it was found advisable to combine the specification data with the standard data used for the establishment of time values. By making this consolidation of information management has the information on the method and a breakdown of the time allowed for each element of the operation. The consolidated form, covering both method and time allowances per operation, has now been projected into a very important management tool because it can, if properly used, serve a variety of uses. For example, it supplies the foreman with specific data on how the job was made the last time and it also supplies in detail the distinct elements of the operation for which time has been allowed.

SPECIFICATION SHEET

CORE BLOWING

					BOX	NO.	No.		
					MIX	NO.			
EIGHT	OF	вох	L	Bs.	(USE		LI	BS.)	
EIGHT	OF	CORE		LBS.	(USE _			LBS.)	

	Allowance			
Constant	07 h			.413
		ox and sand o oven + box	es per plate	
	of sand p			
		Nails		
Per Box	Time	Per Box	Time	
1	.097	4	.182	
2	.127	5	.196	1
3	.157	6	.206	
	-		NUTES/BOX =	
			NUTES/CORE = URS/100 CORES =	

SET BY ____

A simple specification sheet, covering core blowing, will illustrate the necessary details. In this instance, the information needed is the pattern number, the box number, the weight of the box, cubic inches of sand, the cores per box and the boxes per plate. That information is the general job description. The other details needed are the travel from the blower to the oven, if the operator carries his work to the oven, the number of nails, if any, in the box and the sand mix. That is all the information that would normally be required to fill in the specification sheet for core blowing.

Job Requirements Noted

With this information, it is possible to establish an accurate time standard or piece work price on the job. However, the specification sheet for miscellaneous work on bench coremaking does require some additional information. This operation, in addition to the above data, requires the length, width and depth of the core in order to compute the volume of sand and the area for strikeoff. The data must also specify whether it is a solid single box or a solid multiple box, whether it is a split box, single or multiple and whether the sand is put into the box in a horizontal position or vertical position. That is the necessary job information From these data the sand in time, the peening time, the drawing time and the strikeoff time can be calculated. The standard setter must also have information on loose pieces, number of chills placed or tucked, number of straight or bent rods, the number of nails and how they are placed and the other usual details concerned with bench

With this information on one sheet, the standard setter, as well as the foreman and the operator, then know exactly what is required in order to make the job and exactly what details are involved as to rods, nails, chills, whether or not the operator was given sufficient time for plate handling, whether or not an allowance was made for bedding and all of the necessary details that go to make the operation complete. The setting of standards or the setting of prices by this method does not require any great amount of time. After the information is obtained in regards to size of box, amount of sand, nails, chills, etc., it is then possible to set the average bench coremaking standard in approximately three to four minutes and with the simple specification sheet for core blowing, it is possible to compute this standard in roughly two minutes. In other words, the greatest amount of time involved is getting the necessary information posted on the sheet. After it is on the sheet, the balance of the procedure is simple and can be accomplished by practically any one with little, if any, knowledge about actual foundry practice. To further illustrate the simplicity of the calculations, after the information was posted on the specification sheet, the average coremaker can compute his own standard.

Job Method Noted

The details of the information for the specification sheet are usually obtained from the foreman of the particular department. The standard setter getting the specific details of the method from the foreman, automatically brings about a discussion of methods. In other words, through the discussion between the foreman and the standard setter, the details of the method often bring out elements which can be elimin-

ated or which can be reduced.

In addition to methods and standard setting, specification sheets have two other important uses. The first of these additional uses is very important, particularly in today's labor relation programs because the "spec" sheet definitely and specifically will settle all arguments as to whether or not an operator has been given an allowance to perform a certain element. This feature can shorten a lot of arguments about incentive prices because the worker, the union steward, the foreman and the standard setter can all see the same information and all agree on what elements were recognized, and which ones were not recognized, as being necessary.

Use for Estimating Purpose

The fourth feature of this same specification sheet, which is used for methods, standard setting and labor relations, is that it can also be used just as effectively for estimating purposes. The old fashioned practice and one which still exists in a good many foundries today, of estimating prices on molding, coremaking and cleaning by observation can now be done with accuracy by using the "spec" sheet. Given the proper information from the respective supervisors as to the molding and coremaking details, an inexperienced estimator can arrive at a final price that will be more accurate on individual jobs than the old fashioned observation estimate. Another advantage to this type of estimating is that it is in sufficient detail so that the actual methods employed can be checked with the original estimate and corrections made accordingly if there are any great differences.

To carry this same thought a little further, assume that an estimator quotes on a job that is to be made in a flask 20 in. \times 30 in. \times 10-in. cope and 10-in. drag, and according to the customer the pattern will be mounted so that the work can be performed on a rollover machine. In this case the estimator would use the standard data available for a rollover machine and would compile his estimate accordingly for the molding price. Going further on the same job, assume that when the pattern equipment did arrive it was found that the pattern was not mounted and therefore, would have to be made on the floor. When the molding price would be set for these conditions. it would naturally require a greater length of time for the job to be made from a loose pattern than would have been the case if the pattern was mounted.

ti

th

If

th

jo

be

th

th

to

is

The estimator, in good faith, also assumed that the core boxes were in good condition and estimated on the coremaking accordingly. But when the core boxes arrived it was found that they were in poor condition with the result that although the cores could be made as originally planned, a considerable amount of additional time had to be given to the coremaker to patch and repair the cores so that they would be satisfactory for use. With the proper recording of time for molding and coremaking by the operators and a check against the original quotation by the Cost or Estimating Department, it would be revealed that there existed a considerable discrepancy between the original quotation and the actual cost of the job.

By having a detailed specification sheet, the esti-

mator is then in an excellent position to analyze the cost and find out exactly where the differences occurred which caused the discrepancy. With the specification sheet broken down into elements, he can easily quote the greater differences in cost between the method as originally assumed and the actual method that was followed in the shop. With this information in hand he can approach the Sales Department and insist that the original quotation be revised because it was not possible to make the job according to the original specifications as given to him by the customer. Without a detailed specification sheet, the estimator is not in a position to do this and it is a known fact that in many foundries today quotations are made for one set of conditions whereas the job is often made under entirely different conditions and yet no actual check is made to see why there was such a variation in cost.

In many instances where a specification sheet setup was installed, the specification sheet from the Estimating Department is discussed with the foreman when the job comes in so that the foreman can know immediately just how the original quotation was made. If this procedure is carried out, it will often save the foundry a lot of time and headaches because many cases have been found where the foreman has said that the job could not be made as it was originally planned. At that point the Sales Department can inform the customer that the job cannot be made in accordance with the original plans but must be made under a different procedure and the original quotation must be changed accordingly.

Comparison of Estimate and Cost

A good estimating setup will not operate properly unless there is a definite comparison between estimated and actual prices. It is on this point where many foundries are lax, because they do not attempt to compare the two prices either before or after the job is finished. If this is not done, the most exacting and properly made specification sheet will not be of complete value to any organization because an estimate that is made only for the Sales Department and not used to check against actual cost is not performing its full function.

Too much emphasis cannot be placed on this function of the estimating data because it is at this point that so many foundry managements get themselves into trouble from the standpoint of making a profit. If the estimating prices are carefully checked with actual costs, not necessarily every time, but at least the first time the job is run, management is then in a position to know which jobs are good jobs and which jobs are not good jobs. In other words, the jobs on which they are losing money should be re-estimated and the new price submitted to the customer.

In many cases where it is found that the original estimate is too high, it is also quite advantageous in the long run to advise the customer that the job can be made cheaper than was originally quoted. To say this may seem a little unusual in some cases, but in the long run, it will bring about much better customer relations. If this so-called ironing out process is not carefully followed, foundry management begins to realize smaller profits the moment business starts

to decline. The customer who is being overcharged will be the first one to pull his patterns from the foundry and take them to some other foundry where he can get a better price.

By the same token, the customer who is getting his castings at a low price will naturally leave his castings at that place with the result that the profitable jobs are pulled out and the unprofitable ones remain. Although production may have gone down 10 or 15 per cent, the profit may have entirely disappeared. The foundry is then operating at a loss, even in the face of reductions in overhead that may have been made, because, if the original quotations are actually lower than the true cost of the job, all of the reductions in overhead normally would not be sufficient to overcome the losses that would be occurring.

This particular procedure may be common practice in some foundries, but there are numerous foundries where estimates are made and no comparison is made between the estimate and the actual cost. There are also many foundries where the estimate for each operation is made by the molding foreman or the superintendent merely looking over the drawing or pattern and the core boxes and judging that the production will be a certain number of molds per hour and the coremaking would be a certain number of cores per hours. This is often done with no specific regard for the particular methods that must be employed in actually making the job.

Granted that the use of a specification sheet does take a little longer than the observation estimate, management, nevertheless, does have something specific with which to back up its quotation. With the details of flask size, type of molding machine used or floor used, whether or not facing sand was used, etc., it is easier to justify a cost with this kind of information than it is with an approximate estimate.

Useful Data for Management

Another valuable use for this specification sheet, from an estimating standpoint, is that the estimator is in an excellent position to supply management with any details they may require in reference to the change in job costs if a particular change is to be made in the procedure. Because the data on the specification sheet is broken down into elements, the estimator can, for example, tell management what the change in cost would be if the molds were made in a different size flask or if the cope or drag were changed in size or any one of dozens of different conditions that could change. All of this information can be supplied to management without the necessity of an experimental run or without the necessity of going out into the shop to talk it over with the foreman or the superintendent.

In the final analysis, either would have to give an approximate estimate of what the change would be or actually time the job under the old method and the proposed method. Along these same lines, the specification sheet can also become a methods analysis sheet when the quotation is made up after the work has been started. In many cases, use of the "spec" sheet as a method analysis sheet is very valuable.

INDUSTRY PARTICIPATION IN SECONDARY SCHOOL TRAINING PROGRAMS FOR THE FOUNDRY

By T. W. Russell, Jr.*

"In the long run, the future of any business is more dependent on the kind of people in it than on any other one element. How successful a company is as compared with others within an industry largely depends on its success in developing the right kind of men for management responsibilities."

Mr. Given, the chairman of the board of directors of American Brake Shoe Co. said this in a speech in 1950. He has been building the company on this foundation for over 20 years.

Two important parts of this idea are a planned training program and, second, promotion from within the company whenever possible. Naturally the two go hand in hand.

In my five years with American Brake Shoe Co., only one important position in the company has been filled by a man not already within the organization. This is a vital fact to all of us who work for the company today. We know that if we do a good job, it will be recognized. We also know that men in top management will be continually loading responsibility on the younger men. There is no other sure test of ability. We know we are expected to make decisions, in fact, God help us if we don't! At American Brake Shoe Co. it is a much worse mistake not to take a chance on a well considered proposal than it is to stick one's neck out. A man who cannot make a decision will not be often asked to.

In other words, we know what is expected of us. All we have to do is follow through. We will not be forgotten.

How does American Brake Shoe Co. get men who are willing to accept these responsibilities? How does it train them? Our company has ten manufacturing divisions. Each has its own president and its own operating head, sales head. Each division is a unit, an individuality, performing a complete business function. The divisions operate with a headquarters service section, headed by Mr. Given, including the secretary, treasurer, comptroller, general purchasing agent,

advertising manager, export director, research director, and the departments these men handle. These people are specialists in their jobs and are there to free the division personnel for their prime function—production and sales.

Hiring of Personnel

Hiring people is the responsibility of the divisions, because they are responsible for their people after hiring, and because a measure of the success of a division boss is the organization he builds up to carry on. It would be too easy to blame mistakes, even losses, on men hired by headquarters. However, headquarters does offer help. It works this way: the divisions advise headquarters development section how many men they need and the type of job for which they are to train. At present we are only hiring technically trained men. This means engineers and metallurgists. We feel that an interest in technical subjects is a necessity in the foundry industry today. As a general rule engineers have this interest. We also feel that the odds of finding management material from a group of engineers will be the same as from a group without such training in college. This does not mean that emphasis is put only on scholastic ability because we will often hire a man with a good personality and average grades. Perhaps with metallurgists we do give more weight to book learning but this is not always the case. Of course we will not turn down an outstanding man without technical training.

With this background several people in headquarters screen men in approximately twenty colleges and universities; many of them are in the east, some in the south and midwest. We make personal interviews as far west as the Missouri School of Mines and, of course, have applications and hire men from the West Coast. If the people who visit a university for screening like a man they see, they may ask him to report directly to a particular division or ask him to go to New York to meet other people in the New York office including, probably, even the president. This is usually done either because the screener is not sure of his judgment or because he knows he likes the man but would like assistance in placing him in the right spot.

n

01

in

te

aı

ar

ca

in

ne

pr

in

fic

in

th

^{*} Assistant General Purchasing Agent, American Brake Shoe Co., New York.

ec-

to

18,

er

a

en

d-

vi-

W

ch

h-

ıl-

ts

n-

el

m

p

n

se

d

lo

1-

n

d

le

15

of

st

1-

rt

0

e

is

ıt

59

Often we will ask a man to look at two or three divisions, even though he may know the one he wants. The man and the division will make the final decision.

In addition to the men who are selected in this way, there are others who are hired directly by the divisions, either because they seek out a division for a job or are sent there by friends. We consider it a compliment to the company when its friends send men to us. Naturally the same is true when American Brake Shoe Co. people bring their friends to the company.

In addition to the college trainees, there are also available for positions of responsibility the men who move up from the ranks—molders who become foremen, foremen who become superintendents, etc. For example, in one division, in the five-year period immediately after the last war, 41 men in this category were promoted. A few of these had a college education but most did not. These men were clerks, molders, laborers, coremakers, watchmen, and cupola tenders.

I shall concentrate on the college graduates, not because they are most important but because this part of our training fits more appropriately into this discussion.

Training Program

Generally our training program for operating apprentices is planned for a two-year period. It's basis is learning by doing. Most of our foundry divisions have written flexible training schedules: 2 months on floor molding, 2 weeks in the shipping department, 1 month in the metallurgical laboratory, 1 month in the standards department, etc. These written schedules facilitate administration of the course, because there may be three or four apprentices in one plant at the same time. They keep a man from wandering about without direction and they allocate time more efficiently.

In each department the apprentice works with an expert for part of the time and then very often works on his own. For example, in floor molding he will work with a journeyman molder for 1 month, shoveling sand, operating a rammer, helping to draw patterns, set cores, etc. Then he may spend 1 month molding patterns by himself, doing his own risering and gating, fitting his cores, judging mold hardness, and possibly even pouring the iron himself, so he cannot blame anybody else if he makes a scrap casting. In other words he develops a strong back while keeping his mind alert at the same time. It is certainly not a sight-seeing junket.

However, because there is more to the foundry business than making sound castings, the operating apprentice also works in the office—he does cost estimating, production scheduling, shipping, and general office work. This usually occupies about 25 per cent of the course

A sales apprentice will spend considerable time in the foundry—usually 6 months to a year. Then he concentrates on the other sections which are particularly important to him—engineering, estimating, etc. Furthermore, his training will continue for some time after he leaves the plant. He may spend a year working with three or four salesmen out of one office, to learn the customers and his own products and also to observe different methods of selling the same item.

An accounting apprentice will work in the plant for 3 to 6 months, then will work in the bookkeeping section of the plant. After this he will usually work at several other plants, either to learn different systems of accounting procedure or different applications of the same system. He is always encouraged to learn what the numbers mean rather than to be merely satisfied with the correct total. We feel that the young accountant must realize how important original thinking is, how much he can help the superintendent by translating columns of figures into facts which can be easily understood and used to lower costs or even increase production.

Metallurgical Department Trainees

Apprentices in the metallurgical department may take either of two steps. The first is training in a division plant or plants. In this case his course may be very much like that of an operating apprentice except that he will have more time for technical sections and will always go to research headquarters of the company in Mahwah, N. J. The second step is a complete course at research headquarters. Here he will work in each section of the laboratory just as operating apprentices work in each department of the foundry. He will work under the section heads for several months, then will assist a metallurgist on one or more projects, and finally will be assigned projects of his own.

Almost all apprentice metallurgists take this second course.

Some sales apprentices also take a specialized training course in the metallurgical department. It runs from 9 months to 2 years, depending on the product. These men go out, fully equipped with a complete technical background. We believe that if these men use imagination they will have a big impact on the sales of their products.

When hired, an apprentice receives a monthly salary which is the amount we have to pay to get the man we want. Starting salaries for recent college graduates are at a general level just the same as wages for any workers in a free labor market. We normally meet the top half of this average rate. Sometimes we will go over the average to get a man we really want, either for his special training or special qualities which we think will be valuable to the company. However, we do not like to bid against another company because we feel that this makes the dollar sign assume an importance all out of proportion. We would rather lose a good man than have him come to us only because of financial considerations.

The apprentice receives a salary increase either at the end of 6 months or a year; that is division policy. Usually there has been at least one other increase by the time he starts his first regular job. The second and all subsequent increases in salary are based on merit. As a matter of fact the first one is, in a way, because if the apprentice does not get it, he knows he has not been doing satisfactory work and it gives us an opportunity to discuss his faults with him.

There are a number of features which are found in all our training courses which help to develop and broaden the apprentice and also make him feel that he belongs. We think he must understand the balance between human relationships and cold operating functions—that as he is trying to get more production from a jolt-rollover, he must think first of the men who are making it produce, that if he looks at a set of numbers and finds his costs increasing, he must remember that it is more probably lack of harmony between people than inefficient machines or equipment which causes it.

While he is in a plant, the apprentice has the run of the shop-he can ask all the questions he wants of anybody, and if his attitude is reasonable he will get an honest answer. The superintendent's door is always open, whether it be to answer more questions or give advice which was asked for. The superintendent is responsible for the progress of the apprentice and the flexibility of the course permits him, even forces him to make any necessary changes in the outlined course. In fact, if any one man, other than the apprentice himself, can make or break the course, it is the superintendent. If he does not believe in, has not been sold on the course, it cannot succeed. He has to be teacher, boss and friend. All of this work is in addition to his job of running a plant. Nothing but faith in the aim of this training can make him devote the time necessary for the course. From experience under one, I can testify that a good superintendent is a must.

Plant Visitation Program

Another feature of the course is the plant visit program, apprentices will usually inspect at least one other plant in the company and may inspect several outside foundries. Should there be an opportunity, they may work for several weeks in another plant.

After an operating apprentice has spent a year in the foundry, he may substitute for a sick foreman or perhaps fill in for a foreman on vacation. In emergencies he may help out in another plant in a super-

visory capacity.

There is still another part of the course which comes at the end of his training. The apprentice often spends a few days in N. Y. headquarters to learn how he can help the service section and how it can help him when he is in trouble in a plant. It is at this time that he can better realize the relationship of the individual division to the headquarters section. The latter is very much like the army staff headquarters which takes most of the burden for administration off the shoulders of the line outfit so that it can concentrate on fighting. Our headquarters takes over these jobs so that the divison can give all its efforts to production and sales.

The apprentice writes a report at the end of his time in each department. This enables the superintendent of the plant and his boss to find out what the man is learning or, more important, what he is not learning. It also helps this apprentice tie together all the facts he has picked up and makes him put things in the proper perspective.

Another feature which has been valuable for both apprentices and management is the occasional luncheon held in our board of directors room. Usually six apprentices from nearby plants will come to N. Y. and will have lunch with the chairman, president and two or three other men in responsible positions-a division president or the secretary or the accounting or research head. At these meetings no holds are barred and any question can be asked. Overall company policies are discussed; why American Brake Shoe Co. operates as it does, the historical background of the company, profits or sometimes even losses. Usually the main topic is in answer to an apprentice's question. These meetings give the trainee a better idea of the kind of company he is working for and the kind of people who guide the business. In turn, management apparently likes the close contact with the apprentices. The executives learn first hand what the younger men are thinking about and what is bothering them.

In connection with the latter, we have also had another type of meeting, also beneficial to both apprentices and management. From six to ten younger men from different divisions and different sections of a division meet in the head office. At the first meeting a company officer explains that they have been called together with the idea that their personal and collective experiences in the company will lead to criticism and suggestions about the weaknesses of the training courses and the manner in which company policies are carried out in the plants. When the officer leaves the group, it is understood that no holds are barred and an anonymous report will be written. There may be two or three other meetings with no officer or outsider present, so that there is a completely free interchange of ideas. After the final meeting, the group submits a summary of its discussion and recommendations. Then Mr. Given, company president, or someone in top-management goes over the report point by point, to explain where necessary and to try to read between the lines of any obscure suggestion.

These meetings bring together men at different points in their courses, from different divisions, and make them take a look at themselves and then make management take a look at itself. Sometimes there are very basic questions. The apprentices leave the discussion with a better knowledge of company policy and people and a more balanced picture of themselves and their bosses. Obviously management learns where its efforts must be aimed.

th

Transition Difficulties

One of the important facts which management learned from these meetings was the difficulty the apprentices had in the transition from their training course to their first job.

The operating apprentices usually go out to a plant as foreman in a department; sometimes it's as assistant foreman, occasionally as general foreman in a smaller foundry. Sales apprentices may either be given RI

im

th

ch.

six

V

nd

-a

ng

re

m-

oe

of

lly

es-

ea

he

11-

he

he

er-

ad

p-

er

of

t-

'n

rd

to

10

of-

ds

n

10

ly

d

i.

ie

d

g.

nt

d

e

re

e

y

25

re

11

C

g

ıt

a n a territory of their own or work with another salesman. The metallurgist will either remain at research headquarters or go to a division, to work either in the division metallurgists's office, as plant metallurgist, or sometimes in the sales or operating depart-

In all these cases the individual is thrown on his resources to a greater or lesser degree. We have learned that this so-called post-apprentice period is often a most difficult one for him. The transition from student in a plant to foreman on a conveyor line or to salesman in a tri-state area is a severe one. Until the men become settled in their regular jobs, they naturally have many hurdles to jump. Also, there is naturally less frequent contact with his headquarters and boss than there was. That he sometimes feels lost and forgotten is probably natural but also may occasionally be a fact. To overcome this problem, we have one man in each division who acts as a friendly counsellor, giving advice and listening to complaints, suggestions, etc. Perhaps he sees the apprentice and is able to have a talk with him three or four times a vear. We do not think this counselling plan is a substitute for having each apprentice handle his own relations with others in the company or for his efforts to develop his own progress in the most intelligent way he can, but we do think it a worthwhile supplement to the course.

Apprentice Training Program

We feel that actually nobody stops learning or training for a bigger job-at least he does not stop unless he has reached his limit. American Brake Shoe Co. tries to continue the development right on up until the top job is attained or until retirement is reached. The later development and individual growth is another subject, but its basis is the same as the present one. We believe in loading responsibility both as a means of testing a man and of helping him grow. We have flexibility and fluidity. One recent operating apprentice is selling brake shoes, another is a district works manager in his division, another, the writer, is in purchasing. All of us are still with American Brake Shoe Co.

Our training course succeeds to the extent it does because top management believes in it and works on it. It makes it relatively easy to find out, early in the game, how much a man has to offer. We will continue to change the course, to improve it again and again. We do not expect to do away with it.

DISCUSSION

Chairman: W. H. RUTEN, Polytechnic Institute of Brooklyn, Brooklyn, N. Y.

Co-Chairman: G. B. BARKER, University of Wisconsin, Madison, Wis.

F. C. CECH: 1 Do you give your apprentices any text material

to study while going through their apprenticeship.

Mr. Russell: We have no specially prepared text material in our apprenticeship courses. We recommend that the apprentices study certain A.F.S. publications.

MR. CECH: Please elaborate on the training you give the noncollege man.

Mr. Russell: There is very little I can add to what has been given in the paper. We have plant superintendents who contact schools in certain cities. As a general rule we hire the man who applies for employment and train him after that. Generally, a man will start as day laborer, and if he has potentialities and is interested, we keep moving him up. There is no limit to his advancement.

F. W. Shipley: 2 We send our foundry instructors or machine shop instructors to various high schools, even to distant cities, to put on demonstrations and describe in detail all the different apprentice courses we offer. We invite people into the plant to interest their sons in the various training courses throughout the plant.

Mr. Cech: You have an aggressive program.

R. W. Schroeder: 3 I am glad to hear Mr. Shipley make that statement, because it shows that industry does try, and if interested will go out and bring these boys in, rather than have them drift by the door. When industry is interested they will receive the help which is needed.

C. J. FREUND: 4 Is the superintendent of the various plants exclusively in charge of the training, or does he have a staff

member whose job it is to train these young men in the plant?

MR. RUSSELL: We do not have a Personnel Dept. as such except in two small divisions. The plant superintendent is in charge of apprentices in his plant. He reports to the operating vice-president as to the progress and potentialities of these apprentices. Top management feels very strongly the need for this training program.

Mr. Freund: How many apprentices do you have?

Mr. Russell: This year we are hiring 30 to 40 college graduates who will be put through a training program of 6 months to 2 years duration. Six is the maximum that any one superintendent will have.

F. G. SEFING: 5 What about the smaller companies? They will review what your company does and then say, "That is all right for a big company, but how can we do it in our organization?

MR. RUSSELL: I believe strongly in this program to think they cannot afford not to have such a training program. I think their difficulty is that they cannot afford a large number of apprentices and therefore their first choice has to be right. If we hire 40 graduates and guess wrong on two of them, we naturally do not lose quite as much as the man hiring one and who wastes 2 years finding out that that one man is not suitable. He would have to be very careful in hiring.

MR. SEFING: These small companies do not know how they can apply a training course when one superintendent or one technical man comprises their entire technical staff. Small shops are frequently overwhelmed in taking on and training one, two

or three engineering graduates. Mr. Russell: What would they want the man for?

Mr. Sefing: To build up their organization. They are leaning the whole organization on two men and they ought to have four.

Mr. Russell: The first thing you must do is to assure the first two men that you are not trying to get rid of them. Maybe there is just a small cupola and a small shakeout, but an apprentice can take a full 2-year course there and learn much. Our procedure can well be adopted by smaller companies. The top men would have to be sold on the training program. You would have to sell them on the idea that they must help teach this apprentice, that you think the organization is thin and you want additional personnel. You can take the man out to lunch, to safety meetings, do everything to make him feel he is a part of the company and has some future in it. He must feel management is interested in him.

MEMBER: Then the reasonable success of the course in any size company is the interest you take in the man rather than what you teach him?

Mr. Russell: Yes. We do not teach them much. They learn it themselves.

C. T. MAREK: 6 I have attended these meetings quite a few years and each time I heard considerable complaining about in-

Head, Patternmaking Div., Cleveland Trade School, Cleveland.

² Foundry Manager, Caterpillar Tractor Co., Peoria.

² Foundry Instructor, University of Illinois, Chicago.

Dean, College of Engr., University of Detroit, Detroit.
 International Nickel Co., New York.

⁶ Associate Prof., Purdue University, Lafayette, Ind.

dustry not doing their part and also about education not doing their part. I am wondering if it would be advisable to have someone take hold of this program and lead it. I wonder if you recognize the fact that these teachers in our schools are working just 9 or 10 months a year and must procure employment for the remaining 2 or 3 months. It is a wonderful opportunity for industry to say "Come on over to my place. I will give you a job and show you what the Foundry Industry is like." You could make foundrymen out of them pretty fast. They like that security, too.

I wonder if somebody else is going to take advantage of that before we do. Give them a job! Help them and they will in turn do for you what you have failed, in the past, to accomplish

by yourself.

I have had several boys come to me, after a year or two of teaching experience, asking me what they can do to get a foundry course started in their school. I advise them to go out to industry and get acquainted with the people in industry and see what they can do to help. So far, I have not heard of any success. I think I talked to one boy and he said he gave up the idea. In other words, there was not enough enthusiasm on industry's part. I think you could do more toward encouraging and inspiring this educational idea by giving these instructors, your teachers, jobs.

Also, you should have a co-ordinative member in charge of this educational program who keeps a record of those teachers and checks up with the schools to find out whether these instructors in vocational education were offered jobs, and where

they are placed.

MR. FREUND: The foundry superintendent who is asked to take a bright young man into his shop ought to be enthused to have the young man simply as a stimulus. You do not want second rate men around if you have any sense. I think that is an argument you can use on an alert man.

Mr. Russell: We have found that the men who do not want apprentices in the plant are generally the ones not doing the

best that can be done. The top notch superintendents want to have younger men around them.

S. C. Wasson: ⁷ The fact that we hear so often that small foundry organizations or certain foremen will not take in young graduates seems to be getting worn threadbare. I sometimes think we in top management use that as an excuse for following the line of least resistance. If we really want the young in our foundries, and are going to give the men attention and push them along, and can give them further education, we can overcome the resistance to their entering our shops by spending time talking with those in our organizations who seem to resist seeing new blood come in. Many times these older men, who look down on the young engineer, become the young fellow's best boosters once they start working together.

Prof. Barker asks what industry wants out of the schools. Prof. Barker's school is doing a fine job, but the main thing we must do in our high schools and at the college level is to present the Foundry Industry as an engineering, technical, scientific industry. Certainly there is hard work in it, but there is a lot beyond that. If we present the industry to these high school and college students in the right light, most of them will want to come with us. That is what we are trying to do at the F.E.F. schools in a way that has not been done before. Instead of presenting a pile of sand and a cloud of smoke, we are pre-

senting an engineering opportunity.

MR. SCHROEDER: I mention to my students at the University that a number of years ago a graduate of Dartmouth College was asked why he had chosen the foundry field for his life work. He answered, "Because there is so little competition for top positions." I bring out to the engineering students that the foundry is a good place to work in if you are willing to work hard for a few years. I am glad to report that a number of our students are going into the Foundry Industry this year. They like the challenge. But we still must sell those on the high school level.

⁷ Manager, National Malleable & Steel Castings Co., Cicero, Ill.

HEAT FLOW IN MOIST SAND*

By

Victor Paschkis**

1. Introduction

STUDIES ON SOLIDIFICATION OF CASTINGS made on the Heat and Mass Flow Analyzer are based on the knowledge of the thermal properties of the casting as well as the mold. In former reports¹ the influence of thermal properties of steel as well as of the shape of the casting on the solidification rates were examined, but the differences in the sand mold have received little attention. In the report for the year 1949-50,² the influence of varying conductivity of dry sand was examined and an appreciable influence of the conductivity on solidification rates was found.

In foundry practice, however, moist sand is frequently used. This applies not only when casting in green sand, but also when dry sand is being used. After drying, the sand may absorb moisture from the air. It should be remembered that actual foundry sands are composed of sand, clay and binders, and that it is a simplification to talk only of "sand."

For dry sand it is customary to speak of "thermal conductivity," defining same as the rate of heat flow per unit area across a thickness of unit length under the influence of a temperature drop of one degree Fahrenheit. This definition is used even if the heat flow actually occurs in part by conduction from sand particle to sand particle and in part by radiation. Therefore it would be more appropriate to speak of "apparent conductivity" of dry sand. This apparent conductivity is a function of temperature but otherwise constant.

Heat transfer in moist sand is more complex. In moist sand heat transfer occurs in part by conduction from sand particle to sand particle and in part by conduction across the water which fills the interstices of the sand. However, as the moist sand is being heated the moisture migrates under the influence of change of pressure; thus an additional element of heat transfer by convection is introduced. Moreover, practical moisture content in sands is always so small that not all interstices are filled by water. Therefore part of the heat transfer still occurs according to the same pattern as that prevailing in dry sand, namely, across air spaces.

The work covered in the present report started at the end of the business year 1948-49 but had not advanced enough to submit a report. It was completed in the business year of 1949-50 but should still be considered only as a first step towards exploration of these conditions.

The concept of combined heat and moisture migration is so complex and the experimental technique so involved that this report deals entirely with a fictitious assumption, namely, that the "hot surface" of the mold is suddenly raised in temperature from 70 F to a high value and held at that value for a considerable length of time. This hot surface represents the casting-mold interface and actually experiences varying temperatures. In later experiments this condition can be simulated.

It should be noted that even if the concept of constant hot surface temperature does not hold, conditions in actual casting are reasonably similar. The interface between casting and mold remains for a considerable length of time at nearly constant temperature.

II. Program

As mentioned in the introduction, the program was carried out in two steps: the first in 1948-49 was a preliminary experiment based on the more simple concept that conduction in the sand was the same as in dry sand, and that merely convection of moisture was superimposed on this conductivity.

In the winter 1949-50 a paper by Tanasawa³ came to the attention of the author, which describes a technique of studying apparent thermal conductivity of moist sand. Based on this paper the more complete concept indicated in the introduction was developed and was made part of the main body of the experi-

RY

ing nes ing our ush verme

w's

we to

scie is igh vill

re-

ity

ege

rk.

isi-

ra

nts

he

el.

^{*}This is a report on a Research Project sponsored by the A.F.S. Heat Transfer Committee at Columbia University. Members of this Committee are as follows: H. A. Schwartz, Chairman, J. B. Caine, V. Paschkis, A. E. Schuh, C. E. Sims, H. F. Taylor and E. C. Troy.

^{**}Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York City.

ments. However, before accepting this concept the figures of Tanasawa were compared with the considerations of Austin.⁴ This comparison is contained in the Appendix together with some more theoretical considerations which are essential for the continuation of the studies but not of immediate interest to the practical foundryman.

In these main experiments, the influence of the magnitude of the assumed temperature rise of the interface was examined.

III. Properties

The preliminary experiments were made with the sand as assumed in previous work of the laboratory.¹ The conductivity of the sand was 0.9 Btu/ft,hr,F and was assumed not to change with moisture content. The moisture content merely changed the apparent specific heat. As the sand dried out, the heat of evaporation was extracted in one section and introduced in a later one in the form of heat of condensation.

The specific heat of the dry sand was assumed to be 0.28 Btu/lb, F. Density was assumed to be 93.6 lb/cu ft.

In order to be able to compare performance of moist sand with that of dry sand, it was decided to use in the main experiment the sand designated as 20–30 in the paper by Lucks and co-workers.⁵ That sand was also investigated and reported upon in a paper "Influence of Dry Sand Conductivity on the Rate of Freezing of Steel Slabs."²

This sand has a dry density of 108 lb/cu ft.

The thermal conductivity is shown in the above quoted paper by Lucks. By extrapolating beyond his values, the following conductivities were found and were used in the present work (see Table 1).

TABLE 1—THERMAL CONDUCTIVITY OF DRY SAND 20-30 EXTRAPOLATED AFTER LUCKS AND CO-WORKERS

Temperatures, F	Conductivities
	Btu/ft. hr.F
2200	0.704
2000	0.600
1750	0.480
1500	0.417
1250	0.350
1000	0.300
750	0.250
600	0.215
500	0.191
400	0.169
300	0.149
212	0.133

Physical considerations showed that probably the influence of temperature on the conductivity below 212 F is negligible compared with that of moisture content. The computation of the conductivity depending on moisture content is given in the Appendix. At present a sand with a moisture content of 4 percent, referred to the dry sand weight has been investigated; the conductivity values for this moisture content is shown in Table 2. In this table the first column shows the percentage of moisture content expressed in parts of the original moisture content. Column 2

contains the conductivities for 4 percent original mois-

Regarding the designation of moisture contents reference is again made to the Appendix. Moisture contents can be expressed in percentage weight of the dry sand (M_1) or percentage weight of the moist sand (M_2) or finally as percentage of the volume taken up by water. The standard designation of moisture in foundry work is in percentage of the moist sand. Inasmuch as the moisture changes during the drying out it seemed more practical to show both the percentage referring to moist sand (M_2) and in addition the percentage referring to the dry sand (M_1) .

For the specific heat of sand a value of 0.28 Btu/lb, F has been used in previous publications of this laboratory.¹ The specific heat is quite a marked function of temperature and therefore a variation of specific heat with temperature was introduced in the present experiments according to tables in Schack,⁷ Groebers and Norton.⁹ Interpolation was made between the following values:

70 F, 0.19 Btu/lb, F 1100 F, 0.23 Btu/lb, F 2200 F, 0.28 Btu/lb, F

The specific heat of water is of course 1 Btu/lb, F.

IV. Experimental Results

Preliminary Tests—The charts shown hereafter show the behavior of the sand slab by various indices.

Figures 1 and 2 show temperature (vertical axis) – time (horizontal axis) relationships for points at different distances from the surface. The two curves, la and 2a in Fig. 2 show the time required to reach the temperature of 212 F and for complete evaporation for various distances from the hot surface. Figure 2 shows two time-space curves indicating the time required to reach the temperature of 212 F and to complete evaporation for various distances from the hot surface.

In Fig. 3 the temperature gradients (space-temperature curves) are shown for consecutive time elements.

The time-temperature curves all have a common appearance: a more or less rapid rise from 70 F to 212 F, a stop at 212 F, lasting until all water from this section has evaporated; and then a new temperature increase at a steadily decreasing rate.

The space-temperature curves indicate, that at any one time there is a zone in the sand which is remaining constant at 212 F.

At greater distances from the interface the times required for drying become increasingly longer. At

Table 2-Thermal Conductivities for Moisture Content $M_1=4\%$, $M_2=3.85$ Temperature Range 70-212F

Moisture Content in % of Original Moisture Content	Thermal Conductive (Btu/ft.hr.F)		
100	0.597		
	0.531		
50	0.443		
25	0.223		
0	0.133		

80

-30-

Fig

the a d

sho cast liqu NI

is.

re he ist

ne

ist he

he

li-

b,

on fic

nt rs

16

f-

ie

n

e

n.

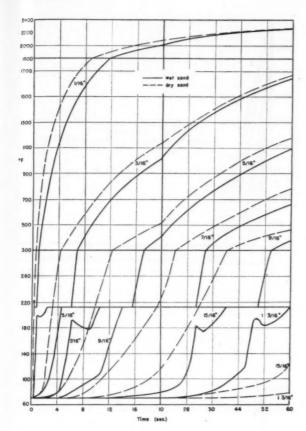


Fig. 1—Time-Temperature Curves. Hot Surface Temperature 2600 F. Distances measured from hot surface.

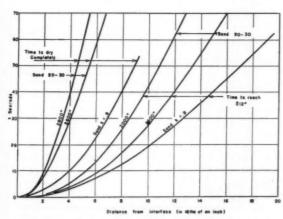


Fig. 2-Time-Space Curves for Heating Moist Sand. Hot Surface Temperature 2600 F.

1/4 in. from the interface the time between reaching the boiling point and complete dryness is 6.8 sec; at a distance of 1/2 in. the equivalent figure is 30 sec, etc. It is also important to note, that the time to dry to

a depth of ½ in. from the hot surface is 40 sec. It is interesting to compare this time with the value shown in Fig. 2 and 3½; a steel slab 4 in. thick cast in sand requires from 500 to 600 sec to reach the liquidus at the center, when poured at 2845 F, and

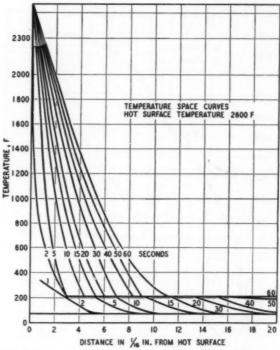


Fig. 3—Temperature Space Curves. Hot Surface Temperature 2600 F. Numbers on curves indicate time after application of heat to hot surface. Note time at constant temperature of 212 F.

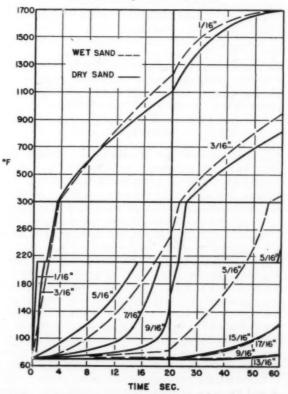


Fig. 4—Time-Temperature Curves for Sand 20-30. 2000 F Interface Temp. Distances measured from hot surface.

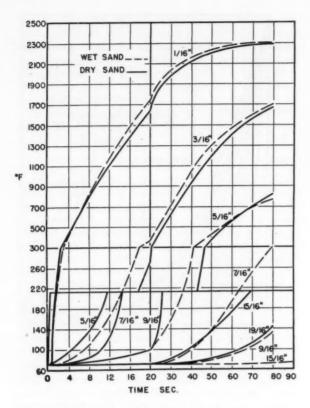


Fig. 5—Time-Temperature Curves for Sand 20-30. 2600 F Interface Temp. Distances measured from hot surface.

750 to 800 sec when poured at 295 F. The vapor migration in the sand will therefore have an influence on the apparent conductivity in the early times, but probably not towards the end of solidification.

Main Tests

(a) Temperature-Time Relationships - Results for the tests at 2000 and 2600 F are shown in Fig. 4 and 5 respectively. The arrangement of the charts is the same as Fig. 2. The latter chart and Fig. 5 hold for the same temperature but different sands; Fig. 4 and 5 for the same sand but different temperatures. The apparent conductivity of the sand 20 to 30 (Fig. 5) is lower than that of the sand used for Fig. 2. Hence heating up is slower. Even with the moisture influencing the apparent conductivity (Fig. 5) the temperature curves for moist (solid line) and dry (broken line) sand are not very different. The time during which each part remains at 212 F is longer under the conditions of Fig. 5. For example the point 3/16 in. is 16 sec at 212, whereas Fig. 2 shows a time of only 4.4 sec. (The hump at 200 F is caused by approximations in electric circuit, eliminated in later tests of Fig. 5).

Lowering the temperature of the hot surface (Fig. 4 and 5) has a marked influence, even more noticeable at larger distances from the hot surface than near that surface. Approximately, for any given time after the start of an experiment the ratio $(t-t_1)/(t_s-t_1)$

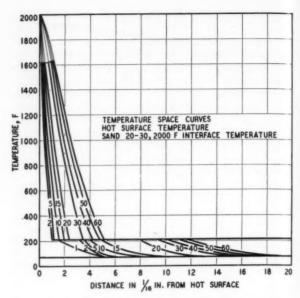


Fig. 6-Temperature Space Curves; Hot Surface Temperature; Sand 20-30, 2000 F Interface Temp. Numbers on curves indicate time after application of heat to hot surface. Note time at constant temperature of 212 F.

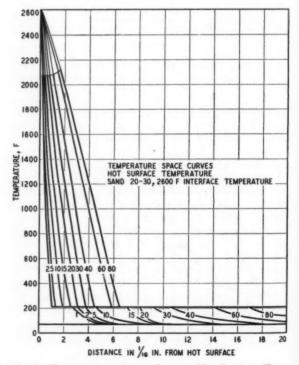


Fig. 7—Temperature Space Curves; Hot Surface Temperature; Sand 20-30, 2600 F Interface Temp. Numbers on curves indicate time after application of heat to hot surface. Note time at constant temperature of 212 F.

should be the same, independent of the constant surface temperature, t_s , if t_s has no unproportional influence. Here t denotes the temperature at a given

Fig

ten sho of sho

Bhu /si

250

200

100,

50

Fig San mmat

ıí

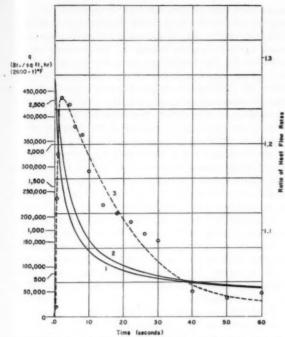


Fig. 8—Rate of Heat Flow From Hot Surface vs Time; Hot Surface Temp. 2600 F. Curves 1 and 2 use left scale, Curve 3 use right scale. Curves 1 and 2 show temperature ½6 in. from hot surface: second left scale shows following values "2600—t". Curve 3 shows ratio of rate of heat flow from hot surface wet/dry. It also shows ratio of temperatures wet/dry expressed in 2600—t.

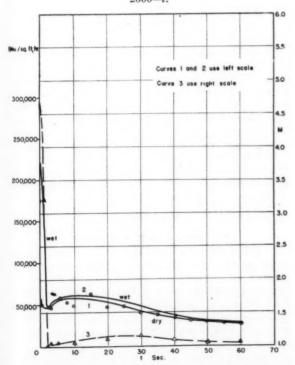


Fig. 9-Rate of Heat Flow From Hot Surface vs. Time; Sand 20-30. 2000 F Interface Temp. Ratio M of rates of Heat Flow are Plotted on Right Scale.

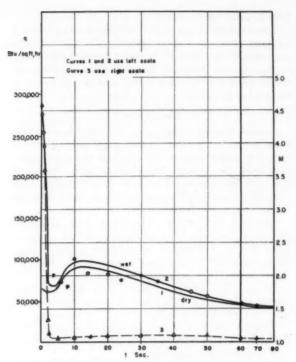


Fig. 10—Rate of Heat Flow From Hot Surface vs Time; Sand 20-30. 2600 F Interface Temp. Ratio M of Rates of Heat Flow are Plotted on Right Scale.

time and t_i the initial temperature of the sand. For example, 60 sec after "casting" the point $\frac{1}{16}$ in. under the surface (wet sand) has reached a temperature of 2270 F (2600 F "pouring temperature") or of 1690 F (2000 F "pouring temperature"). If the influence of the pouring temperature is proportional, then the ratio (2270-70)/(2600-70)=0.87 should be the same as the ratio (1690-70)/(2000-70)=0.84. The temperature at $\frac{1}{16}$ in. is lower than expected. For a point $\frac{3}{16}$ in. below the surface the two values of ratio are (both for 60 sec) (1540-70)/(2600-70)=0.58 and (940-70)/(2000-70)=0.45.

At $\frac{1}{16}$ in. the deviation was only 0.84/0.87 = 0.97, whereas at $\frac{3}{16}$ in. it deteriorated to 0.45/0.58 = 0.78. (b) Temperature-Space Relationships—Corresponding to Fig. 3 two graphs are shown for sand 20–30: Fig. 6 for 2000 F and Fig. 7 for 2600 F.

They illustrate in a different way the characteristics, discussed in (a). The lower conductivity (sand 20–30 compared with the sand of the "preliminary experiments") results in a steeper temperature gradient (Fig. 3 vs Fig. 7). Similarly, the higher interface temperature (Fig. 7 vs Fig. 6) results in a steeper gradient.

(c) Time-Space Relationships—These are shown in Fig. 2, curves 1b and 1c for the time to dry completely and 2b and 2c for the time to reach 212 F. Curves 1b and 2b refer to a hot surface temperature of 2600 F and 1c and 2c to a hot surface temperature 2600 F.

Both, the times to reach 212 F, and to dry completely for any given distance from the interface increase in the following order: lowest time for the sand used in the preliminary experiment; sand 20-30,

2600 F; sand 20-30, 2000 F.

The horizontal spread between the curves 1 and 2 carrying the same letters indicate the layers of the sand which are held at constant temperature. The vertical distance between the same pairs of curves indicate the length of time during which each point in the sand is at 212 F.

Conclusions

General Considerations—The foundryman is interested in the temperatures in the sand only to the extent that such temperatures influence the heat extraction from the casting. Properties or temperatures in the sand are significant only in so far as they have bearing on the solidification, either at early times or in the overall solidification rate.

Although the present experiments were all based on the assumption of a constant interface temperature, that is, a constant temperature of the casting-mold interface, certain conclusions on the heat extraction from an actual casting can be drawn. In order to study the heat extraction from the casting, curves and charts Fig. 8, 9 and 10 are presented on all of which the abscissa scale is "time" in seconds.

Figure 8 refers to the preliminary experiment, Fig. 9 and 10 to the sand 20–30 for 2000 and 2600 F inter-

face temperatures respectively.

The charts have an ordinate scale of Btu/sq ft, hr which is a measure for the rate of heat extraction from the casting. It is important to find out what influence the moisture content has on the heat extraction and therefore for each of the three cases a further curve was added showing the heat extraction for the same sand but assuming that the latter is entirely dry from the beginning of the test run.

Finally each of the charts shows a curve indicating the ratio of the rates of heat extraction of the moist

sand/dry sand.

In the case of the preliminary test, Fig. 8, where the conductivity of the sand was not varied with temperature it was possible to apply another ordinate scale to the same curve. This ordinate scale indicates the difference between the initial temperature and the temperature at any time at a point ½6 in. away from the hot surface.

Discussion of the Curves—In all instances the rate of heat extraction is originally very high and quickly decreases to a much lower value. Theoretically this rate of heat flow is much larger than that observed. The so-called "lumping technique" used in the computer experiments tends at very early times to decrease the rate of heat flow.

At very early times the rate of heat extraction of the moist sand is much larger than that of the dry sand. This observation holds for all three curves. Most markedlý it is true for the sand 20–30, 2000 F temperature; less marked for the same sand at 2600 F temperature and least marked for the sand used in the preliminary experiment.

As time goes on, the rate of heat extraction of the dry sand and the moist sand becomes closer and

closer.

The maximum ratio of heat extraction rates, moist/ dry is fairly low in case of the preliminary experiments and does not exceed 1.26. It is much larger in case of the sand 20-30, 2600 F where it reaches a value of 4.6. In case of the low temperature the maximum ratio is approximately 5. In the case of the sand 20-30 the maximum ratio is obtained at the start of the experiment and drops down rapidly thereafter In the case of the preliminary experiment the rate increases rapidly and reaches maximum a few seconds after the start of the experiment and thereafter drops slower than the case of the sand 20-30. It should be understood that the difference in the shape of the curves is not a characteristic of the sand but rather of the assumptions underlying the experiments as explained in Section II. Thus the shape and nature of the curves as shown in Fig. 9 and 10 are considered more reliable than those of Fig. 8.

The moisture content thus exerts a stronger chilling action if the pouring temperature is lower. In all instances the chilling action is only of brief duration. It may contribute to the rapid formation of a thin layer of frozen metal but probably does not contribute to a change of total solidification time of the

center of a casting.

Practical Considerations—Casting in green sand is more commonly practiced in gray iron foundries than in steel foundries. Although the experiments reported above were made with properties of steel the results are more than indicative for gray iron practice. The fact that lower pouring temperatures are used in iron foundries (as compared with steel foundries) is the reason why moisture content is of more importance in iron foundries than in steel foundries. The iron casting will undergo a severe chilling effect for a brief time if moist sand is used.

Further Work

tł

ir

PI

CO

Further work will explore the influence of various degrees of moisture content and later on it is necessary to drop the assumption of constant interface temperature. Actually the interface temperature will drop quite rapidly from the initial temperature to a much lower value which then is held for a relatively longer period almost constant. The rapid drop of temperature during the early stages of solidification will probably increase the influence of the moisture.

In future years attempts should be made to see to what extent the influence of the moisture content is offset in some parts of the casting by the drying effect of the flow of liquid metal into the mold.

APPENDIX

I. Some Physical Concepts

1. When the surface of a slab of moist sand is suddenly raised to a temperature well above the boiling point at normal pressure, then the layers next to the hot surface will dry out rapidly. Provided the steam thus formed cannot escape through the hot surface, the steam will be pushed out towards the cold surface. It is well to remember that as long as the entire moisture of a layer is not evaporated, the temperature of that layer cannot exceed the boiling temperature.

The steam being pushed towards the cold surface starts to condense as soon as it reaches layers of sand which are below the boiling temperature. AND

in

s a

axi-

the

the

ere-

the

few

ere-

11

ape

but

nts

ure

red

ill-

all

011.

nin

tri-

the

is

an

ted

ilts

he

on the

ice

on

ief

ous

ary

er-

op

ich

ger

ra-

ob-

is

ect

ıd-

ng

he

m

ce

II-

ire

ire

ur-

of

3. Layers of sand which increase in temperature, but yet are below the boiling point, also lose liquid. This occurs by two mechanisms. According to the elevated temperature the partial pressure of the waterair system existing in the interstices of the sand changes and under the influence of pressure the vapors migrate. This process is well understood, qualitatively and quantitatively. In addition, moisture will flow by capillary action through the sand. Very little is known about this capillary flow.

4. The present report covers only the phase of heat transfer by steam. The migration of moisture in parts of the sand which are below the boiling point, is not considered. This greatly simplifies the investigation, and is, at least as first step, a reasonable approxima-

5. The analysis makes wide use of the electric analogy method, which in general may be considered to be known.10 Special adaptations were necessary to make it suitable for the present studies, the adaptations being described in Appendix V. Here it is merely necessary to explain that this technique includes a "lumping procedure." The slab (sand mold) is thought of as divided into several sections; in the center of each section the specific heat, moisture, etc., are concentrated. Thus if e.g. in the report one reads "the first section remains at constant temperature . . ., it should be understood that this implies an approximation. Actually only infinitely thin sections should be considered; and the description of occurences in sections of finite thickness involves an inaccuracy.

II. Definitions of Moisture Content

The standard definition of moisture content of sand in foundry work is by expressing the water content in per cent of the weight, moist sand.6 Inasmuch as the weight of the moist sand changes during drying out, this definition is not convenient for studying any processes in which drying takes place, such as the one covered by this report. For this purpose it is more convenient to compare the weight of the water with that of the dry sand, obtained after all moisture has been driven out.

Finally for certain calculations and considerations covered in Section IV of the Appendix, it is helpful to consider the moisture content on a volume basis, indicating what per cent of the total volume is filled

Inasmuch as all three definitions are used in the text, it is important to know the relationships between the three operations.

In the following comparisons, notations are used as follows:

M₁ = moisture content in % (by weight), referring to dry sand as standard

M₂ = moisture content in % (by weight), referring to moist sand as standard

M₃ = moisture content in % (by volume)

p_D = density of the dry sand in lb/cu ft p_M = density of moist sand in lb/cu ft

pw = amount of water in lb/cu ft of mixture

Relationships between M₁ and M₂-In moist sand of PD density and M1 moisture content the total water content may be found from

 $p_W = (M_1 \times p_D)/100 \text{ lb/cu ft}$ (1)

The total weight of 1 cu ft of the moist sand, p_M, can be found by adding the weight of the dry sand cu ft (pD) and the water content pw. The volume of the moist sand does not change by adding water.

$$p_{M} = p_{D} + p_{W} \qquad (2)$$

The moisture content M2 can be found by comparing the water content pw with the weight of the moist sand pm. Hence

$$M_2 = 100 p_W/p_M = 100 p_W/p_D + p_W = 100 M_1/(100 + M_1)$$
 (3)

Relationships between M₁ and M₃-The definition of M₃ means that 1/cu ft of sand contains M₃/100 cu ft of water. 1 cu ft of water weighs 62.4 lb and therefore 1 cu ft of sand contains water weighing 0.624 M_3 lb.

According to the definition of M1, I cu ft of sand contains $p_D \times M_1/100$ lb of water. The quantity of water calculated by either method must be the same,

$$p_D \times M_1/100 = 0.624 M_3 \text{ or} M_3 = p_D \times M_1/62.4$$
 (4)

III. Some Numerical Considerations

The moisture content of foundry sands (using definition M₁) varies within wide limits, possibly from 0 to 10 per cent.

The specific heat of (dry) foundry sands is in the order of magnitude of 0.2 to 0.3 Btu/lb,F. The specific heat of water is 1.0 Btu/lb,F and heat of vaporization at atmospheric pressure is 970.3 Btu/lb. Consider the amount of heat required to heat the sand and the water to from 70 to 212 F, and to evaporate the water. Note, that in no element of the mold can the temperature rise above 212 F, before all the heat of vaporization has been supplied. Note also, that the heat introduced into the mold is being extracted from the casting. Table 3 is prepared in way of example and is based on a specific heat of the (dry) sand of 0.28 Btu/lb,F and a density of 93.6 lb/cu ft.

TABLE 3-HEAT IN BTU/CU FT TO BE SUPPLIED

Moisture Content M ₁ , %	To Sand	To Water in Sand	Subtotal for the section to reach 212 F	Vapori- zation	Total, before section of mold can increase beyond 212 F
0	3720		3720		3720
4	3720	531.6	4252	3640	7892
8	3720	1063.2	4753	7270	12053
10	3720	1329.0	5049	9090	14139

If heat is introduced to a section containing a certain amount of moisture, and this moisture is evaporated, then the steam must escape as rapidly as it is formed, or the pressure must increase. Now, say that the moisture content $M_1 = 4\%$. Then in a layer $\frac{1}{8}$ in. thick and having 1 sq ft cross-section, the moisture content is $0.04 \times 93.6 \times (1/8) \times (1/12) = 0.039$ lb. The required heat of evaporation is $0.039 \times 970.3 =$ 37.8 Btu. In the layer next to the casting-mold interface one might expect a temperature difference not to exceed 2000 F. Heat would flow to the center of the section at the rate of

$$\frac{0.9 \times 2000}{1/(8 \times 2) \times \frac{1}{1_2}} = 345600 \text{ Btu/sq ft,hr}$$

= 96 Btu/sq ft,sec.

Hence it would take only 37.8/96 = 0.395 sec, to supply the heat of evaporation. If the steam is to be withdrawn during this period, a quantity of 0.039 lb has to be forced out. The specific volume of steam at atmospheric pressure is 26.8 cu ft/lb. Therefore a total of $0.039 \times 26.8 = 1.04$ cu ft have to leave the section. The lowest permeability used in foundry sands is $6 \text{ cm}^4/(\text{g} \times \text{min}) = 5.25 \times 10^{-5} \text{ft}^4/\text{lb,sec}$.

Now one can examine what pressure would be required to force the 1.04 cu ft of steam out of the 1/8 in, thick section in 0.395 sec.

Pressure =
$$\frac{1.04}{0.395 \times 5.25 \times 10^{-5} \times 96 \times 1}$$

= 522 lb/sq ft = 3.63 lb/sq in.

This pressure is small enough to warrant the assumption that the steam is drained off as fast as it is formed. In other words that there is no time lag in transferring the steam from one section to the other.

This assumption becomes also more valid because the required pressure becomes smaller with greater distance from the interface, because the rate of heat inflow is smaller at such a greater distance.

The increase in pressure is accompanied by an increase in temperature of approximately nine degrees. This increase is so small as to be negligible in first approximation.

IV. Tanasawa's Values3 and Austin's Study4

Tanasawa presents in his paper a set of curves for thermal conductivity, specific heat and diffusivity obtained in the diffusivity tests. These curves were redrawn using great care to read the small graphs and are included as Fig. 11 in this report.

Tanasawa's values are apparently the only ones known showing the conductivity as function of moisture. It should be noted that he shows a marked increase of conductivity with relatively small amounts of moisture and a gradual tapering off of the moisture influence at higher moisture contents. Before accepting this relationship for the present investigations, it was thought worthwhile to compare the findings of Tanasawa with the equations presented by Austin.

The first step was to compare the conductivities for the completely dry sand. For this condition, Austin's equation 10 can be used. Austin defines the porosity P as the ratio of the total volume of pores to the volume of the over-all body under investigation. Assuming that the pores are filled with air, as would be the case for completely dry sand, the porosity P can be calculated from equation (5).

$$P = 1 - (p_D/p_s) \tag{5}$$

In this equation p_D is as before the density of the dry sand and p_8 the density of the solid sand particles if they would form a continuous body with no air interstices.

Austin bases his investigations on a porous brick but the findings should hold equally well for sand. Calling the conductivity of the sand k_B ("brick") and that of the solid without interstices k_B (corresponding to p_B) and the conductivity of air, k_A , then Austin's equation reads as follows:

$$k_B = k_A \frac{(1 - P)^{2/8}}{(1 - P)^{2/3} - (1 - P)}$$
 (6)

To evaluate this equation for Tanasawa's values, k_A and P have to be known. For k_A , one finds for 45 C by interpolation from Jakob¹¹ page 69, the value 0.0158 Btu/ft,hr,F.

In order to calculate P, the values of pD and ps must be known. pp is given in Tanasawa as 1100 kg/m3. The value of ps is not indicated in Tanasawa's paper. Jakob11 page 94 shows as limits for this value the figures 2400 to 3100 kg/m3. Using these two limits one obtains for P, 0.54 to 0.64. Evaluating equation (6) for these two values one finds as limits for k_B 0.068 to 0.057 Btu/ft,hr,F or 0.101 to 0.085 kcal/m. hr,C. The smaller value is obtained with a larger porosity. Tanasawa shows for dry sand as may be seen from Fig. 11, a value of 0.12 kcal/m,hr,C. The check seems close and would indicate that a value near the lower porosity is correct. The lowest value of porosity (P = 0.54) however cannot hold true, because Tanasawa shows moisture contents beyond 0.56 and the porosity of course cannot be any smaller than this value.

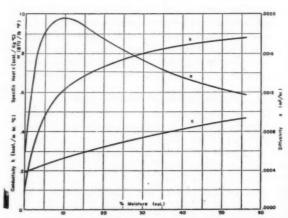


Fig. 11—Diffusivity, Conductivity and Specific Heat for Moist Sand (After Tanasawa).

Next it was attempted to check the upper end of Tanasawa's curve which holds supposedly for the condition that the sand is saturated with water and cannot absorb more water. As limiting case one should then assume that all interstices, are filled with water.

For this case two different physical concepts are possible, both introduced by Austin. One would be, that the water is continuous and is interrupted by sand particles. The other implies that the solid is continuous and encloses water in its cavities. The first concept is covered by Austin's equation (7), the latter by his equation (8).

Using the notations as before these equations are as follows:

$$k_{\rm B}/k_{\rm S} = \frac{{
m P}^{2/3} + k_{\rm S}/k_{\rm A} (1 - {
m P}^{2/3})}{{
m P}^{2/3} - {
m P} + k_{\rm S}/k_{\rm A} (1 - {
m P}^{2/3} + {
m P})}$$
 (7) for the concept of continuous water, and

ND

45

00

sa-

iis

VO

ia-

or

m,

er

be

he

ue

ue

te,

nd

er

or

of

he

nd

ld

er.

re

e.

by

st

lt-

re

7)

$$\begin{array}{c} k_{B} = \\ k_{B}/k_{A} \; (1-P)^{\;2/3} + 1 - (1-P)^{\;2/3} \\ k_{A} \overline{k_{B}/k_{A}[\; (1-P)^{\;2/3} - (1-P)\;] + [1 + (1-P) - (1-P)^{\;2/3}]} \\ \end{array}$$

for the concept of continuous solid.

In order to solve these equations the conductivity k_s must also be known. Moreover, if it is assumed that all interstices are filled with water the conductivity of water must be introduced for " k_A ." For the conductivity of solid, Schack⁷ page 345, gives the limiting values 4.2 to 5.9 Btu/ft,hr,F. For water, one may interpolate from Jacob¹¹ page 78, for 45 C, the value $k_A = 0.366$ Btu/ft,hr,F.

Using again the two limits for porosity one obtains a total of four different values for the conductivity of the completely moist sand, two each for each of the two equations (7) and (8). The results from equation (7) and (8) do not differ greatly and therefore only those for k₈ of 4.2 are shown here. The values for the two different porosities are 1.59 and 1.37 kcal/m,hr,C, the higher value belonging to the lower porosity. In comparison Tanasawa shows for a moisture content of 56.3 per cent a conductivity of only 0.88 kcal/m,hr,C. The check is obviously much less close than that for dry sand but still gives the same order of magnitude.

For intermediate values of moisture content no comparison of Austin's and Tanasawa values is possible. An attempt failed to compute the conductivities for intermediate moisture contents based on the conductivities for completely dry and completely moist sand by a law of mixture sand as shown by Tanasawa.

However in view of the good check for dry sand and the fair check for completely moist sand, it was decided to accept Tanasawa's function of increase of conductivity with moisture content.

The reasoning was as follows: the sand 20–30 has a porosity of P=0.35 whereas Tanasawa used the sand with a porosity of 0.575.

Using Austin's equations for dry sand and for sand completely saturated with water, both for the sand 20-30 and for Tanasawa's sand, one can compare the conductivities of moist and dry sand.

For the sand 20-30 one obtains:

for moist sand—1.51 Btu/ft,hr,F and for dry sand—(at 45 C)—0.114 Btu/ft,hr,F the ratio of the two figures is: 1.51/0.114=13.2

For Tanasawa's sand, one obtains:

for moist sand—1.07 Btu/ft,hr,F for dry sand—0.068 Btu/ft,hr,F

The ratio of the two figures is: 1.07/0.068=15.8

The two ratios are close enough together that lacking better information it seemed better justified to vary the conductivity according to the moisture content by volume in the same way as Tanasawa showed. With Tanasawa's figures and calling the conductivity of dry sand and (as reference) "one," one obtains a curve as shown in Fig. 12. The conductivity values for sand with some moisture were found by multiplying the conductivity of dry sand (.133) with the appropriate ordinate values from Fig. 12.

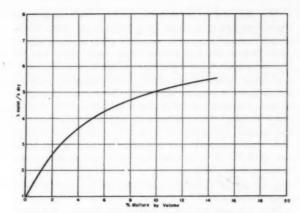


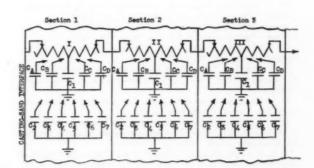
Fig. 12-Ratio of Conductivities of Moist vs Dry Sand (After Tanasawa) Plotted Against Moisture.

V. Experimental Technique

The setting up of electric circuits to compute heat flow problems has been published previously and will not be described in detail here. The technique is described in several publications by this laboratory. However, certain modifications had to be made to represent the migration of steam.

In Fig. 13, a wiring diagram representing three of the several "sections" is shown. Each "section" I, II, III, represents a layer of finite thickness of sand. The resistances on either side of the capacitances represent the thermal resistance to heat flow in sand comprising all modes of heat transfer in the sand: conduction from sand particle to sand particle, conduction in the water and convection and radiation through the air. In other words the resistances are based on the apparent conductivity of the sand.

The capacitances represent the specific heat of the sand and of the water condensed in each section. The capacitances $C_1 - C_7$ represents the specific heat of the sand and the capacitance $C_n - C_d$ that of the water.



C_A to C_D are moisture capacitors which are originally on the sections indicated but which move as the sand heats up.
C₁ is the dry sand capacitance at temperatures below 212 F.
C₂ to C₇ are capacitors added to C₁ to represent the increase of specific heat with increasing temperature. They are precharged to voltages representing the temperatures at which they are added.

Fig. 13-Schematic wiring diagram for three sections.

The migration of moisture from each section is represented in four steps, each step taking place when 25 per cent of the total moisture content has evaporated. When the evaporation has taken place one of the condensers Ca - Cd is transferred to a section further away from the interface. Simultaneously, corresponding to a change in conductivity with changing moisture content, the resistance is changed by operating a switch. The capacitance which is taken out of the section is transferred to a next section unless that next section has already reached the boiling temperature 212 F. If it has reached that temperature, the water cannot condense but is transferred to a following section; the latter is selected as the first section being below 212 F.

When a section has reached the boiling temperature, the resistance is changed according to the change in conductivity as shown in Table 1. At the same time the capacitances C2 - C7 are inserted to

represent the change in specific heat.

The operation described so far does not include the heat of evaporation. The heat of evaporation acts as a heat sink. As soon as a section has reached the boiling temperature of 212 F current is withdrawn according to the heat of evaporation. This current is integrated in a current integrator, a newly built device, and when one-quarter of the heat of evaporation has been withdrawn one of the capacitances Ca - Cd is transferred to the later section as described above. As the steam transfers to a later section the content of the later increases, and a heat source becomes active which is of equal strength to the heat sink in the previous section. Current is fed to this later section at a constant voltage (temperature) corresponding to 212 F.

Bibliography

la. V. Paschkis, "Studies on Solidification of Castings," TRAN-SACTIONS, A.F.S., vol. 53, p. 90 (1945).

1b. V. Paschkis, "Influence of Properties on Solification Of Metals," Transactions, A.F.S., vol. 55, p. 54 (1947).

Ic. V. Paschkis, "Studies on the Solidification of Aluminum Castings," Transactions, A.F.S., vol. 56, p. 366 (1948).

ld. V. Paschkis, "Studies on the Solidification of White Iron Castings," Transactions, A.F.S., vol. 56, p. 371 (1948).

le. V. Paschkis, "Studies on the Solidification of Steel Spheres," Transactions, A.F.S., vol. 56, p. 373 (1948).

1f. V. Paschkis, "Comparative Solidification Studies," TRANsactions, A.F.S., vol. 57, p. 100 (1949).

2. V. Paschkis, "Influence of Dry Sand Conductivity on Rate of Freezing of Steel Slabs," Transactions, A.F.S., vol. 58, p. 147

3. Y. Tanasawa, Transactions Of The Society Of Mechanical

Engineers, Tokyo, vol. 1, 1935, p. 217.
4. J. B. Austin, "Factors Influencing the Thermal Conductivity of Non-Metallic Materials," Symposium on Thermal Insulating Materials, ASTM, Phila. p. 1 (1939).

5. C. F. Lucks, O. L. Linebrink and K. L. Johnson, "Thermal Conductivity of a Sand Mixture," Transactions, A.F.S., vol. 56, p. 363 (1948).

6. FOUNDRY SAND TESTING HANDBOOK, 5th edition 1944, American Foundrymen's Association, Chicago.

7. A. Schack, G. Goldschmidt, E. P. Partridge, Industrial Heat Transfer, J. Wiley & Sons (1933), New York.

8. H. Groeber, Warmeuebertragung, J. Springer, Berlin (1926).9. F. H. Nortoh, Refractories, McGraw-Hill Book Co., New

York (1942).

10a. V. Paschkis and H. D. Baker, "A Method for Determining Unsteady-State Heat Transfer by Means of an Electrical Analogy," Trans. ASME vol. 64, p. 105 (1942).

10b. V. Paschkis, "The Heat and Mass Flow Analyzer—a Tool for Heat Research," Metal Progress, vol. 52, p. 813, Nov. (1947). 11. Max Jakob, Heat Transfer, vol. I, J. Wiley & Sons, New York (1949.)

DISCUSSION

Chairman: H. A. SCHWARTZ, National Malleable & Steel Cast ings Co., Cleveland.

Co-Chairman: E. C. TROY, Foundry Engineer, Palmyra, N. J. JOHN PERKINS: 1 This paper changed my thinking considerably. We have a 31/2-in. diam camshaft where the analysis is such that we instigate a chill in a certain part of the casting. The balance of the casting has to be gray iron to machine easily. With the analysis being more or less constant, the pouring temperature constant, and as far as we can determine other variables being constant, with the exception of the moisture in the sand, we have been experiencing an inverse chill in the 11/8-in. flange of this casting. We theorized that if we have more than 3.4 per cent moisture in the sand this inverse chill would form and if the moisture content was less than 3.4 we would not get this chill. But following along the author's theories in the paper, we are wrong in thinking so. Is that correct?

CO-CHAIRMAN TROY: Was your assumption that the inverse

chill was a function of cooling rate?

Dr. PASCHKISS: Would you say then that you had rapid heat extraction over a thin layer. Is that the concept?

CO-CHAIRMAN TROY: No, the reverse; on the outside.

DR. PASCHKIS: Moisture content will increase the rate of heat

MR. PERKINS: That was my idea but in the paper the author states that it does very little of it.

DR. PASCHKIS: What is your explanation of this soft layer?

Mr. Perkins: A high rate of heat extraction.

Dr. PASCHKIS: That would mean you have a chill at that point. MR. PERKINS: No. The moisture was 3.2 per cent at that point to eliminate this inverse effect. My idea was' that if the moisture content was increased to 3.8 per cent there would be a high rate of heat extraction which caused this inverse chill.

Dr. PASCHKIS: When saying you had an inverse chill you mean

that you got a soft layer.

MR. PERKINS: That is right. If I kept the moisture below 3.4 per cent we got a good gray iron layer about 1/8 in. thick.

DR. PASCHKIS: We found, as stated, a chilling effect very near the surface for a very short time (about 60 sec). I do not

remember how deep the solidification goes in 60 sec. W. S. PELLINI: 2 The effects that Dr. Paschkis speaks of (very high rates of heat extraction during the first few seconds) pertain to the time when superheat, if present, is being liberated. The fact that he finds a marked increase (five times) in the heat extraction rate if water is present at the surface of the mold may have no significance for somewhat later times. Thus, the presence of water in the sand at early times cannot be considered to develop metallurgical effects equivalent to extraction of heat at rates equal to five times that for dry sand.

There is, however, a continuing effect of water presence which while not in the order of five times is felt throughout the solidification rates are increased moderately during the entire course

of solidifications

DR. PASCHKIS: How does Mr. Pellini explain the continuing presence of water after the first layer. With the interface temperatures which we have, it took less than 60 sec to get completely dry at quite a distance from the interface.

CO-CHAIRMAN TROY: We may be dealing with a phenomena of nucleation that may or may not be related to heat transfer. This problem of inverse chill can be tied up to an analysis problem just as well as it can be tied to heat transfer. There

are so many things that can cause an inverse chill.

CHAIRMAN SCHWARTZ: There is another explanation that when you suddenly cool the iron, it shrinks and you set up tensile forces. This is not my explanation. The separation of graphite is accomplished by an increase of volume so that forces tending to increase dimensionally favor graphitization, so you can make this inverse chill problem quite complicated.

¹ Asst. General Supt., Ford Motor Co. of Canada Ltd., Windsor, Ontario,

² Metallurgist, Naval Research Laboratory, Washington, D. C.

NI

001

47

ast

bly

ach

The

ily.

m

oles

nd.

nge

and

his

er,

eat

eat 101 2 nt. hat the e a an 3.4 ear oc ery ed. he ble he ed eat ch dirse ng m-

na er. sis ere nat up of ces

rio,

R. P. DUNPHY: Hydrogen has a very important effect on this.

MR. PELLINI: In deference to the man who raised the question we have to accept that he has statistical evidence relative to the water effect so we are not talking about happenstance. I concluded that there was statistical evidence that when the water was high inverse chill was obtained.

In regard to how deep and how quickly you eliminate the water, I cite that in 2 min the water was eliminated to a depth of $\frac{1}{2}$ in. In 3 or $\frac{3}{2}$ min the water was eliminated to a depth of $\frac{1}{2}$ in. Since gray iron sections of the size described take over 1 hr to freeze there is a long time during which the effect of water is felt by the casting and solidification rates are affected.

DR. PASCHKIS: It seems to me that if your figures are correct it will prove that moisture has no influence on the effect of time.

Mr. PELLINI: Let us take one figure. In 2 min water was eliminated to a depth of $\frac{1}{2}$ in.

DR. PASCHKIS: And you say a large casting may take 1 hr or more to freeze. The 2 min are negligible for solidification. After 2 min you would have no more influence.

MR. PELLINI: The first layer has been dried and behaves as dry sand but there are other layers which contain water—the process is a dynamic one.

DR. PASCHKIS: That is true theoretically, but it would seem to me that it is not likely to have practical significance. Possibly

the constant interface temperature which we assumed enters

C. E. SIMS: ^a I would like to give further consideration to Mr. Dunphy's remarks about hydrogen. Inasmuch as the inverse chill has been so definitely related to the moisture content of the sand, an explanation based on hydrogen effects would appear to be logical if it agrees with the known behavior of hydrogen. It should be noted that this particular iron has a critical composition. By that I mean, it is readily influenced to freeze either white or gray. The composition is chosen deliberately so that the chilled parts will be white for wear resistance and the parts that cool against sand will graphitize for machinability. A small change in composition may exert a large effect on the freezing characteristics of such an iron.

Hydrogen has been shown to have a potent effect on the graphitization of iron. This has been shown particularly for malleable iron. Also, contact of water vapor with hot iron is apt to introduce hydrogen. The reason for the inverse chill seems to be that hydrogen is driven inwards by dendritic segregation and that it is annealed out of the quickly frozen surface layer by the super heat escaping from the interior. It is quite unlikely that any set of conditions could cause the interior to cool faster than the surface.

² Assistant Director, Battelle Memorial Institute, Columbus, Ohio.

INCREASING RISER EFFICIENCY

By James O'Keeffe, Jr.*

Introduction

FACED WITH A CRITICAL SHORTAGE of many basic metals and alloys, with a continually shrinking supply of skilled personnel, and with increasing costs for materials, fuel, power, and labor, it is particularly timely that we investigate every possible means to increase the efficiency of our processes in an effort to maintain or even increase production and reduce costs wherever possible.

The methods utilized to assure sound, high quality castings in many cases mean that the yield of saleable product is less than 50 per cent of the total metal melted. The efficiency of the risering system used to produce these castings is therefore very low—probably less than 10 per cent.

It is the purpose of this paper to point out several methods which are available to increase riser efficiency, and at the same time actually to increase product quality.

Risering Fundamentals

First let us review briefly the fundamental considerations in risering. The first question which we must answer is just why a riser is necessary. Why it is necessary to melt and pour 1000 lb of metal to get 500 lb of saleable castings?

Unfortunately, fundamental data on the shrinkage characteristics of many alloys which we produce are not available in the literature. We are fortunate at least in having excellent data for pure copper. From these data we realize that 560 lb of copper will occupy 1728 cu in. at room temperature. When sufficient heat is added to this mass to melt the copper and superheat the liquid to 2400 F (1315 C), the same 560 lb will occupy a volume of 2000 cu in.—an overall expansion of 15.7 per cent.

When the superheated molten copper cools from 2400 F to 1981 F and solidifies, the resulting mass should occupy only 1862 cu in.—a volume contraction of 6.9 per cent. The reason for saying that the resulting mass should occupy only 1862 cu in. is that the specific volume of solid copper at this temperature

is 0.286 lb per cu in. Under proper conditions of solidification, i.e. controlled directional solidification so that there is a continuous supply of liquid metal to compensate for the volume contraction which takes place as the metal cools and solidifies, the result will be a sound dense metal structure of optimum density.

In the production of ingots and castings, the method of achieving sound dense metal structures involves maintaining a reservoir of molten metal in the hot top or riser to supply the feed metal required to compensate for the volume contraction or shrinkage which the metal undergoes during cooling and changing from liquid to solid.

Recent Published Information

Fundamentals of risering have been reported in the TRANSACTIONS of A.F.S. in numerous articles. Only the more recent publications will be mentioned here, but these authors cite many references to previous reports.

The subject of risering has been summarized by Caine^{2,3} in two papers which contain a great deal of information which is of practical significance and also points the way for a scientific method of risering castings. He emphasizes that "risering can be separated into two phases—positioning and dimensioning." The author then develops the necessity for proper positioning of riser so that the particular riser will feed the casting or casting section to which it is assigned. He then proposes a method for determining the optimum dimensions for risers. This method is based on the relative cooling rate of the riser and the casting, and includes two constants which must be determined empirically. For a complete understanding of this important subject, the original articles are recommended.

Caine also suggests that "Any method which will decrease the freezing rate of the riser in relation to the casting is quite useful in increasing riser efficiency."

The advantages of insulating sleeves and the method of producing these sleeves was reported by Taylor and Wick.⁴ These authors stated that by using insulating sleeves, sounder castings could be produced, with increased yields, and at reduced cleaning costs. O'Keeffe and Taylor⁵ show that it is necessary to insulate the riser of a bronze test piece ½ in. thick at base,

T

T

ca

ec

ar

ca

fre

pr

Wa

ph

ing

erl

is

fie

wil

(a

me

in

ods

liqu

it i

din

ma the

^{*} Metallurgical Engineer, Exomet, Inc., Conneaut, Ohio.

of

on tal

ces

ill

IV.

od

res

ot

m-

ch

ng

he

he

ut

ts.

by

of

so

st-

ed

he

n-

he

Te

ım

he

nd

m-

m-

ed.

to

ffi-

h-

or

u-

d,

ts.

u-

se.

-38

 $1\frac{1}{2}$ in. thick at the top, 3 in. deep, and 4 in. high. They report microporosity in a sand casting of this size even though a riser $2\frac{1}{2}$ in. by 3 in. by 3 in. was used—a riser volume $2\frac{1}{2}$ times the casting volume.

Bishop and Pellini⁶ report experimental data for the proper positioning and dimensioning of risers in order to assure adequate soundness in plates of various sizes. The work reported by these authors emphasizes the necessity for proper location of risers. Their data indicates the limited distance which the riser can be expected to feed. Bishop and Pellini conclude "... That plate castings with a width to thickness ratio of 3 to 1 or greater, containing 0.20 to 0.35 per cent carbon, can be made sound for a distance from the riser edge equal to about $4\frac{1}{2}$ times their thickness $(4\frac{1}{2}T)$. Of this distance $2\frac{1}{2}T$ results from the chilling effect of the casting edge; the remaining 2T is made sound by the riser."

Most of the data in the literature was obtained by experimentation using cast steel. However, Taylor and Wick, as well as Caine, emphasize the validity for non-ferrous alloys of the approach which is used for cast steel.

Riser Efficiency Defined

There is no general agreement at the present time as to the precise definition of riser efficiency. For the sake of terminology, the term as used in this paper will refer to the weight of metal which actually moves from the riser into the casting during the solidification process (W_1) , divided by the total weight of metal poured into the riser (W_2) , and this ratio converted to per cent, thus,

$$\frac{W_1}{W_2} imes 100 = ext{Riser Efficiency (per cent)}$$
.

It is assumed in this case that the risering practice was adequate to produce a sound casting.

It is important to bear in mind that there are two phases to consider in the problem of improving risering practices. In the first place, the risering system must be designed so that adequate casting soundness is assured. This means that the risers must be properly located and of adequate size to assure that there is liquid metal in the riser after the casting has solidified, i.e. the last metal to solidify will be in the riser.

Increasing the efficiency of properly located risers will then involve methods by which it is possible to (a) increase the rate of solidification of the casting, (b) decrease the heat loss from the riser metal, thus keeping the riser molten, and (c) add heat to the riser metal to remelt and superheat the reservoir of metal in the riser. And preferably we might use these methods in combination for maximum riser efficiency.

Increase Solidification Rate

The basic requirement of the riser is that it remains liquid longer than the casting or casting section which it is to feed. The longer it takes the casting to solidify directionally to the riser, the longer the riser must remain molten. When the riser is molded from the same material as the casting, solidification will take place in the riser at a rate approximating the rate of solidifi-



Fig. 1-528-lb bearing casting produced in Navy "G" metal. Casting dimensions: 11/2 in. I.D., 121/2 in. O.D., 131/4 in. high. Eight graphite chills 4x4x15 in. (as shown at right) were used. Two insulated risers 5.5-in. I.D. were poured 71/2 in. high. Yield for this casting was 84.9 per cent.

cation in the casting itself. This means that the riser volume for castings which solidify slowly must be very large.

An example of this is the 528-lb bearing shown in Fig. 1. By probing through the insulating cover on the riser of this casting we learned that this casting solidified in about 8 min when the casting was produced using chills as shown in Fig. 1. We estimate that this casting would have taken 45 min to solidify without chills. In this latter case, the riser would have to be considerably larger than would be the case for the casting which required only 8 min to solidify.

Use of metal chills to increase the rate of solidification and to assure proper directional solidification is a useful foundry tool. It is also a good method of increasing riser efficiency, since smaller, more efficient risers can be used when solidification takes place rapidly.

Minimum Riser Surface Area

To keep the riser open as long as possible, and thus increase the amount of metal which will go into the casting from a given size riser, it is necessary to minimize the loss of heat from the metal in the riser. This is emphasized by Caine,³ who explains the reasoning for this in detail. He points out that the ideal shape for a riser would be a sphere, since this geometric form presents a minimum surface area for the volume enclosed. This is not a practical shape for use as a riser in the sand foundry. However, investment molding operators might find a riser of this shape advantageous, coupled with a method for maintaining atmospheric pressure on the liquid in the riser.¹¹ The

writer has not used this type riser, but suggests the possibilities of such an approach for risering invest-

ment castings.

In sand molding practice, the most practical shape which will have a minimum surface area for the volume enclosed is a cylinder with a metal height equal approximately to the diameter. As an example of this consider Fig. 2. The rectangular riser 4-in. by 8-in. section dimensions, poured to a height of 6 in. will contain 192 cu in. of molten metal, and will have 144 sq in. of surface area exposed through which heat will flow to the molding material. The cylindrical riser of 6 in. diam, poured to a height of 63/4 in. will also contain 192 cu in. For the same volume contained, this riser has only 127 sq in. of exposed surface—a decrease in surface through which heat will flow to the molding sand of 13.4 per cent.

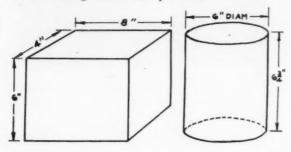


Fig. 2—Cylindrical riser has same volume but 13.4 per cent less side surface than the rectangular riser at left.

The important thing to bear in mind here is that whenever possible the riser should be cylindrical in shape or as close to a cylinder as possible. In many cases it is not practical to use a cylinder, but at least the riser should be as close to round in shape as possible for best efficiency. In this way, you minimize heat loss through the molding sand by decreasing the amount of surface exposed for the volume of metal, or heat, contained in the riser.

Another method of minimizing heat loss from the riser metal is to pre-heat the riser cavity by gating through the riser. This practice is widely used, and has two advantages. In the first place, the metal which goes into the casting is colder than the riser metal, since there has been continuous loss of heat during the pouring and running of the casting. This contributes to the formation of proper temperature gradients in the solidifying casting for directional solidification. Secondly, having the metal enter the casting through the riser keeps the hot metal at the risercasting junction where it should be, thus preventing the possibility of a shrinkage defect at the gate.

Insulate Top of Riser

Heat loss from the riser metal can also be decreased by covering the top surface of the metal in the riser. There are two classes of materials in general use for this purpose. Insulating materials, such as charcoal, cork, rice hulls, finely powdered carbon or graphite, lime, mineral perlite, diatomaceous earth, gypsum, or vermiculite, are inexpensive and provide top insulation. There is no published data on the relative value of these materials for riser insulation to the author knowledge. Dry floor sand is also used for top covering. However, heat losses when sand is used are known to be greater than when more efficient insulators aroused.

Mildly exothermic riser compounds are also used to a considerable extent to cover the tops of risers. It is probable that there is not sufficient heat generated by the reaction of these compounds to add any significant volume of heat to the riser metal. However, there is sufficient heat evolved to increase the temperature of the compound to that of the metal in the riser, and in some cases to a temperature in excess of the metal temperature. There is therefore no loss of heat from the top surface of the riser metal. For this reason a mildly exothermic riser compound is preferable in many cases to an insulator.

Several applications where mildly exothermic riser compounds are preferable can be mentioned. One case is when the metal in the riser is on the cold side, and the surface of the riser is essentially solid immediately after pouring is completed. In this case the heat from the exothermic compound is sufficient to prevent solidification and the use of a mildly exothermic compound is preferable to using an insulator.

Another case where mildly exothermic compounds are of value is on small risers. In this case the volume of heat in the risers is comparatively small, and the heat provided by the exothermic reaction is of value. It might be that the insulators do take a small amount of heat from the metal, whereas the mildly exothermic materials do not take any heat since they are raised to a temperature in excess of the metal temperature as a result of the chemical reaction.

You will notice from the data in Table 2 that the metal in a 3½-in. diam insulating sleeve remained molten 5 min longer when exothermic hot topping was used than when powdered gypsum was used for top covering. For this reason, the hot topping compound is used with insulating sleeves, and is of particular advantage in assuring maximum riser efficiency.

T

he

th

m

ta

th

si

sle

ad

It is also possible to use a layer of exothermic hottopping compound ½ in, thick on large risers, and then cover this with dry floor sand. In this manner you heat the covering layer by means of heat generated from the exothermic reaction, but you minimize the cost by using dry sand as the insulator. The heat loss to the sand in this case comes from the heat supplied by the exothermic reaction. There is thus a minimum heat loss from the riser metal itself.

Insulating Riser Sleeves

By using insulated risers, it is possible to produce castings of dependable soundness, and do this with a yield in most cases of over 80 per cent saleable castings.⁸ In this case the efficiency of the riser is over 40 per cent, compared with riser efficiency in normal practice of 10 per cent or less.

Consider in this instance the bearing casting shown in Fig. 3. This casting was produced from 80 Cu, 10 Sn, 10 Pb alloy. To assure adequate feeding, six insu-

e

c

a

d

r

ı

t

ıl

0



Fig. 3—Casting made of 80-10-10 metal. Gross weight 1100 lb; net weight 948 lb; yield 86 per cent. Six 3.50-in. I.D. insulated risers poured 7½ in. high were 4¼ in. high after solidification.

lated risers were positioned as shown. The risers were 31/2 in. in diameter, and poured 71/2 in. high. After solidification, the risers were flat across the top and averaged 41/4 in. in height. This means that 31.2 cu in. of metal went from the riser into the casting from each of these risers, or a total volume of 187 cu in. of feed metal was supplied to the casting from the risers. This casting had an estimated volume of 2960 cu in. It is therefore evident that metal volume equivalent to 6.3 per cent of the casting volume has been transferred from the riser to the casting during the solidification process. It is also evident that the riser efficiency is about 50 per cent—that is to say half of the metal which was poured into the riser went into the casting to compensate for volume contraction of the metal which was poured into the casting originally.

The data in Table 1 indicates the relative insulating value of the gypsum-base insulating formulation. These data were obtained by pouring molten bronze into the cavities listed, and then probing the open risers to determine the time required for the metal to solidify completely. It is interesting to note tha 69.7 cu in. of 80-10-10 bronze can be kept molten six times as long in an insulated cavity covered with exothermic hot topping as would be the case for a normal open riser of the same size.

The data in Table 3 was obtained by placing a thermocouple in the center of the mold cavity. This more exact method substantiates the information obtained by simply probing the top of the riser.

The high efficiency of a riser of this type is due to the fact that the metal does not solidify against the side of the riser. The insulation of the formulation is so effective that not even enough heat is removed to cause a solid skin to form at the riser-metal interface. This is evident in Fig 3. The insulating sleeve is shown in the right-center background, and this sleeve was poured to a height of ½-in. below the top of the sleeve.

The greatly increased riser efficiency which is possible through the use of insulated risers offers many advantages in casting production. In the first place,

TABLE 1-DENSITY OF PURE COPPER¹

		Specific \	Volume of	
Temperature	State	Grams per	Lbs per cu in.	560 lb cu in.
68 F (20 C)	solid	8.96	0.324	1728
1981 F (1083 C)	solid	8.36	0.302	1862
1981 F (1083 C)	liquid	7.92	0.286	1958
2400 F (1315 C)	liquid	7.76	0.280	2000

TABLE 2—TIME REQUIRED FOR BRONZE* TO SOLIDIFY IN SAND AND INSULATED CAVITY

			Time Requir	ed for Meta	al to Solidify
Cavity Diam, in.	Dimensions Height, in.		Sand Cavity, Min	Insulated Cavity Top Open, Min	Cavity
2.00	5.50	17.3	11/2	3	41/2
2.50	5.50	27.0	2	6	10
3.50	7.25	69.7	41/9	111/2	22***

* Metal was 80-10-10 poured at 2050-2075 F.

** Previously used sleeves crushed and used to insulate surface.
*** When exothermic hot topping was used to cover the top
of this size cavity, the metal remained molten for 27 minutes.

TABLE 3-TIME REQUIRED FOR ALUMINUM BRONZE*
TO SOLIDIFY

Cavity Dimensions	0. 1 3½ in. dia.	× 7 in. deep	
Metal Liquid	Gypsum	Green Sand	
Metal Liquid plus	10 Min	Less than	1 Min
some solid	20 Min		5
	30 Min		6
aring temperature 2200 F	2040 F		

First recorded temperature 2040 F Mildly exothermic material used as top covering.

Pou

Cavity Dimensions	0. 2 31/₂ in. dia. × 7	in. deep
	Gypsum	Dry Sand
Metal liquid	5 Min	0 Min
Metal liquid plus		
some solid	18 Min	5 Min
Total	23 Min	5 Min
First recorded temperature	1900 F	1860 F
Mildly exothermic material use	d as top coverin	g.
* 8-aluminum, 31/2-iron, bal. co	pper	

these risers can be used to assure adequate feeding on troublesome or critical jobs. Numerous illustrations of applications for insulated risers have been reported in recent technical articles.^{4,8,9}

The use of insulated risers also provides a means of increasing output where melting capacity is limited. This is true on an over-all basis or for a particular job. As an example consider the manganese bronze valve stem in Fig. 4. This casting required only 330 lb of metal to pour when insulating sleeves were used as shown in the illustration. Normally this casting would have required 520 lb to pour. By using sleeves the casting was produced with 190 lb less metal than would be required without insulation—a 36.5 per cent saving in metal.

In addition, the riser for this casting without insulation would be of 10-in. diam. This riser would be difficult to cut off, and would also have to be quartered by sawing or cut-off wheel for remelting. On the other hand the 3.5-in. diam risers can be removed readily, and there is no problem with remelting.

Another advantage is that the diameter of the riser when properly insulated is usually one-half the diameter of a riser for the same casting when the riser is formed in natural molding sand. This means that insulated risers can be used to advantage in many cases where there are restrictions as to the size riser which can be used. This might be due to bars in the flask, actual flask dimensions, or other similar causes.

It is also possible to use an exothermic moldable material to produce riser sleeves. This material would function in a manner similar to the mildly exothermic hot-topping compound. The writer is not familiar with the details of this method, and for this reason is not in a position to evaluate this method for increasing the efficiency of risers.

Highly Exothermic Riser Compounds

It is also possible to remelt and highly superheat the riser metal by means of a highly exothermic reaction. The reaction between properly sized aluminum powder and suitable metallic oxides produces liquid metal of controlled composition. This method has been reported by Lutts, Hickey, and Bock.⁷ The method involves pouring the riser short, waiting, and then superheating the riser by the addition of the correct amount of exothermic material.

This method is used commercially in the production of pure nickel and nickel alloy castings. The method is of particular value for these castings, since the gypsum-base insulating material will flux when used with nickel and nickel alloys.

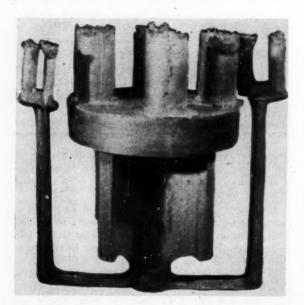


Fig. 4—Manganese bronze valve stem. Yield of 78.5 per cent was achieved on this 259-lb casting by using four 3½-in. I.D. insulated risers poured 5½ in. high.



Fig. 5—The three bearings shown were fed from the 2.50-in. I.D. riser. Gross weight, 68 lb; net weight 51 lb; yield 75 per cent.

The method is not as attractive for aluminum and copper-base alloys, since these castings can be made with high riser efficiency by using insulating sleeves. In addition the high temperatures involved cause volitilization of zinc and lead. Further, the metallic oxides used to prepare the mixtures are expensive and difficult to obtain at the present time.

Lutts, Hickey and Bock report "A word of caution is desirable... as to the explosive character of copper oxide when inadvisedly used alone or in certain combinations with other oxides, particularly aluminum not properly graded. This situation makes the production of exothermic powders for copper bronze rather more difficult than is the case with monel metal or steel." They add "No particular problem exists with ordinary steel, nickel copper and chromium copper mixtures."

There is one other method for increasing riser efficiency which should be mentioned—that of feeding several castings from one common riser. This method is illustrated in Fig. 5. This procedure has numerous advantages. In the case illustrated; the method, combined with the use of an insulated riser, enabled the foundry to produce three pieces per mold. In previous practice it was possible to make only two castings per mold due to the riser size.

Summary

Production of sound, dense castings requires first that risers be located so that solidification will proceed directionally to the source of feed metal in the riser. Secondly, the dimensions of the riser must be such that there will be sufficient liquid metal in this reservoir to compensate for the volume contraction of metal which was originally poured into the casting.

The actual efficiency of the riser can be increased by any method which will (a) prolong the time that the riser metal remains molten or (b) increase the rate of solidification of the casting.

Use of circular risers for minimum riser surface area, use of chills to increase the rate of solidification of chunky castings, and gating the casting through the riser are methods which should be used to assure efficient feeding.

Heat loss from the top surface of open risers should be minimized by the use of suitable insulation or exothermic hot-topping compound. Insulated risers can be used to achieve yields of over 80 per cent, even on castings made from high-shrinkage alloys such as high-conductivity copper, aluminum, and manganese bronze, and high-shrinkage aluminum alloys. The advantages of riser insulation should be considered by producers of non-ferrous castings.

Highly exothermic riser compounds may be used to greatly increase the efficiency of risers for nickel and nickel-alloy and steel castings.

References

- 1. ASM Metals Handbook, 1948, p. 903.
- 2. J. B. Caine, "Risering Castings," Transactions, A.F.S., vol. 57, p. 66 (1949).
- 3. J. B. Caine, "A Theoretical Approach to the Problem of Dimensioning Risers," Transactions, A.F.S., vol. 56, p. 492 (1948)

- H. F. Taylor and W. C. Wick, "Insulating Pads and Riser Sleeves for Non-Ferrous Castings," Transactions, A.F.S., vol. 54, p. 262 (1946).
- 5. J. O'Keeffe, Jr. and H. F. Taylor, "Melt Quality of Bronze Castings," *The Foundry*, Feb. 1950, pp. 94-96.
- 6. H. F. Bishop and W. S. Pellini, "Soundness of Cast Steel Plates," Transactions, A.F.S., vol. 58, p. 185 (1950).
- 7. C. G. Lutts, J. P. Hickey, and M. Bock II, "Exothermic Materials," Transactions, A.F.S., vol. 54, p. 337 (1946).
- 8. W. H. Piper and J. O'Keeffe, Jr., "Insulation Increases Riser Efficiency," *The Foundry*, Dec 1950, pp. 73-75.
- 9. S. W. Brinson and J. A. Duma, "Kncck-off Risers for Non-Ferrous Castings," Transactions, A.F.S., vol. 54, p. 795 (1946).
- 10. L. W. Eastwood and J. A. Davis, "Microporosity in Magnesium Alloy Castings," Transactions, A.F.S., vol. 54, p. 254 (1946).
- 11. H. F. Taylor and E. A. Rominski, "Atmospheric Pressure and the Steel Casting," A.F.S. Transactions, vol. 50, p. 215 (1942).

DIE CASTING MAGNESIUM ALLOYS

R. C. Cornell*

MAGNESIUM is the baby of the die casting materials in this country. Not until 1931 did Charles Pack first die cast magnesium in his vertical cold chamber machine. Despite this good beginning no progress was made until 1935, when the horizontal cold chamber machine came into use. In the interim, considerable experimentation served to demonstrate that the airgooseneck machine was unsafe and incapable of producing quality magnesium castings. Steady progress has been made since 1935.

In Germany the start was earlier, in 1925, again with the air machine. By 1932, however, the Germans were producing a considerable tonnage of magnesium die castings, largely with aid of the Polak cold chamber machine. The trend then reversed, and by the start of World War II the Germans were producing a large proportion of their magnesium die castings on hot chamber machines, both piston and air types. The hot chamber machine has not been commercially reverted to in this country although we may see it before long. Considerable research is being carried on in this direction. The hot chamber machine has a strong incentive in its high casting speed, as well as in the complexity of the castings it can produce.

Special Methods Required

The physical and chemical characteristics of magnesium differ somewhat from those of other metals, and special methods are required for casting and finishing. Magnesium tends to oxidize quite readily at higher temperatures (1000 F). It may also be oxidized by sulphur. This tendency to oxidize makes it necessary to melt magnesium under a protective flux blanket of sulphur dioxide.

An interesting paradox exists with respect to sulphur. Briefly, sulphur dioxide inhibits burning until burning starts. Then it feeds the burning because of magnesium reaction with both the sulphur and the oxygen of the decomposed sulphur dioxide. This peculiarity is the cause of some of the fires occurring in handling magnesium. Water and sand are also dissociated by burning magnesium with resulting violent reactions, and are not suitable as fire extinguishers.

The metal is tarnished quite readily by moist and salty atmospheres. Its oxide coating (unlike that on aluminum) is not a sealer, and therefore cannot protect the appearance of castings subjected to exposure.

Melting and Refining: Because of its tendency to burn when molten, magnesium requires special and experienced handling in melting and refining. The casting pot is protected by sulphur dioxide by means of a special, self-generating hood over the pot. The casting scrap (sprues, dripping and flash) cannot be properly cleansed in the casting pot, and must be returned to an open refining pot where a flux is used to clean and protect the melt. From there it may be transferred in the molten state to the casting pot or re-pigged.

The casting pot must be fluxed out periodically. This fluxing operation is carried out with considerable care to ensure that no flux remains to contaminate the castings. It is very important to keep the casting pot free from dross accumulations at the metal line. Such accumulations seem to promote burning of the metal.

Minor alloy additions must be made to the remelted die cast scrap to maintain the proper composition. The most commonly used alloy, AZ91 (R), is not susceptible to alloy contamination in the manner of



Magnesium die castings have many critical applications, especially in aircraft engines and accessories.9

zinc-base alloys. Nickel and copper are the detrimental contaminants because they increase the corrosion rate of the metal when exposed to salty atmospheres. Excess zinc in the alloy serves only to increase its already strong tendency toward hot cracking. Aluminum above the composition limits makes the metal hard and brittle.

Fumes from the melting fluxes are very corrosive toward buildings and castings. An unprotected metal building can be completely eaten away in several year's time. Special precautions should be taken in the construction of a building to be used for magnesium melting. Any steel work must be well protected with acid-resisting paints. Castings must be protected during storage and processing. In the author's plant, the magnesium castings are removed from the casting department promptly, and then each box is covered with a canvas.

Die Casting Technique: Unquestionably, magnesium presents more difficulties than any of the alloys commonly used in die casting. Fortunately, the magnesium die caster has to contend with only one prac-

^{*} Vice President, Litemetal Dicast, Inc., Jackson, Mich.

ni-

he

al

ng

ed

of

is

ıl

ıl

n

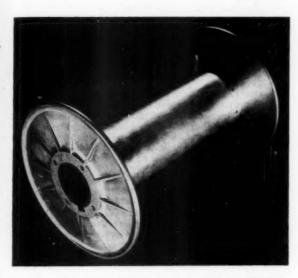
d

d

d

tical alloy. This difficulty in casting stems from two characteristics of the alloy: its low heat content at the pouring temperature, and its wide freezing range, the latter being the cause of hot cracking. When cast from the same pouring temperature, magnesium must give up only about half as much heat as does aluminum before being removed from the die. This means that magnesium will not flow as far or into as thin a section as would aluminum. On the other hand, it means that magnesium can be removed from the die with less delay, thereby greatly speeding the casting cycle.

In a given die, magnesium can be cast 20 to 30 per cent faster than aluminum. On the other hand, the casting efficiency of magnesium is definitely lower. For example, a job which produces 8 to 10 per cent



This textile warp beam or spool has 13-in. diameter magnesium die cast flanges, joined by central tubing.

scrap in aluminum may run 15 to 20 per cent in magnesium, which offsets some of the casting speed advantage of magnesium.

Gating for magnesium is not too different from that for aluminum. The major difference, in the author's experience, is in the aspect ratio of the gate. The broad, thin gate so often seen on aluminum die castings is not often used for magnesium. The area of the gate will depend on the method of filling desired. The great majority of castings are made with a small "spray" gate. Venting must be more carefully thought out and more generous, in view of the faster injection normally used with magnesium.

Extremely high pressures are sometimes used when exceptional soundness is required. These high pressures are used only when absolutely necessary in view of the detrimental effect on dies and the accompanying difficulty of casting injection. While on the subject of high pressures, it might be well to point out that special means of attaining those high pressures should be very carefully studied before purchasing equipment for die casting magnesium.

Much money and energy have been wasted in try-

ing to produce aircraft quality magnesium die castings with such devices as pre-fill injection. Injection multipliers may have the same drawbacks unless they have the capacity of getting the metal into the die fast and hard without a letdown or hesitation at the end of the stroke. The greatest success has been achieved with a large accumulator on a fairly large cylinder plus generous valving and piping to put the metal in hard and fast.

Magnesium places more of a demand on die casting equipment than any other metal. Most important, the machine must be flexible, i.e., have a wide range of plunger speed and pressure. Cooling of the dies is much less of a problem with magnesium than with aluminum. This is one reason why dies tend to last longer with magnesium. In some cases it is figuratively necessary to horsewhip the operators to get them to cast fast enough to keep the dies at the optimum operating temperature. Dies must be sturdy to withstand the terrific mechanical shock. By the same token, the die clamping must be more powerful to hold the dies together, which means a larger machine.

Magnesium is prone to give more trouble with surface imperfections than other metals. These imperfections consist of cracks, misruns, shrinks, swirls (turbulence marks) and oxidation. The last mentioned is the roughness resulting from oxide buildup on hot spots of the die. It can sometimes be combated by selective water cooling in the die, but more often can be eliminated only by changing the design of the casting. One automotive casting produced in the author's plant had to be converted to aluminum because oxidation prevented a smooth paint job.

Cores Easily Removed

Little difficulty of this sort has been experienced in aluminum so long as water lines are kept clear. On the other hand, little difficulty is experienced with magnesium galling on cores where draft is inadequate, nor does die washing result with magnesium. Fundamentally, magnesium is a slippery metal. Cores have been pulled 2½ in. deep with no draft and no tendency to scuff. Long, thin cores give less trouble in aluminum because of the reduced mechanical shock.

Trimming: Die trimming of magnesium die castings is done in much the same manner as with aluminum or zinc. Less shaving action on the trim dies is required. In fact, a chamfer on the cutting edge is often used to give a smooth or burnished job. Trim dies are seldom hardened as is necessary in aluminum trimming. Also, it is never necessary to oil the castings or trim dies to prevent scuffing.

If grinding or buffing are to be done, special precautions must be observed in collecting and disposing of the grinding dust. The author prefers individual and portable water-wash collectors for each grinder. Any type of collector must be properly maintained and regularly cleaned to guard against serious explosions. Magnesium dust mixed with 5 to 30 per cent water makes a very potent explosive. Only a spark is required to set it off.

In six years of experience with individual collectors this plant has never had a magnesium fire that

resulted in a lost-time personal injury or property loss. To prevent hydrogen and magnesium dust explosions any collector must be kept clean and well ventilated. The sludge should be burned on a dump.

Drilling, tapping, turning and milling may be done dry without the need of dust collectors. In these operations it is important to maintain sharp tools and to prevent the accumulation of chips so that any fire will be a minor one. Trimming flash is returned to the melting room for reclamation. Chips and grindings are not economically recoverable and are usually burned on the dump. If large quantities of coarse chips are produced they may be sold to a smelter, which is better equipped to obtain economical recovery. The little machining necessary on magnesium die castings will demonstrate its excellent machineability. Tool life is much greater than with aluminum.

Finishing: After trimming, one of two treatments is given the castings, depending on their use. They may be either chrome-pickled or oiled. This is necessary to forestall undue corrosion of the castings during subsequent shipment and storage. The oil dip actually provides the best protection, but sometimes is objectionable with respect to finishing operations.

The pickle provides a good base for commercial painting operations. For aircraft work the castings are given (by the painter) an alkaline dichromate boil, followed by the usual zinc chromate primers, etc. Anodizing is possible but is not often used because of its cost. Plating of die castings has lagged behind the development for other magnesium forms.

Applications: The light weight of magnesium, twothirds that of aluminum, is its main selling point. Notable exceptions to this are in textile bobbins and lawn-mower gears. In the former, aluminum is undesirable because it tends to smudge and mark light colored yarns. In the latter, the wear resistance of magnesium is just as good as that of aluminum but, due to the current high price of secondary aluminum, magnesium is at present the most economical material. However, the current shortage of primary and secondary magnesium alloy may soon push this job into aluminum. Other applications are optical instruments (cameras-binoculars), portable business machines, portable tools, aircraft parts, and auto parts.

Relative Cost Factors Govern

The cost of the castings is a large factor in their salability in competition with aluminum or zinc. Roughly speaking (with normal metal prices), if an aluminum die casting weighs ½ lb or over, magnesium may be competitive. The same would be true in the case of a zinc casting weighing over 1 lb. Several years ago, when aluminum and zinc ingot prices were high and magnesium was normal, it was found that most of the usual arguments against magnesium

fell into insignificance because of the cost savings that could be offered. The fact that magnesium parts must often be painted where an aluminum or zinc part could exist without protection is a definite deterrent to their use.

The foregoing discussion boils down to this—there is a best use for each of the die casting alloys. Their continued advancement can be brought about only by careful application engineering. Each of the alloys, magnesium, brass, aluminum and zinc, must take its place alongside the others and stand or fall on its own merits.

Bibliography

- 1. A. Beck, Technology of Magnesium and Its Alloys, F. A. Hughes, Ltd.
- 2. W. Muller, Werkstoff Magnesium, 2nd ed, 1939, pp. 43-55, Die Casting.
- 3. Machinery (supp.), June 6, 1940.
- 4. Report of Magnesium Die Casting Conference, Jack & Heintz, Inc., Dec. 7, 1945.
- 5. H. Chase, "Die Castings in Magnesium Alloy," Metals & Alloys, Feb. 1941.
- R. Cornell and C. E. Nelson, "Principles of Die Casting Magnesium Alloys," A.F.S. Transactions, vol. 53, pp. 328-332 (1945).
- 7. Dow Chemical Co. Technical Bulletins DM 26A, DM 11a, DM 42, DM 73, DM 31a, DM 53, DM 47 and DM 14.
- 8. E. E. Stonebrook and W. E. Sicha, "Correlation of Cooling Curve Data With Casting Characteristics of Aluminum Alloys," A.F.S. TRANSACTIONS, vol. 57, pp. 489-496 (1949).
- 9. A. W. Winston and M. E. Brooks, "Magnesium Castings Production and Use," AMERICAN FOUNDRYMAN, Jan., 1948.

DISCUSSION

- Chairman: M. E. BROOKS, The Dow Chemical Co., Bay City,
- Mich.

 Co-Chairman: W. D. DANKS, Howard Foundry Co., Chicago.
- D. L. LAVELLE: 1 Why does the spray type of gating seem to work best in your practice whereas for magnesium you would not basically think that type would be best because of magnesium's tendency to oxidize and form dross?
- Mr. CORNELL: Since magnesium has a very small heat content and since the majority of die castings have relatively thin sections, it is not practical to keep the metal flowing in the die at a velocity at which it can complete filling of the cavity within the time limit allowed by the heat content of the material. In other words, the metal would freeze before it got there if in every case you would get the spray filling or squirting. We do fill some castings with a solid front, distinct from spray filling. However, the sections and the gates must be generous and it takes extremely high pressures to keep the metal moving.
- J. J. WARGA: ² I would like to have the author elaborate on pressure considerations in magnesium die casting, especially with regard to hot pressure machines opposed to cold pressure machines. Is there a limiting pressure below which magnesium cannot be cast readily or soundly?
- MR. CORNELL: I do not think there is a limiting pressure. The pressure required to put the metal in the die and fill the die properly without misruns is a function of several factors, among which are temperature of the metal and temperature of the die, insulating coatings on the die, etc. Before long there may be increased use of the hot chamber die casting machine in this country.
- ¹ Research Metallurgist, American Smelting & Refining Co., Barber, N. J.
- ² Process Engineer, Piasecki Helicopter Corp., Morton, Pa.

SILICON-CHROMIUM ALLOY IN COMPLICATED IRON CASTINGS

By

Ralph A. Clark*

ABSTRACT

Four automobile cylinder castings were poured from a typical cylinder grade of cast iron as follows:-

1. Conventional unalloyed cylinder iron.

2. The same iron as No. 1 alloyed with sufficient nickel shot and high carbon ferrochrome to yield an alloy level of 0.28 per cent nickel and 0.23 per cent chromium.

3. The same iron as No. 1 alloyed with 0.34 per cent chromium and 0.20 per cent silicon as 30 per cent silicon-50 per

cent chromium alloy.

lat ISL art n

re eir

ly VS. its

its

A. 55.

8

è

32

2,

1

g

ld

g.

ni

C.

at

in

in

n

lo

h

re n

ie

e,

4. The same iron as No. 1 alloyed with 0.51 per cent chromium and 0.30 per cent silicon as 30 per cent silicon-50 per cent chromium alloy.

Sectioning and exploration for Rockwell B hardness and microstructure indicated that all four irons were pearlitic in nature and quite similar in mechanical properties in arbitration bar sections. The chill depths of the four irons in a hatchettype specimen were similar. However, variation in cooling rate brought about wide variation in the microstructure and hard-

ness in critical sections of the cylinder castings.

The unalloyed iron and that containing 0.28 per cent nickel and 0.23 per cent chromium showed a decided tendency toward formation of primary ferrite along the surface of the cylinder bore and a tendency toward formation of ferrite in critical sections of the cylinder bores. The irons alloyed with increased amounts of silicon-chromium alloy showed less tendency toward formation of a ferritic surface layer and a more uniformly pearlitic structure throughout the casting. All of the siliconchromium additions resulted in an increase in the uniformity of hardness in the cylinder bores at a slight increase in the hardness level. No massive carbides were found in any of the specimens examined.

THE EFFECT OF VARIATION in section thickness and consequent cooling rate of gray iron castings is generally recognized. Work done by various investigators has fixed the fact that rate of solidification and subsequent cooling rate of the casting as determined by the thickness of section has considerable effect on the metallographic structure of gray iron and its physical properties. Generally speaking, the hardness and strength of a plain cast iron will decrease with increase in section size, and the amount of such decrease to be expected in castings of relatively simple design has been rather well determined.**

This is true as long as the iron has random graphite distribution and a normal pearlitic structure. However, when freezing rates are rapid and approach a rate which will yield a chilled or mottled structure, we obtain a fine, sooty fracture and a microstructure characterized in extreme cases by type D and E graphite with a matrix which may contain primary ferrite in either large or small amounts. A freezing rate which is not quite drastic enough to cause formation of primary ferrite may yield a pearlitic matrix with some abnormal graphite. The presence of primary ferrite in the sub-chill eutectiform structure results in lowered hardness and poor resistance to wear and scuffing. Thus, light sections of a casting or corners and flanges which freeze rapidly may be softer than heavier sections in the same casting.

Inoculation of Cast Iron

Inoculation of cast iron by the ladle addition of silicon in one form or another is a well established method of avoiding eutectiform structure. The addition of a carbide-forming alloy such as chromium is a common method of stabilizing the pearlitic structure and avoiding decomposition at high temperature, thus avoiding soft areas in slowly cooled portions of a casting. However, chromium as an alloying element has the effect of increasing chill depth, and the amount that can be used alone as ferrochrome without difficulty from chilled corners and flanges may be limited. For this reason, additions of chromium are often combined with nickel or with one of the graphitizing inoculants in proportions which will avoid an increase in chill depth.

A silicon-chromium alloy analyzing approximately 30 per cent silicon and 50 per cent chromium has recently been introduced to the foundry industry as an addition for cast iron. This material is so balanced in composition as to make practical the addition of chromium with no significant change in chill depth, while the silicon content has a mild inoculating effect. The silicon also confers excellent solubility compared to the conventional high carbon and foundry grade ferrochromes generally used as ladle additions

in the iron foundry.

^{*} Electro Metallurgical Company, a Division of Union Carbide and Carbon Corporation, Detroit.

^{**} See Cast Metals Handbook, 3rd edition, published by A.F.S. (1944) for data.

It would seem that the structure of the metal in any given part of a complicated iron casting would depend on several factors. These are: (1) The rate of freezing as determined by the thickness of the section; (2) The rate of subsequent cooling as determined by section thickness and, more particularly, by design; (3) The chemical analysis of the iron, and (4) Any processing which the iron may have received. The following is a report of an investigation of four six-cylinder automobile cylinder block castings of typical design cast from four irons of similar arbitration bar properties and chilling tendency.

Procedure

In this investigation four cylinder-block castings were poured by a regular producer of this type of gray iron casting from metal which had been alloyed in various ways.

Sample No. 1 represents a conventional unalloyed cylinder iron containing nickel and chromium only in residual amounts.

Sample No. 2 represents the same iron as No. 1 to which had been added sufficient nickel shot and high-carbon ferrochrome as a ladle addition to yield an alloy level of 0.28 per cent nickel and 0.23 per cent chromium. This is a low alloy composition which has been rather widely used in passenger car and light truck cylinders.

Sample No. 3 represents unalloyed iron to which had been added a balanced 30 per cent silicon, 50 per cent chromium alloy in sufficient amount to yield a chromium level of 0.34 per cent.

Sample No. 4 represents unalloyed iron to which had been added a sufficient amount of 30 per cent silicon, 50 per cent chromium alloy to yield a chromium level of 0.51 per cent.

The cupola charge was:

1,075 lb Malleable pig iron (2.24 Si, 0.92 Mn, 0.13 P, 0.03 S)

1,025 lb Steel scrap plate and flashings 1,900 lb Mixed returned scrap and purchased iron scrap

15 Silicon briquets-5-lb size

7 Silicomanganese briquets

650 lb Coke-mixed brands

155 lb Limestone

8 lb Cornell flux

Metal was tapped continuously into a small receiving ladle holding about two tons, using a calibrated tap hole and rear slagging. The temperature of the metal at the spout was approximately 2800 F.

Four cylinder block castings were poured from previously used 1,000-lb capacity ladles to which the alloys were added as the ladle was filled. The castings were poured at a controlled temperature of 2670 to 2680 F, as determined by optical pyrometer. Immediately after pouring each casting, 2 core sand mold containing 1.2-in. arbitration test bars was poured. Three hatchet-type chill specimens were also poured from each ladle.

While metal for these castings was taken from a receiving ladle containing some 2 tons of molten iron, the cupola was operating under average conditions with a continuous flow of iron into the receiver, and there was some turnover of metal in the receiver resulting in a slight drift in the analysis of the metal during pouring of these sample castings. In addition to samples No. 1, 2, 3 and 4 representing cylinder castings, chill specimens No. 5 were poured from the base metal during the filling of ladle No. 4 to determine any change in the analysis of the base iron during the progress of the test.

Chemical Analyses

Chemical analyses were made on samples prepared from the arbitration bars and from chill block specimen No. 5. The results are given in Table 1.

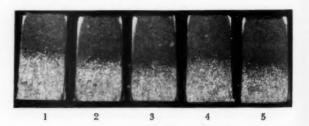
TABLE 1 % Total % Sample Treatment Cr Carbon Mn 0.85 0.11 0.13 2.12 0.08 Base Iron 3.24 0.14 2 Ni-Cr 3.26 2.21 0.28 0.23 0.34% Cr 3.26 2.31 0.11 0.34 4 0.51% Cr 3.24 2.45 0.13 0.51 0.91 0.11 0.14 2.26 0.13 0.15 5 Base Iron 3.24 (Final)

Chill Test Specimens

The chill depth of each iron as determined by fracturing the specimens is shown in Table 2.

TABLE 2										
Specimen No.	1	2	3	4	5					
•		0.28 Ni	0.54 Cr	0.51 Cr	Base					
Treatment	Base	0.23 Cr	5	*****						
	Chill	Depth in	1/32 in.							
	17	15 ·	12	15	12					
	17	15	8	15						
	17	15	10	14	14					

It will be noted that there was little difference in the chilling properties of the various alloy combinations. Specimens No. 1 and No. 5 of the base irons indicate a slight reduction of chill depth during the progress of the test.



1. Base Iron
2. 0.28 Ni-0.23 Cr Iron
3. 0.34 Cr Iron
4. 0.51 Cr Iron

5. Base Iron Taken at Time No. 4 Casting was Poured.

Fig. 1-Fractures of Broken Chill Specimens

VC:S

II.

ns

nd

re-

tal

on st-

ne

he

IC-

te

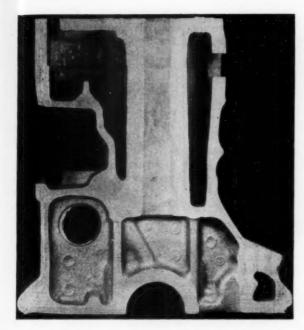


Fig. 2—Cross-section of typical cylinder block at line of removal of longitudinal section, showing interior contours and wall thicknesses.

Tests on Arbitration Bars

The results of transverse tests over an 18-in. span and hardness determinations on the arbitration bars from each of the four casts are given in Table 3.

TABLE 3

Sample No.	Type of Iron	Breaking Strength, lb	Deflection, In.	Hardness Bhn
1	Base	3020	0.270	212
1		2900	0.230	
2	Ni-Cr	3000	0.265	212
2	**	3000	0.250	
3	0.34 Cr	2920	0.255	212
3	9.0	3100	0.250	
4	0.51 Cr	3060	0.250	217
4	9.0	2380	0.190	207

Strength and deflection values of one of the No. 4 specimens were somewhat low, but in hardness and transverse properties the irons do not appear to be significantly different.

Metallographic specimens were prepared from each of the arbitration bars and each iron was found to be pearlitic in the 1.2-in. round section. There was comparatively little difference in graphite characteristics among the four bars.

Examination of Cylinder Block Castings

In order to obtain specimens for hardness surveys and metallographic examination, the cylinder block castings were cut to obtain longitudinal and circular sections from the two central cylinder walls. The longitudinal sections were cut from the valve chamber side and included a portion of one valve seat. The

circular sections were taken at a position approximately 2 in. below the top of the cylinder block which, as may be noted in Fig. 2, is just below the heavy section at the top of the block.

Hardness Surveys

Rockwell B hardness surveys were made on the longitudinal sections as shown in Fig. 4. Readings given are the average of three or four individual readings depending on the section thickness. It is evident from Fig. 4 and Table 4 that use of the silicon-chromium alloy resulted in a slight increase in the maximum hardness, combined with a considerably greater increase in the hardness of the soft zones at the center of the bores, resulting in a more uniform hardness along the length of the cylinder bore.

TABLE 4

		Rockwell B Hardness							
Sample No.	Treatment	Maximum at Top or Bottom	Minimum Thin Section	Hardness Difference					
1	Base Iron	-89.0	79.8	9.2					
2	Ni-Cr	90.8	81.4	9.4					
3	0.34 Cr	90.3	88.2	2.1					
4	0.51 Cr	93.9	88.9	4.0					

Examination of the individual hardness readings of the longitudinal samples revealed a difference in hardness between those taken near the surface of the section and those taken at the center of the cylinder wall. The silicon-chromium treated irons were uniform in cross-sectional hardness. See Table 5.

The maximum variation in hardness at a given position was about 6 Rockwell B units for samples 1 and 2, while it was 2 to 3 units in samples 3 and 4.

The untreated and low nickel-chromium irons tended to be soft in the water jacket area, while the irons treated with silicon-chromium alloy were rela-

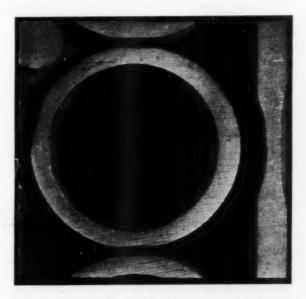


Fig. 3-Cross-section of cylinder block at location of removal of ring specimen.

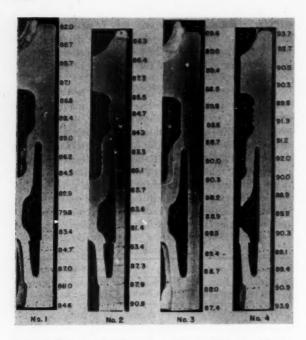


Fig. 4—Longitudinal sections from cylinder blocks, with average Rockwell B hardness values across wall thickness at approximately ½-in. intervals.

TABLE 5

No.	Treatm	ent Section	Dista						
No.		Treatment Section			from	Cyli	nder	Top	In.
			-1	2	3	4	5	6	7
1	Base	Center	87.6	88.5	90.5	86.4	83.5	88.0	88.5
		Bore Surface	85.2	84.4	85.0	83.0	81.8	81.8	87.0
2	Ni-Cr	Center	88.0	87.5	87.0	87.0	86.2	89.8	92.0
		Bore Surface	88.5	82.2	83.5	81.2	83.0	84.0	92.0
3	0.34 Cr	Center	86.8	91.5	89.0	90.0	89.5	89.8	88.5
		Bore Surface	87.8	89.0.	89.0	90.5	89.0	89.0	86.0
4	0.51 Cr	Center	90.0	90.0	91.0	92.0	89.5	90.5	91.5
		Bore Surface	91.0	87.7	90.5	91.0	90.0	87.5	92.7

tively uniform in this area. These differences were most significant near the bore surface where the lowest values were found in the untreated iron.

It will be noted in the hardness surveys of the circular sections (Fig. 5) that there was a sector at the bottom of each specimen that was relatively soft. This zone corresponded to the large flat in Fig. 3. Adjacent to this flat was the bottom of a valve port separated from the bore by the jacket core which was relatively light at this location. As mentioned previously, the thinning of the section here would tend to increase the rate of solidification, while the adjacent sections separated by the thin core retarded subsequent cooling. The result was that this section of the cylinder wall tended to have more eutectiform graphite and a large amount of ferrite which was reflected in the soft zone. It was evident that the silicon-chromium treatment in samples 3 and 4 eliminated the soft zones at the small flats between the cylinders, as well as at the valve port wall and effectively minimized the effects of change of section and cooling rate so that relatively uniform hardness was obtained throughout the cylinder walls.

Metallographic Examination

Specimens for metallographic examination were prepared from both ends of the longitudinal specimens, as well as at 1-in. intervals along their length, while four specimens at 90-degree intervals were taken from each of the circular sections, one being from the thinnest part of the cylinder wall.

In examining the specimens cut from the longitudinal sections, it was observed that, with respect to microstructure, the top and the bottom portions of the cylinders were approximately similar, but that there was a marked difference in some cases between these and the intermediate portions. A difference was also noted between microstructures adjacent to the bore and those at the center of the cylinder walls. To illustrate typical structures over the length of the cylinder, photomicrographs were taken of specimens prepared from the top, and from positions $2\frac{1}{2}$ in. and 5 in. from the top of the cylinder block.

There was little difference in the microstructure of the arbitration bar specimens of these four irons. All were of random graphite distribution and the matrix of each specimen was entirely pearlitic in nature. This was also true in some locations in the cylinder cast-

In Fig. 6, 7, 8 and 9 are shown the microstructure at the center of the cylinder wall at the top of the longitudinal section of each casting. It will be noted that in each case, the structure at this location is composed of the desired pearlitic matrix with type A random graphite distribution. The relatively heavy section at this location results in a freezing rate which

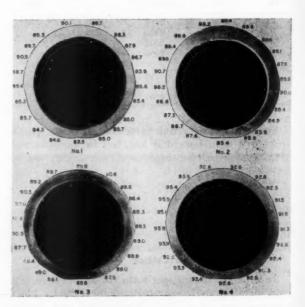


Fig. 5—Circular sections from cylinder blocks, with average Rockwell B hardness values across wall thickness at approximately ½-in. intervals.

NGS

and was

rere ecigth, ken om ngiect ons hat een nce to lls. the ens nd of All rix his stre he ed is

ch

Fig. 6-Block No. 1

Fig. 7-Block No. 2

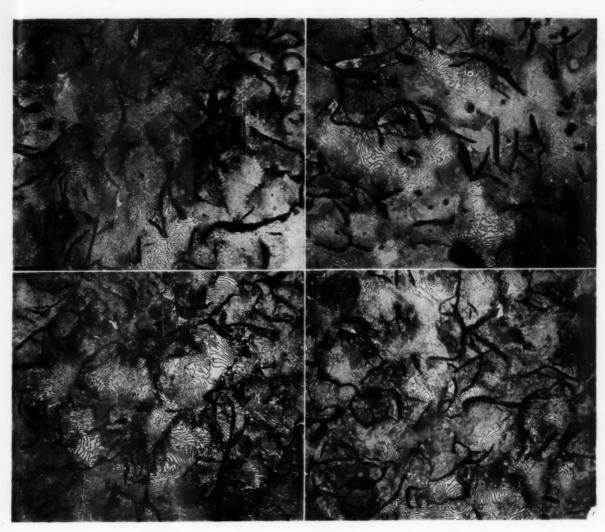


Fig. 8-Block No. 3

Fig. 9-Block No. 4

Microstructure at center of wall near top of longitudinal section of cylinder block. Nital etch. Mag. -- x 250.

is slow enough to allow the development of the random graphite structure, while the cooling rate after solidification due to location at the outside of the casting is rapid enough to discourage self annealing. While addition of alloy had little effect on microstructure at this location, it did result in a mild increase in hardness and the higher chromium irons exhibited a finer graphite flake size.

On the other hand, Fig. 10, 11, 12 and 13 show that adverse cooling conditions in some portions of the casting imposed by the design of the piece can result in a wide difference in the microstructure depending on the chemistry and treatment of the iron. These micrographs illustrate the structure of the metal at the thinnest portion of the ring specimens. It will be recalled that these ring specimens were removed from the casting at a location approximately 2 in.

below the head contact surface. The thin portion illustrated here was directly opposite the bottom edge of the valve port (Fig. 2 and 3).

The jacket core at this point was thin and became rapidly heated to a high temperature by the metal on either side, thus delaying cooling of the iron in this area. The result of rapid freezing was formation of a substantial amount of type D graphite associated with primary ferrite in the lower chromium irons. The result of delayed cooling of the solidified metal was self annealing, and formation of some secondary ferrite associated with type A graphite around the edges of the abnormal areas. The amount of ferrite and abnormal graphite present became progressively less as the chromium content increased, but this phase did not entirely disappear until a 0.51 per cent chromium level represented by block No. 4 (Fig. 13)

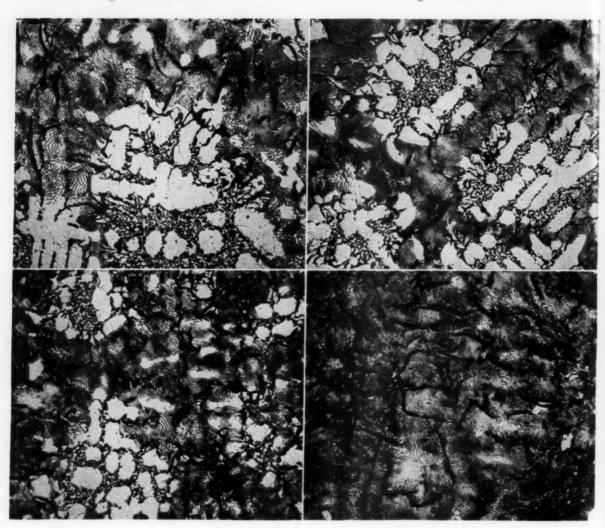


Fig. 12-Block No. 3

Fig. 13-Block No. 4

Microstructure at center of thinnest wall portion of circular section of cylinder block. Nital etch. Mag. -- x 250.

was reached.

The effect of slightly delayed solidification and more rapid subsequent cooling is shown in Fig. 14, 15, 16 and 17 taken from the thicker portion of the ring specimens. The heavier section delayed solidification, and the jacket cores were more massive at this location and exerted some cooling action after the metal was frozen. The result was that the abnormal ferrite-type D graphite structure showed up only in the unalloyed iron from block No. 1 illustrated in Fig. 14. All the alloyed blocks showed a pearlite-random graphite structure in sharp contrast to specimens taken from the same ring samples at a more critical area (Fig. 10, 11, 12 and 13).

Figures 18, 19, 20 and 21 illustrate the microstructure at the edge of the thin wall portion of the circular section of the cylinder bore. Accelerated freezing at

the surface of the cylinder bore resulted in a surface envelope of abnormal structure largely consisting of type D graphite and primary ferrite. The depth of this highly ferritic envelope decreased with increasing chromium content, and it practically disappeared at the 0.51 per cent chromium level, sample No. 4.

Figures 22, 23, 24 and 25 show the microstructure at the edge of the valve seat near the top of the longitudinal section of the block. Graphite structure was finer at this location than in the cylinder wall, but was mostly random in distribution, due to the relatively heavy section of the head contact. The graphite flake size again decreased with increasing chromium content of the iron.

CO

In summary, the metallographic examination of specimens taken from the center bores of these cylinder castings showed the following trends:—

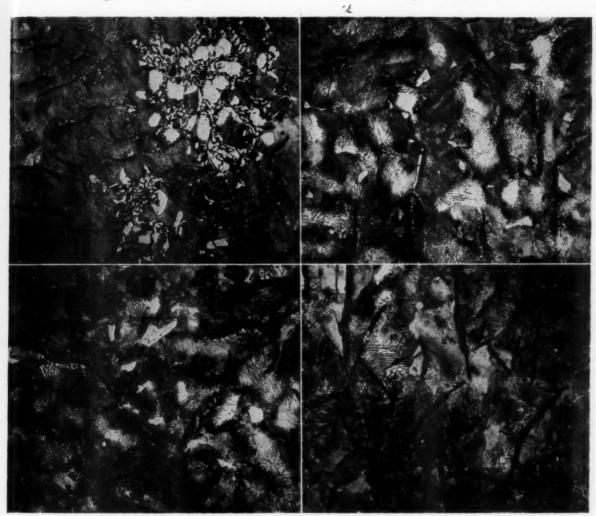


Fig. 16-Block No. 3

Fig. 17-Block No. 4

Microstructure at center of thicker portion of circular section of cylinder block. Nital etch. Mag. -- x 250.

1. In thin portions of the cylinder bore, there was a decided tendency toward formation of segregated areas of ferrite with type D graphite. The proportion of this eutectiform structure was increased where the construction was such as to promote rapid freezing of the iron followed by subsequent slow cooling. This condition, resulting in areas of reduced hardness at central portions of the cylinder barrel, was particularly pronounced in castings No. 1 and No. 2. The addition of sufficient chromium in the form of siliconchromium alloy effectively suppressed the ferritic structure and promoted the desirable random distribution of graphite in a pearlitic matrix, thus reducing the spread in hardness in castings No. 3 and No. 4.

2. At heavier portions of the cylinder barrel, somewhat slower solidification of the iron resulted in the

desirable random graphite in a pearlitic matrix in all the castings. The result of the higher chromium content of castings No. 3 and No. 4 was an increase in the hardness level of the pearlite and refinement of the graphite structure.

3. All the castings showed a ferritic envelope at the surface of the cylinder bore. The thickness of this ferrite layer decreased with increasing chromium content, being relatively heavy in casting No. 1 and becoming progressively thinner in castings No. 2, 3, and 4. The result of this envelope was the wide variation in Rockwell hardness readings at given locations in castings No. 1 and 2 which was mentioned earlier. Castings No. 3 and No. 4 did not show this variation in hardness, as the soft surface layer was too thin to come under the Rockwell penetrator.

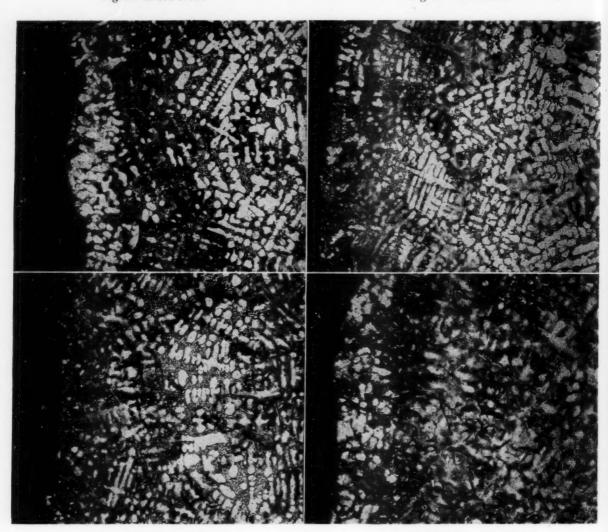


Fig. 20-Block No. 3

Fig. 21-Block No. 4

M

di

al di

de

re

ire

fo

th

ar

th

Microstructure at edge of thinnest wall portion of circular section of cylinder block. Nital etch. Mag. - x 250.

Conclusions

Examination of samples of automobile cylinder castings of an untreated iron, an iron alloyed with 0.28 per cent nickel and 0.23 per cent chromium, an iron alloyed with silicon-chromium alloy containing 0.34 per cent chromium, and another treated with the same alloy with 0.51 per cent chromium, revealed that, as compared to the other two irons, those treated with the silicon-chromium alloy had a greater uniformity in hardness over both the length and girth of the cylinder. In addition, samples of these last two irons were more uniform in hardness over the cross-section of the cylinder wall. This uniformity in hardness was combined with a slight increase in the maximum hardness and without materially effecting chill depth or transverse properties.

The metallographic examination and hardness explorations serve to illustrate that there is often variation in the structure of the iron at closely adjacent areas in an intricate casting as the result of minor variation in the freezing rate and subsequent cooling rate of the metal as controlled by the design of the casting. Spot checking of hardness or microstructure is useless as a method of determining suitability for the service application unless such checking is preceded by careful exploration which fixes the critical areas at which unsuitable hardness or microstructure are likely to occur. As an example, examination of a specimen taken from the top of the cylinder bore of any of these castings would have indicated a satisfactory pearlitic microstructure in spite of the fact that all of them except the No. 4 casting contained free ferrite and abnormal graphite in certain critical

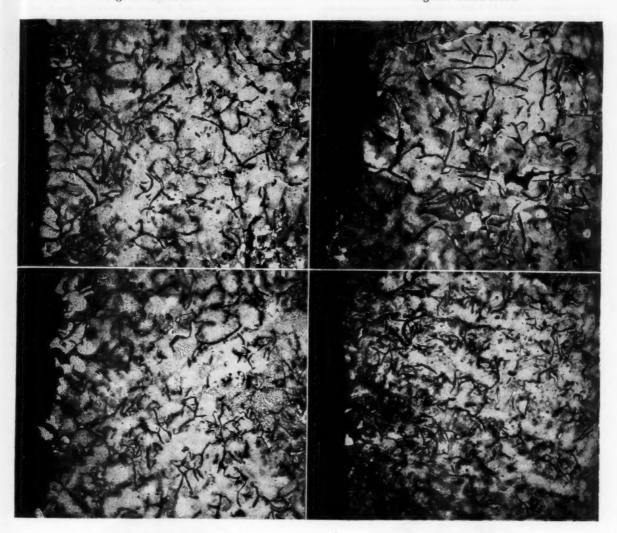


Fig. 24-Block No. 3

Fig. 25-Block No. 4

Microstructure at valve seat edge near top of longitudinal section of cylinder block. Nital etch. Mag. - x 250.

locations. Use of a sufficiently high chromium addition in this case in the form of a silicon-chromium alloy was one method for ensuring a random graphite distribution and the pearlitic matrix which is usually desired in cast iron to be used in applications where resistance to wear is required.

The silicon-chromium alloy additions produced an iron that had a higher hardness and greater uniformity of hardness and microstructure than either the untreated iron or the iron treated with the nickel and ferro-chrome alloy selected for comparison. The greater uniformity of microstructure at the bore of the cylinder should greatly increase wear resistance. It may be concluded that the silicon-chromium alloy is a suitable addition for the improvement of cast iron automotive cylinder blocks or other intricate iron castings.

Acknowledgment

The author wishes to thank the personnel of the Research Laboratories of Electro Metallurgical Company, Niagara Falls, N.Y., for chemical analyses, photomicrographs, advice and valued assistance in many phases of the preparation of this paper. Assistance and advice received from Mr. Walter Crafts and Mr. John L. Lamont were particularly valuable.

DISCUSSION

Chairman: H. H. WILDER, Vanadium Corp. of America, Detroit.

Co-Chairman: J. L. Brooks, Muskegon Piston Ring Co., Sparta, Mich.

E. A. LORIA (Written Discussion): We can confirm the author's findings on the decided tendency toward formation of

¹ Senior Engineer, Metallurgy, The Carborundum Co., Niagara Falls, N. Y.

R

et

pi ki

ca

CO

ire

ca

92

sil

th

ire

Wa

ire

res

W

ofi ch of ha be

primary ferrite along the surface of the bore of defective automotive cylinder blocks. In a metallographic and chemical analysis study of thin sections where cracking occurred, a coarse cellular structure comprising fine particle graphite and ferrite and a low total carbon content were the only features that could have caused the defect.

In a piece (designated as sample A) taken from a cracked water jacket, a survey of the general area of the 3/16 in. crosssection and the periphery of the circulae portion revealed large cells that could be observed visually, the microstructure being Type D graphite and primary ferrite. In traversing the $\frac{3}{16}$ in thick section, large patches of ferrite associated with the 85 per cent type D graphite were observed. In another 3/16 in. thick section (sample B), the microstructure along the crack and perpendicular to it was type D graphite and primary ferrite, the coarse cellular structure being observed visually or macroscopically all the way across the section. There was no sulphide inclusion segregate on either surface. Again, in a heavier piece taken from the bulkhead (sample C) with a variable 1/2 in. to 1 in. thick section, large cellular outlines could be observed visually at the corners. Microscopically, types D and B graphite and primary ferrite were seen. On traversing the 1/2 in. thickness, from edge to edge, mainly type A graphite was found except for the edges where the types B and D graphite were evident. At the center, the matrix structure was pearlite with a significant amount of secondary ferrite adjacent to the flake graphite. Still, there was no evidence of sulphide inclusions creating the defect. Representative micrographs illustrating these statements are presented as Fig. 26.

The chemical analysis of drillings taken from the samples just described were:

Sample	TC	CC	GC	Si	Mn	S	P
A	3.14	0.24	2.90	2.10	0.76	0.19	0.13
В	3.31	0.43	2.88	2.10	0.76	0.20	0.13
C	3.16	0.39	2.77	1.99	0.82	0.20	0.12

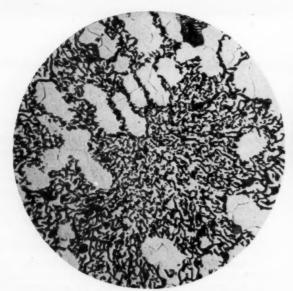
The significant feature here is the low total carbon content and the rather high sulphur content in the defective areas of an iron which should contain around 3.40 per cent total carbon and less than 0.15 per cent sulphur. The low total carbon content appears to be the most important consideration since there is considerable evidence to indicate that the lower total carbon irons are more prone to undercooling for any given cooling rate than the higher carbon irons. This is where inoculation comes in. It nucleates the solidification centers preventing undercooling and its attendant type D graphite, but the important thing

is to have the proper composition in the base iron for it will influence the need for and the type of inoculant that is necessary. The base iron cylinder blocks studied by the author contain only 3.24 per cent total carbon and this brings up the question of whether or not the type D microstructure at the wall portion of the circular section could not have been minimized (reduced in amount) if a higher total carbon iron had been produced? Would the author be able to show the same microstructural difference in ring specimens of base iron cylinder blocks analyzing 3.25 to 3.35 per cent total carbon, 1.95 to 2.25 per cent silicon compared to his silicon-chromium alloy treated blocks? Of course, some silicon bearing inoculant would also be used in the former to produce an even microstructure across the section.

In Table 1, the higher silicon contents of the silicon-chromium alloy treated cylinder blocks 3 and 4 are to be noted. Besides its graphitizing action, silicon lowers the amount of carbon required for the eutectic concentration. This is one of the reasons why higher silicon, lower carbon irons are generally finer grained than lower silicon, higher carbon irons of approximately the same strength and why the evaluation of grain or cell size (and consequently the tendency for formation of type B graphite) for the silicon-chromium alloy treated and untreated irons has to be made on the basis of comparable chemistry. Can the author show ring specimen structure and hardness values on this basis? That is comparing his iron with nickel-chromium stabilized or SMZ inoculated iron.

MR. CLARK (Written Reply to Mr. Loria): It is gratifying to learn that Mr. Loria's experience confirms our findings on the tendency toward formation of primary ferrite along the surface of the bore of automobile cylinder blocks.

Taking Mr. Loria's comments in order, we believe that cracked water jackets in cylinder blocks (or for that matter, cracking in any gray iron casting) usually result from conditions which cause residual stress in the casting of a magnitude exceeding the strength of the metal, thus resulting in fracture. Such failure may result from a variety of causes either mechanical in nature or partially the result of metal quality. We agree that low total carbon and high sulphur may promote cracking. Residual stress is the result of varying cooling rate in different portions of the casting causing contraction of the metal at a differing rate. Such uneven contraction may leave portions of a casting stressed in either tension or compression. The degree of such stress is affected by the design of the casting, the degree of collapsibility of cores, shake-out temperature, etc. Low carbon content of the metal may affect cracking due to the fact that iron which is low in carbon contains less graphite and contracts to



A-Type D graphite and primary ferrite.



B-Type A graphite and secondary ferrite.

Fig. 26-Representative microstructures in defective automotive cylinder blocks. Nital etch. Mag. 500x.

iG:

vill

Ott

the

all

zed

een

ro

de

.25

ted

he

OSS

its

red

hy

ed

he

nd

e)

he

on

m

to

he

ce

er,

ns

X-

re.

ni.

ee

ıg.

a

of

of

on

m

a greater extent during cooling, thus setting up somewhat higher stresses than when metal of higher carbon content is employed. The primary ferrite type D graphite structure probably also has a detrimental effect on cracking. As we know, this structure shows rather poor mechanical properties, and thus would be more likely to fracture under conditions of high residual stress. Primarily, however, we believe the problem of stress cracking is one of design and foundry practice rather than strictly a metal problem.

Mr. Loria questions whether the 3.24 per cent carbon content of the iron studied in this paper is not lower than desirable. We believe that typical automotive cylinder irons will vary in carbon content from 3.10 per cent for one large manufacturer to as high as 3.50 per cent for another. The average of the industry probably is in the range of 3.25 to 3.30 per cent total carbon content. Cupola operation which results in a carbon range of plus or minus 0.10 per cent from the desired level is considered to be under rather satisfactory control and, thus, we believe the 3.24 per cent carbon level of this iron representative of the

metal produced by the industry.

The phenomena reported in this paper have been checked at a number of automotive foundries melting somewhat varying irons. As an example, strips cut from a tractor cylinder block cast from an unalloyed iron containing 3.38 per cent carbon and 2.07 per cent silicon showed a Rockwell B hardness variation of 92.2 to 80.2 on the soft side of the cylinder bore. A similar casting alloyed with 0.31 per cent chromium in the form of the silicon-chrome alloy showed a Rockwell B hardness range of 94.0 to 90.8 and at the 0.44 per cent chromium level 95.5 to 89.7. The depth of the abnormal surface envelope was drastically reduced by use of the silicon-chrome alloy in a manner similar to that reported in this paper.

We believe that the results we have reported are typical for irons of carbon content normally used in cylinder castings. Carbon levels much above 3.40 per cent may result in a coarse graphite structure which gives an open appearance and poor finish in honed cylinders (this defect is often reported as porous bores). High carbon content may also result in leakers during

water testing of cylinder castings after machining.

We agree with Mr. Loria that effective inoculation of cylinder iron with a suitable silicon-bearing graphitizing inoculant would result in evening out of the graphite structure and the formation of a reduced amount of type D graphite at critical cooling rates. We believe, however, that such an unalloyed iron, particularly at a higher carbon content, would be more likely to show secondary ferrite and low hardness in slowly cooled portions of the cylinder bores due to lack of stability of the pearlitic structure. It has often been demonstrated that the presence of a small amount of chromium in the 0.30 to 0.50 per cent range confers a stability of structure which promotes retention of pearlite and satisfactory hardness under adverse cooling conditions. For this reason, we believe an inoculant of the stabilizing type such as the silicon-chromium alloy here described is desirable for cylinder castings.

In regard to the suggestion that these irons should have been compared on the basis of the same silicon level, we made an effort to compare the various irons on the basis of similar chill depth. The depth of chill is a rather definite limiting factor regarding the composition of unalloyed iron which can be used in cylinder castings. The base iron reported here is rather typical and in common use. The stabilizing inoculant used introduced sufficient silicon to counteract the chilling tendency of the chromium addition, and we believe that a foundry using it in this type of casting would normally produce a somewhat higher silicon content metal than if an unalloyed iron containing no chromium were used. We appreciate Mr. Loria's comments and hope that we have answered his questions in a satisfactory manner.

H. W. LOWNIE, JR.: ² In the author's No. 4 casting, his ladle addition consisted of roughly 0.30 per cent silicon and 0.50 per cent chromium. How would 0.30 per cent silicon alone have acted?

MR. CLARK: We did not make a casting treated with 0.30 per cent silicon alone. Obviously, I am unable to give an exact answer. I believe that silicon inoculation would have been beneficial in eliminating the eutectiform structure, but that the Brinell hardness or Rockwell hardness of such a casting would have been considerably lower, probably in the range of about 30 points Brinell. In other words, we would have obtained a pearlitic structure, but one with a softer pearlite than when the chromium treatment was employed.

MR. LOWNIE: I am curious about whether the author tried a straight silicon inoculant to solve this problem, and how it

acted.

MR. CLARK: We did not try a straight silicon inoculant. I presume the people who operate this foundry have tried this during the past years. I produced automobile castings myself for some eleven years, and an effective inoculation practice has never been very popular in most of the production automobile foundries. There are no doubt several reasons for this condition, and I believe some of them could be overcome very readily. Probably the biggest reason is that the lack of uniformity of structure in castings of this type is not recognized by the foundry operators. Our metallurgists realize the effect of cooling rate, but the average engineer is inclined to overlook it. We usually check the hardness of a casting at the cylinder head, and if a satisfactory range is maintained, we are inclined to overlook variations in the cylinder bore down in the center of the casting.

MEMBER: What was the analysis of that chrome-ferro-silicon

MR. CLARK: In round figures, 50 per cent chromium and 30 per cent silicon. This alloy is just a practical and convenient way of adding chromium.

MEMBER: Would there be a noticeable reduction in leaks in the use of that alloy?

Mr. CLARK: There have been quite a few carloads of this material used in cylinder work. As to its effect on leakers, I certainly would not claim that it could be used to reduce the frequency of this defect. I do not expect that use of this alloy would have much, if any, effect.

² Assistant Supervisor, Battelle Memorial Institute, Columbus, Ohio.

MODERN CORE SAND PRACTICE

By R. H. Greenlee*

MANY FOUNDRIES during recent months have been faced with the problem of increasing core production. In many instances this increase has been limited due to inadequate core oven facilities for baking. As a result, many foundrymen have shown considerable interest in the use of resins and the fast baking, highly polymerizing core oils as a means of increasing their core oven capacity. It is the purpose of this paper to endeavor to show how these core binders can be successfully incorporated into a well-balanced core sand mix which will produce a desirable core. The advantages and disadvantages of each will be discussed.

In the use of resins or the faster baking core oils to obtain desirable baked core physicals, it is essential that we have a fairly good working knowledge of the correct use and effect of all the raw materials which are used in the production of a well-balanced core sand mixture containing these bonding agents.

Success in the use of resins depends almost entirely upon having the proper ratio of cereal to water with a given resin content. It will be noted in Table 1 that extremely high baked tensile strengths are obtained when moisture and cereal are progressively increased. Results were obtained by using laboratory batches. Green compressions are the average of three tests and baked tensile strengths are the average of five standard briquets which were baked in a laboratory oven for 15 min at 500 F.

Michigan Dune Sand Used

It will be noted that a Michigan dune sand was used in carrying out many of the laboratory tests instead of a standard A. F. S. silica sand. This was done because a Michigan dune sand was used for core production and it was the thought that the results of laboratory tests using this sand would be more indicative of what could be expected when laboratory batches were converted to production mixes.

Figure 1 shows the grain distribution curve of the Michigan dune sand used for these tests.

TABLE I-RESIN SAND MIXES-LABORATORY BATCHES.

Effect of cereal and water on baked tensile strength.

*Cereal,		alculated D		Green Compression,	Baked Tensile Strength
%	%	%	%	psi psi	psi
0.5	0.5	2.0	2.0	0.25	145
0.5	0.5	3.0	3.15	0.28	150
0.5	0.5	4.0	4.1	0.29	184
0.5	0.5	5.0	5.15	0.30	193
1.0	0.5	2.0	2.05	0.33	191
1.0	0.5	3.0	2.8	0.38	217
1.0	0.5	4.0	4.1	0.40	232
1.0	0.5	5.0	5.05	0.32	253
1.5	0.5	2.0	2.1	0.51	230
1.5	0.5	3.0	3.3	0.54	284
1.5	0.5	4.0	4.15	0.50	299
1.5	0.5	5.0	5.0	0.42	295
2.0	0.5	2.0	2.15	0.68	172
2.0	0.5	3.0	3.0	0.64	268
2.0	0.5	4.0	4.2	0.60	319
2.0	0.5	5.0	5.0	0.56	338
2.5	0.5	2.0	1.9	0.82	114
2.5	0.5	3.0	3.1	0.80	181
2.5	0.5	4.0	4.1	0.73	328
2.5	0.5	5.0	5.2	0.71	294
3.0	0.5	2.0	1.95	0.98	90
3.0	0.5	3.0	3.2	0.94	161
3.0	0.5	4.0	4.05	0.89	301
3.0	0.5	5.0	4.85	0.90	264

Addition-Per cent Weight to Dry Sand Sand -Michigan Dune Sand

Cores bonded with the phenolic resins bake 30 to 50 per cent faster than cores bonded with the conventional core oils. Polymerization of the resin is completed shortly after the elimination of water in the core. This rapid polymerization, compared to the rather slow oxidization process characteristic of the slower baking core oils, accounts for the savings in baking time. Baking temperatures may range from 350 to 500 F. However, care and good judgment must

sta

^{*} Cereal-Corn Flour

^{**} Resin-Phenol

^{*} Foundry Supt., Auto Specialties Mfg. Co., St. Joseph, Mich.

IE9.

ed

ile

gth

5

0

4 3

n-

m-

he

he

he

in

m

ISL

be exercised in the selection of this temperature. Unless baking conditions are optimum, the lower temperatures will prove to give the best results. Cereal burn-out becomes more severe at higher temperatures and baked tensile strengths fall off rapidly once the core has been completely baked. The only advantage of high oven temperatures will be to shorten the oven cycle. Figure 2 shows the baking curve of a phenolic resin compared to a slower baking core oil. As will be noted, the baked tensile strength of the resin falls off sharply at 500 F compared to 400 F.

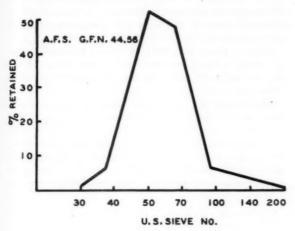
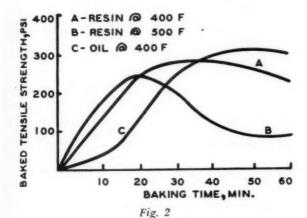


Fig. 1



Another advantage of the phenolic resins is their thermosetting properties. As a result, they do not require oxygen during the baking process. This property of being thermosetting is used to good advantage in the curing of resin cores by means of dielectric heating.

Because of their water solubility, resins migrate—along with the water—toward the surface of the core, the curing of which produces extremely hard surfaces. The interior will be slightly softer. This characteristic is desirable—that of having a core which will withstand handling with a minimum loss in core breakage—providing, of course, that core collapsibility is not impaired.

Moisture content of a core is an important factor in controlling surface hardness. Higher moisture content will result in correspondingly higher surface hardness. It appears that resin-bonded sands will tolerate a higher moisture than sands bonded with oil. In fact, it is only with the use of higher moisture ranges that resin-bonded sands may be successfully used to obtain optimum physical core properties, particularly hardness and baked strength. Surface hardness of cores whether bonded with resin or oil may be materially increased by spraying with water. It will be found that the baked strength of the core will be effected very little. Dextrin, the soluble component in cereal, also plays an important role in the development of surface hardness. Those cereals which contain a higher dextrin content will generally produce a core with a harder surface. During the baking process, dextrin, being soluble, will migrate to the surface along with the water. There, it is deposited and upon curing imparts a certain degree of skin hardness to the core surface. From this may follow the thought that cereals containing high dextrin contents are to be desired. This is true to a certain extent. Cereals containing exceptionally high dextrin contents, although increasing surface hardness, reduce the baked strength of the core. Green compression is also impaired.

Table 2 illustrates the effect on green compression and baked tensile strengths of a given core mix containing increasing amounts of dextrin. Mix "A" representing that mix with the lowest dextrin content, Mix "F" that with the highest. All cereals were heavy weight. Results are the average of three separate mixtures.

TABLE 2

Sand-Silica A. F.	Laboratory		10 lb
Cereal			
Moisture (Calcu	lated)		3.7-3.9
Mulling	Cycle and	Baking T	ime
	Cycle and	Durang 1	anne
Mulling Cycle:			
			0 .
Dry			2 mu
Wet			

	Cereal	Moisture,	Green Compression, psi	Baked Tensile Strength, psi	Surface Hardness
L-	A	3.8	0.64	146	54
	В	3.7	0.64	140	58
	C	3.7	0.60	132	63
	D	3.6	0.55	121	70
	E	3.9	0.48	100	72
H-	F	3.7	0.32	87	75

Sand stickiness is characteristic of core sand containing resin unless the sand grains are properly lubricated with a release agent. Sand mixtures of cereal and water only will tend to be sticky and as a result will not work well in a core box. With the addition of an oil to this mixture, stickiness is reduced or entirely eliminated. This is primarily due to the lubricating effect of the oil on the grains of

wo oi ter

sand. There are no lubricating qualities inherent in a resin. In fact, quite the opposite is true. Resin, by nature, has a definite tendency to make any core mixture sticky. As a result, certain agents must be added not only to eliminate the stickiness due to the cereal but also that due to the resin. Materials quite frequently used are the lighter oils, particularily kerosene in combination with the fatty acids, such as stearic, naphthenic, or oleic. Concentrations of these must be watched closely. These so-called release agents while eliminating core sand stickiness, do however, reduce certain physical properties of the baked core. Surface hardness and baked tensile strength are most noticeably affected.

Table 3 shows the effect on surface hardness and baked tensile strength with the addition of 0.3 per cent by weight of a release agent to a resin-bonded sand. Laboratory mixtures were made in which the composition was varied as shown in Table 3. Mixtures were made with and without the addition of a release agent. It will be noted that green compressive strengths are also affected, being considerably lower in all batches which contained a release agent. All materials added are in per cent by weight to dry sand.

Resin-bonded sands present somewhat of a problem in striking-off the surface of a green core. This is primarily due to the higher moisture content necessary. The sand has a tendency to tear and as a result creates a rather rough surface. In many cases, this condition is relatively unimportant. However, there are many instances in which the surface of the core plays a very important part in determining the overall dimensional length of a catting. Here, it is necessary for the strike-off surface to be smooth and beheld to certain close tolerances. It will be found that an aluminum strike-off bar, machined to an edge, will work satisfactorily. Resin-bonded sands have a greater tendency to stick to teel than to brass or aluminum. It may even be necessary to use aluminum inserts in place of steel inserts in certain locations of some core boxes.

Table 4 shows the effect of composition on strikeoff, using a flat aluminum bar which was machined to a cutting edge. All materials are given in per cent by weight to dry sand.

The storage life of a liquid phenolic resin is somewhat limited, particularly if stored as received. The solubility of a resin in water is a good indication of its efficiency in producing a well-balanced core mix. Over a period of time, this solubility will gradually reach a point where very little water can be added to a given amount of phenolic resin without causing the

TABLE 3—RESIN SAND MIXES, LABORATORY BATCHES. Effect of release agent on baked tensile and surface hardness.

Sand, lb	Iron Oxide, %	Silica Flour, %	Cereal, %	Resin, %	Release Agent, %	Calculated Moisture, %	Determined Moisture, %	Green Compression, psi	Baked Tensile Strength, psi	Hardnes
18*	1.0	0.9	1.5	0.75	0.3	4.0	3.75	0.63	288	85- 90
18	1.0	0.9	1.5	0.75		4.0	3.85	0.65	330	95-100
18	1.0		1.5	0.75	0.3	3.0	3.0	0.62	322	80- 90
- 18	1.0		1.5	0.75		3.0	2.9	0.70	349	95-100
18	1.0		1.5	0.75	0.3	2.5	2.55	0.62	255	80- 90
18	1.0		1.5	0.75		2.5	2.5	0.66	316	95-100
18	1.0	0.9	1.5	0.75	0.3	2.5	2.4	0.76	242	80- 90
18	1.0	0.9	1.5	0.75		2.5	2.4	0.89	298	95-100
18	2.0		1.5	0.75	0.3	3.5	3.55	0.71 ·	310	95-100
18	2.0		1.5	0.75		3.5	3.5	0.75	343	95-100
18	2.0		1.5	0.75	0.3	3.0	2.7	0.77	225	80- 85
18	2.0		1.5	0.75		3.0	2.7	0.80	280	90- 95
18	2.0		1.5	0.75	0.3	2.5	2.5	0.71	235	80- 85
18	2.0		1.5	0.75		2.5	2.5	0.85	290	90- 95
Sand-N	lichigan du	ine sand								

TABLE 4

•Sand, lb	Cereal, %	Silica Flour, %	Iron Oxide, %	Resin, %	Release Agent, %	Calculated Moisture, %	Determined Moisture, %	Green Compression, psi	Strike off
15				* *	* *	4.5	4.5	0.32	Good
15				0.5	0.3	4.0	4.2	0.25	Good
15			1.0	0.5	0.3	4.0	4.2	0.27	Good
15	* *	0.9	1.0	0.5	0.3	4.0	4.2	0.26	Good
15	0.5	0.9	1.0	0.5	0.3	4.0	4.2	0.37	No Good
15	1.0	0.9	1.0	0.5	0.3	4.0	4.1	0.41	No Good
15	1.5	0.9	1.0	0.75	0.3	2.5	2.3	0.71	Fair
15	1.0	0.9	1.0	0.75	0.3	2.0	1.6	0.60	Good
15	1.5	0.9	1.0	0.75	0.3	2.0	2.1	0.70	Good
15	2.0	0.9	1.0	0.75	0.3	2.0	2.1	0.90	Fair
15	1.5	0.9	1.0	0.75	0.5	2.0	2.0	0.80	Fair
15	1.5	0.9	1.0	0.75	0.75	4.0	3.9	0.71	Fair
15	0.5	0.9	1.0	0.75	0.75	4.0	4.1	0.40	Good
15	0.75	0.9	1.0	0.75	0.3	4.0	4.1	0.41	Good
15	1.5		2.0	0.5	0.3	4.0	3.8	0.76	Very Good
15	1.0		2.0	0.5	0.3	4.0	3.8	0.50	Very Good
15	0.5	4.0	0.5	0.75	0.3	4.0	4.2	0.90	Very Good
Sand-Mi	chigan dun	ie							,/

It will be noted that as the percentage of fines is increased, the strike-off surface becomes progressively better.

CE

63

1 t

nis

re

re

er-

es-

be

at

ge.

a

or

m

of

ce

ed

nt

10

he

of

X.

ly

to

he

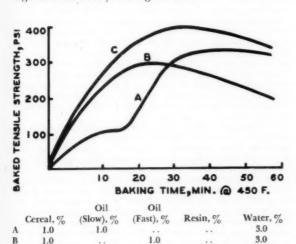
C

1.0

mixture to precipitate resin residue out of solution. If used in this condition, cores will possess very little surface hardness and other physical properties will fall off sharply. Solubility tests should be run daily and records filed in order to recognize and determine that point in the solubility curve which will indicate when the resin is no longer suitable for use. The storage life of a resin may be prolonged by diluting with water. Keeping resin stored at lower temperatures will also increase the storage life. Storage life is reduced considerably when the relin is stored at higher room temperatures.

Cleanliness is a criterion always to be stressed when handling the phenolic resins. Certain individuals are allergic to free phenol formaldehyde (always pre ent in varying amounts in all liquid phenolic resins). These individuals, however, generally constitute a small proportionate number of the total working force. Instructions in cleanliness should be put into effect, for cleanliness does appear to be a determining factor in many borderline cases of allergies of this nature.

Recently, several fast baking core oils have been made available to the foundry industry. Several of these oils will bake out equally as fast as the resin. Figure 3 shows the typical baking curve of this type of oil compared to a liquid phenolic resin and a regular moderately baking core oil.



All additions are in per cent by weight to dry sand.

1.0

3.0

Fig. 3

It will be noted that the faster baking core oils have lower tensile strength compared to either the resin or slower baking oils. However, in many applications, this difference does not prove to be significant.

Using the faster baking core oils, best core sand workability is obtained at lower moisture levels. These oils, when worked at higher moisture levels, have a tendency toward stickiness. Other properties such as transverse strength, collapsibility, and surface hardness compare favorably with the conventional core oils.

One disadvantage in the use of the fast baking core

oils is the fact that they rapidly air dry the core sand mixtures. This presents a problem, e-pecially in the making of small cores where the consumption of core sand is relatively small. However, in the making of larger cores where the volume of core sand used is larger and the turn-over is rapid this characteristic proves to be advantageous. Cores set up rapidly and as a result will be less prone to sag during transportation to the oven. Sagging, as a result of this movement, is reduced to a minimum.

Core Sand Mixing Procedure

Regardless of the type of core sand mixing equipment in use in any one plant or the type of core sand being produced, the same basic principles apply for the order in which each material is added to the sand during the mixing cycle in order to produce a core mixture which will have the maximum physical core sand properties for that particular core mix. To illustrate, consider a simple mix consisting only of sand, cereal, water and oil. All dry ingredients-in this case sand and cereal-should first be thoroughly mixed dry. If the cereal is not thoroughly mixed with the sand before water is added, the finished core sand mixture will contain very small lumps of cereal which have only partially been activated with water. This constitutes not only a watte of material but also a mix which is nonuniform in physical properties.

The time cycle for dry mixing should be carefully determined and closely followed thereafter. Next in order is the addition of water to the mix. Here. cereal which is now uniformly distributed in the sand is activated by the water and by continued mixing, green compressive strength is developed. The time element in this part of the mixing cycle should also be carefully determined. Mixing should continue until maximum green compression is reached. Otherwise, more cereal will be added than necessary resulting not only in higher core cost but also in the production of a core with a higher gas content. Oil should then be added and again thorough mixing is necessary to produce a core of maximum hardness and strength for a minimum amount of oil. Complete mixing cycles may range from 11/2 min to 7 min depending upon the type of mixing equipment. In any case, these cycles should be carefully determined and when once established, should be put into operation and enforced. A more uniform, workable and economical mix will be the direct result.

A rigid and close laboratory control of any core sand mixture cannot be over-stressed. Many core room problems come as a direct result of improper mixing and composition. In many core mixes, a difference of 0.2 per cent or 0.3 per cent moisture makes the difference between a well-balanced core sand mixture and one which will give trouble in the core room.

Conclusions

1. Thermosetting resins offer the foundryman a means of speeding up core baking time without sacrificing essential core properties.

As with conventional core oils, moisture content is an important factor in controlling core sur-

R

face hardness. With the resins, however, this factor appears to be more critical and such sands are usually worked at higher moisture contents.

- 3. By nature, the thermosetting resins impart a stickiness to core sands and impair workability. The use of release agents is helpful in overcoming this condition.
- 4. Storage life of phenolic resins is somewhat limited. A simple solubility test may be adopted to indicate the condition of the resin.
- 5. Fast baking, highly polymerizing core oils are available and offer rapid baking. These oils do not seem to offer the same high tensile strengths as the resins, but may often be used to advantage in core room practice.

DISCUSSION

Chairman: E. C. ZIRZOW, Deere & Co., Moline, Ill.

Co-Chairman: F. J. PFARR, Lake City Malleable Co., Cleveland. O. J. Myers: 1 I would like to congratulate the author on the presentation of an excellent paper. Only through such precise and practical work can we "of research" be able to develop new and better products for the foundry industry.

T. W. CURRY: 2 What is the frequency and severity of dermatitis among colored employees when using resins as core binders?

Mr. Greenlee: Approximately 2 or 3 per cent of the employees are allergic to phenolic resins. To some employees it presents no serious problem. However, to others the effect is more drastic. In fact in one or two cases, it has been necessary to take certain individuals out of the core room entirely due to their allergy to phenolic resins.

Mr. CURRY: Colored personnel are not immune to dermatitis from synthetic resins and neither are others. We have had cases where colored personnel have had slight cases of dermatitis and also the dark complexioned or swarthy people. We consider dermatitis a problem with all people in the core room.

H. W. DIETERT: 8 Has the author succeeded in baking core oiland resin-bonded cores in the same oven? Does he favor use of bentonite or clay substance in the core sand mixture when resin

binder is used?

Mr. Greenlee: We have placed both resin-bonded cores and oil-bonded cores in the same oven without impairing the physical properties of either type of core. However, core size must be watched closely. Clay substances in general are not desirable in most core sand mixtures. Physical properties of a sand mixture containing clay, with the exception of green compressive trength and hot strength, are generally lower.

MR. DIETERT: I was wondering whether or not you used any bank sands with your resin bond, where the clay substance would

be very low?

MR. GREENLEE: At the present time, we are not using bank sand in any resin-bonded core mixture. It is all Michigan dune sand with an A.F.S. No. 50 to 52. I think you will find that the clay content in this sand is exceptionally low, probably much below 0.5 per cent.

MEMBER: Does the resin bond give off fumes in the foundry

which would be bad for the personnel?

MR. GREENLEE: There is no noticeable odor during the baking of a resin-bonded core. However, there is a definite pungent odor present after a mold containing a resin core has been poured. The same is true of a mold containing an oil core. Which is the least desirable is a matter of opinion.

ERIC WELANDER: 4 Would the author care to elaborate on the

actual baking of resin-bonded sand?

MR. GREENLEE: I am not in a position to state exactly what occurs. However, phenolic resins, when baked at temperatures as low as 300 F, transform rapidly into bakelite. At least that is my understanding. I believe this rapid transformation accounts for the savings in baking time compared to the baking of an oil core. It is not an oxidation process.

MR. GREENLEE: The release agent should be added after the

addition of the resin. In this manner sand stickiness is reduced

7 Metallurgist, Archer-Daniels-Midland Co., Minneapolis.

MR. WELANDER: I suppose much of your baking time is consumed in driving off the water after that.

Mr. Greenlee: As soon as the moisture is driven out of the core, it is only a matter of minutes, perhaps seconds, before the core is completely baked. No oxygen is required to bake a resinbonded core.

CHAIRMAN ZIRZOW: In other words, it is actually a case of driving off not only the moisture in the core sand mixture, but the moisture in the resin itself.

J. R. Somers: 5 In your process of making crankshafts in a dry sand stack mold do you use resin binders or do you use core oils?

MR. GREENLEE: Both resin and core oil are used as binders in making dry sand cores for crankshafts. Some cores are made of resin, others of oil.

W. G. FERRELL: 4 The author did not discuss the durability or the life of resin binders. Life of resin binders is measured in weeks, not months. They must be kept or stored in compara-

tively cool rooms or storage places.

In regard to dermatitis, it is our experience that colored employees have no trouble with dermatitis resulting from handling of resin binders. Lighter skinned, especially red-headed employees do get it. It is important to remember that a man can get dermatitis working at a sand mixer when the air is hot and humid in the room. Such a case can be difficult to clear up. We had a case like that. The man was confined to a small space while working at the sand mixer. It will take months to clear up the condition.

CHAIRMAN ZIRZOW: The author discussed cereal:water ratios. I believe he said that if you increased the cereal content to 2 per cent that you doubled the strength of the core mixture. We know that to be rue. Then is it not a case of obtaining most of the strength from the cereal binder rather than from the resin? What is actually happening in the use of phenolic resins is water proofing the cereals so we obtain the maximum amount of strength from the cereal rather than from the resin.

MR. GREENLEE: To a certain extent that is correct. Many foundrymen failed to take into account the tremendous amount of tensile strength which can be gained by properly blending the correct amount of water with cereal. Practically in no case can we add enough water to completely activate all the cereal. I would say if we have a sand mixture containing 11/2 per cent cereal we would have to add 7 or 8 per cent water to completely activate all the cereal. Although this sand would not be usable for making cores, it would make a very hard core possessing very high tensile properties.

J. D. Johnson: 7 We performed many experiments on water: cereal ratios in trying to get the maximum green properties and maximum baked properties from cereal in sand. We find that we can get near maximum or maximum green strength at the

low ratio of 2 parts of water to 1 part of cereal.

The maximum baked strength is obtained at approximately 4 to 5 parts of water to 1 part cereal. In other words, if you are going to use 11/2 per cent cereal in your application, you must use 6 to 71/2 per cent water to get maximum baked strength. That represents, as you know, a very sticky, gummy core sand mixture. You will lose most of your green properties by going to that high cereal:water ratio to gain baked strength.

CHAIRMAN ZIRZOW: In other words, it resolves itself into whether you are adding cereal for green strength or dry

to a minimum.

The author stated that when mixing the sand, you add all the dry ingredients first. There is a difference of opinion on that procedure. We mix the sand long enough with water so that we get even distribution of the cereal in the mixture. That phase of the mixing with water is most important in coarse sand mixing, even more so than after the oil has been added because the more you add oil the more you destroy some of the green strength. Inasmuch as you are adding cereal in the mixture primarily for green strength, keep your mixing with the oil down to a minimum. At what time do you add your release agents to the mixture?

Director of Foundry Research, Archer-Daniels-Midland Co., Minneapolis.

Metallurgist, Lynchburg Foundry Co., Lynchburg, Va.
 President, Harry W. Dietert Co., Detroit.

⁵ Chemical Div., General Electric Co., Detroit. ⁶ V. P. In Charge of Mfg., Auto Specialties Mfg. Co., St. Joseph, Mich.

⁴ Metallurgist, Union Malleable Iron Works, E. Moline, Ill.

ICE

on.

the

the

sin-

of

but

n a

use

lers

ade

or

in

m-

ing

m-

an

ot

ip

all

ths

OS.

2

re.

ng

m lic m in. ny nt ng se al. nt n be S T id at 16 ly u NI h. d g 0

II n at at ac d

h

CHAIRMAN ZIRZOW: In other words, you suggest that you add kerosene, stearic napthenic or oleic acid immediately after your resin addition or your oil addition.

MR. GREENLEE: That is correct.

CHAIRMAN ZIRZOW: I would agree with you as far as stickiness is concerned if you were trying to, in the case of kerosene or light fuel oils, to dilute your oil and get a better spread of it. I think your release agent should not be added until about 10 or 20 sec of the mixing time to give you best results.

MEMBER: The author stated that he bakes phenolic resin cores at 550 F. Does surface hardness deteriorate when the cores are

baked at that temperature?

MR. GREENLEE: If cores are properly baked at 500-550 F we will have cores with very hard surfaces. However, cereal burnout is drastic at these temperatures. Unless we have a cycle which is closely watched and controlled, edges on the cores will be soft. This, I believe, is primarily due to the fact that the cores have been overbaked and cereal burnout has occurred.

Mr. Myers: When baking at 500 or 550 F, we should keep in mind core temperature and oven temperature. Time is the critical factor. If you put a core in an oven at 550 F perhaps at the end of 10 or 15 min or even 30 min, depending on the size of the core and the amount of moisture in the core, the core has not approached the oven temperature. For that reason, if you control the temperature of the core, it will be better, than just controlling the temperature of the oven.

There are pills on the market which you can imbed in cores

and determine the temperature exactly.

Clays are going to be used more and more rather than less and less in core sand mixtures. We are finding out that there is more to green strength of sand than compression tests. By that I mean an impact and deformation tester that can be procured today, will tell you that green strength is only half of the

story in a green core mixture. You must consider the impact properties, the sagability of the core, the ability of the core to sustain an overhang.

Mr. DIETERT: Where oil-bonded cores are baked simultaneously with resin-bonded cores, I feel it would be advantageous to add a drier to the oil-bonded cores so that they may be baked more uniformly. Speeding up of baking of the oil-bonded cores will help in getting a uniform bake on the oil-bonded cores and not hurt the resin-bonded cores with a slight overbake.

Mr. DIETERT: Does the author apply parting materials to the core boxes to improve lift by an addition of a release agent?

MR. GREENLEE: No. When first using a core box which previously has been used for making oil sand cores, this box must first be completely cleaned of all oil film. This may be done by washing with carbon tetrachloride or any other solvent.

There are several points which I have neglected to mention. Resins do seem to lower the green compressive strength of core sand mixtures more drastically than do most core oils. This means that if you deem it necessary to hold a certain green compressive standard it will be necessary to increase the addition of cereal, in the core sand mixture. I also failed to mention that a core made with a release agent will have slightly softer edges. This is somewhat of a disadvantage. This, however, can be overcome by spraying the green core with water or a solution of water and resin.

Mr. DIETERT: I am glad Mr. Greenlee mentioned cleanliness of the core box because that is one of the paramount steps we have to take when we use resin binders.

A resin that imparts greater green strength than oil has been developed. It is a simple chemical that you can add to increase green strength of your core mixture.

EQUIPMENT AND METHODS OF STRAIGHTENING AND DIMENSIONAL INSPECTION OF MALLEABLE IRON CASTINGS

By Leslie N. Schuman*

THE PHENOMENAL STRIDES of American industry in the matter of high speed production and its associated cost reduction coupled with improved quality has required that our industry strive constantly to improve our product. Were we to stand idly by we would see many conversions to competitive materials that would be a detriment to the foundry industry.

The Foundry Industry through close cooperation with customers has traveled a long road of improved materials and methods resulting in great savings in both manpower and materials for our customers. There were many conversions to malleable iron from competing materials with resultant savings in weight, cost and product improvement.

It is not too long ago that the present day functions of the Straightening and Inspection Departments were handled by either the Shipping Room or the Sorting Room.

Later as customers required a casting a little closer to their blueprint dimensions the use of wood gages or sheet metal gages made to scale dimensions was practiced. There were definitely of questionable accuracy. This brought about the development of crude cast dies, in many cases made from plaster patterns poured on the production pattern, cast of soft steel and run on drop hammers or hydraulic presses of rather low capacities compared to today's 500 to 2,000 ton hydraulic presses. Other castings were hand straightened or "custom straightened" by local application of pressure at what appeared to be the warped area.

Further inspection and straightening improvement indicated a great need for more accurate pattern work particularly wherein duplicate pattern equipment was used. In many cases it was found that certain contours that did not affect the utility value of the product and were rather optional insofar as the blueprint was concerned became important to the effectiveness of the straightening dies. This brought about a change from the old scale, caliper and combination square bevel protractor to micrometer, vernier bevel protractor or sine bar in the Pattern Shop.

Today all previous concepts seem to be outmoded by the improvements in straightening and inspection. That the new thought is beneficial to our customers is best indicated by requests for quotations asking for price on "Pattern, Die and Gage" whereas it used to be "Pattern Cost." Gages have developed to be exact duplicates of customers first operation setup with suf-

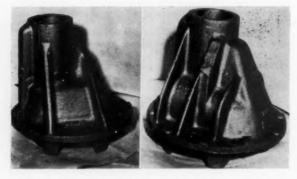


Fig. 1 (Left) and Fig. 2 (Right)—Typical passenger car differentials.

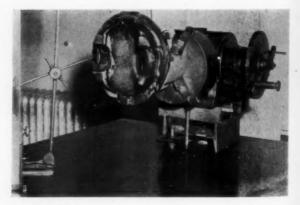


Fig. 3-Carrier in dividing head ready for layout.

Plant Supt., National Malleable & Steel Castings Co., Cleveland, O.

rns eel of 000 nd olied nt rk nt in he ievea on

el

n.

ers

or

to ct if-

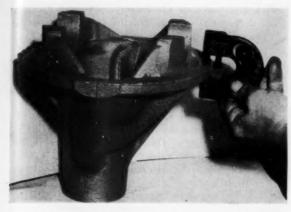


Fig. 4-Adjustable snap gage as used on flange.

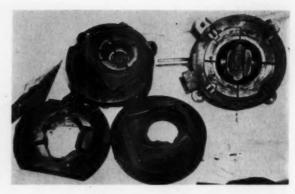


Fig. 9-Opened bottom die.

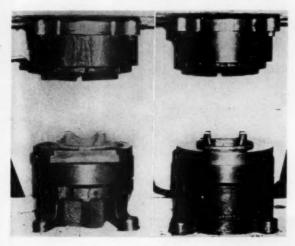
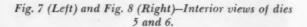
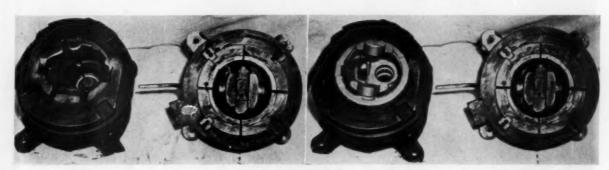


Fig. 5 (Left) and Fig. 6 (Right)—Typical differential carrier dies as mounted on hydraulic press.



Fig. 10-Differential carrier gage.





ficient tolerance targets and limit gages to assure detecting any dimensional deviations that would result in machine scrap.

As an example of the functions of Straightening and Dimensional Inspection of automotive differential carriers similar to those shown in Fig. 1 and 2, the writer will explain the procedure followed at his plant. After a check of the pattern equipment the pattern is mounted in the foundry on a regular production set-

up. Five or ten molds are produced under actual production conditions. Castings are then thoroughly shot blasted and two sets are delivered to the Inspection Department for a complete layout. The use of a milling machine dividing head equipped with a three-jaw universal chuck has proven to be a great time saver and a help to greater accuracy in the layout room by enabling a complete check in all planes from a single setup as shown in Fig. 3. Any deviations are

reported to both the Foundry Superintendent and the Pattern Shop Foreman for the necessary corrective measures. The other castings made are usually broken for a check on physical defects such as shrink and porosity. When samples satisfactory from a dimensional viewpoint are available, one set is sent for customer's approval and several sets are annealed for a trial in the straightening dies, later to be shipped for proof machining in customer's regular production equipment.

Assuming the pattern has been approved for production runs, the Hard Iron or Trimming Department periodically break a few castings each day to check for

shrink and porosity. On large running jobs such as these differential carriers, Quality Control Charts are maintained based on ten castings per shift. These charts are on such items as pinion bore diameter, pinion bore off center, flange thickness, flask shift and wall sections. All carriers are checked for cope to drag dimension by an adjustable snap gage used on the flange thickness as shown in Fig. 4. This gage is set to tolerances in accordance with customer's desires and the casting accuracy necessary to provide for proper functioning of the Straightening Dies.

After annealing, grinding, chipping and shot blasting the castings are straightened in conventional type



Fig. 11-Bearing leg template as used.



Fig. 12-"Go and No Go" gage for pinion end.

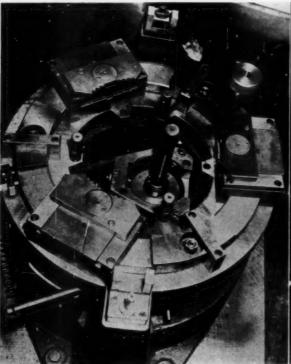


Fig. 13-View of a differential carrier proving gage.

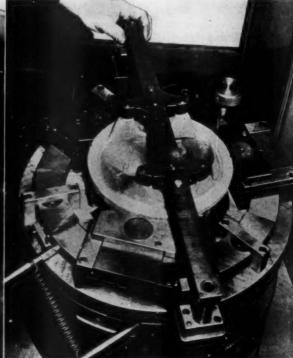


Fig. 14-Method of radial location.

NGS

are are ter, and to on e is defor

astype

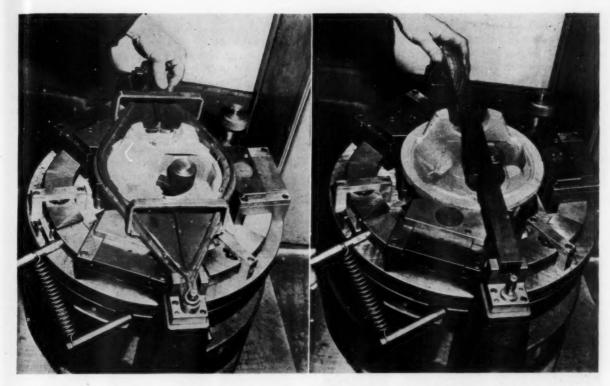


Fig. 15-Contour template for flange.

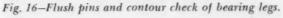




Fig. 17-Side gear seat gage for differential gear case.

hydraulic presses of 500 to 1,000 tons capacity. Output is 300 to 500 per hour depending on the travel required to load and the travel required at high pressure to close the loose piece dies used. For safety, the presses are usually provided with two control buttons so spaced that they cannot be actuated by one hand, requiring that both hands be "in the clear" to actuate press.

The dies used have loose pieces assembled inside of standardized holders as shown in Fig. 5 and 6. The top die has contracting loose pieces to grasp the outside diameter of the flange and the outer section of the cross bore bearing legs and expanding loose pieces that contact at the inner side of cross bore bearing

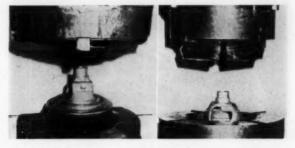


Fig. 18 (Left) and Fig. 19 (Right)—Typical differential gear case dies.

legs as shown in Fig. 7 and 8. The bottom die shown in Fig. 7 and 8 is composed of a holder for the contracting loose pieces to grasp the outside diameter of the pinion bore or stem as we call it. Above this is a combination spacer and stop for bottom loose pieces on top of which is a pressure plate as shown in Fig. 9. This provides for easy maintenance of dies and replacement of worn die pieces without complete die replacement.

Each time a die is mounted, or every 8 hr, whichever is first, 100 pieces are gaged in a fixture that duplicates customer's first operation (Fig. 10). The points gaged are cross bore bearing legs by means of a template as shown in Fig. 11, clearance for pinion boring bar and the pinion bore end or stem as we call it. The pinion end is gaged with a "go and no go"

rotating finger gage calibrated to assure sufficient wall section after machining. This target and its method of use is shown in Fig. 12. This type gage plus a "go and no go" for flange diameter has proven quite satisfactory for normal production operation control.

There has been some tendency to elaborate on inspection work by more elaborate proving gages such as is shown in Fig. 13, 14, 15 and 16. However, from a production standpoint, their use is entirely too time consuming. It is possible that to provide a check for quality control charts they may serve a worthwhile investment.

One-piece differential gear cases of the type shown in Fig. 17 follow the same checking and proving as previously described as being used on differential carriers. Figure 17 shows the method used to check the position of the core up and down as in the mold. It will also be noted that a flange thickness gage is included with the core set gage.

The straightening of gear cases may be somewhat simplified if the flange outside diameter is in that part of the mold opposite the body of the casting. Figure 18 shows a die that has proven quite satisfactory. The entire effort to remove shift and runout being by the top loose pieces. This type die is rather free from the dirt trouble encountered when loose pieces are at the very bottom of a die. Figure 19 shows a further effort to guide top die into bottom die.

Inspection, as on the carrier, includes a periodic 100 per cent gage operation in a fixture similar to customer's first operation. A rotating finger gage as shown in Fig. 20 is used to check eccentricity of the



Fig. 20-Inspection fixture for gear case.

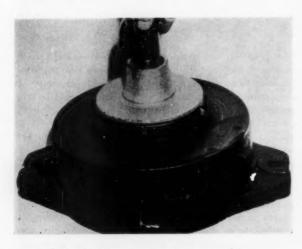


Fig. 21-Hub and gage for eccentricity.

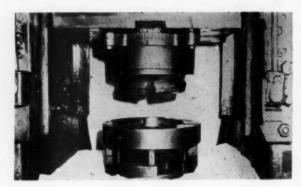


Fig. 22-Loose piece hub die mounted on press.

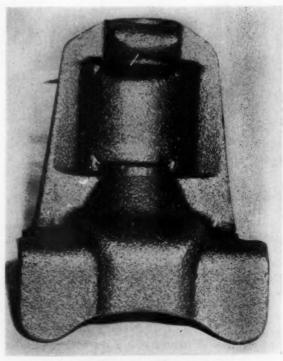
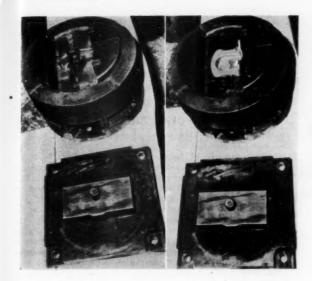


Fig. 23-Truck spring hanger.

d.
nat
re
ne
ne
ne
ne

ic

e



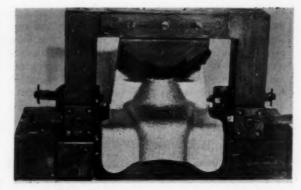
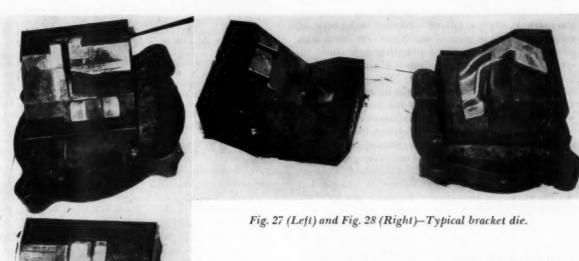


Fig. 26-Gage used on truck spring hanger.

Fig. 24 (Left) and Fig. 25 (Right) - Views of truck spring hanger die, with and without casting.



end opposite the chuck jaws.

Single flange passenger car hubs follow the usual pattern approval and proof procedure previously outlined and due to the small amount of permissible "runout" or shift are checked closely in the Trimming Room for this condition as shown in Fig. 21. Experience has indicated that a loose piece die of the type shown in Fig. 22 will remove about one-half of the runout of a shifted casting. Here again as on the differential gear case the top loose pieces contact both cope and drag of the casting. Periodic checks are made at the press to assure that the die is functioning properly and any rejections for "runout" require immediate corrective measures.

The truck spring hanger shown in Fig. 23 is one that requires straightening in all planes to tolerances of about 0.050 in. and the frame contact areas are to be coined to about 0.015 of flat and true 90-degree angle. It will be noted from Fig. 24 and 25 that loose pieces contact from all sides. After breaking down the contact areas of this type die repeatedly it was concluded that due to normal molding variations and the close fit that it was necessary to either mold to much closer tolerances than were practical or to induce our customer to allow some part of this casting to have the accumulated error from the flowing of excess metal in the pressing operation. Allowance for the displaced metal has corrected the breakdown condition considerably. The gage used is shown in Fig. 26, the numerous targets and flush pins being self evident.

Figures 27 and 28 are of a solid type bracket die with replaceable inserts. This type die is of the simple type and when it can be used provides one of the less troublesome types in current use.

The materials used in the dies mentioned vary from low-carbon steel carburized to high-carbon alloy tool steels, the selection of the material being based on the application. Torsion, compression, tension, shock and desired hardness being some of the considerations. The various angles used on the loose pieces are varied depending upon the pressure desired and the amount of effective travel desired. Serrations are sometimes used to provide for flow of metal on heavily coined areas. A large percentage of the dies are constructed in the plant, a small percentage being farmed out to local tool shops. Design is entirely by plant die engineers working under the supervision of the Finishing Superintendent.

The gages used are developed after either the Chief Inspector or one of his assistants has visited the customer's plant and discussed the job requirements and locating point of the casting involved after which a gage is developed to adequately gage all essential dimensions. All gaging after anneal is in close proximity to presses to minimize handling and transport, the number of inspectors per job being geared to maximum press output.

The responsibility for the functioning of the operations are briefly as follows: Layout and check of samples, design of gages, supervision of hard iron trim and inspection, soft iron gaging and visual inspection are under the jurisdiction of the plant Chief Inspector who works under the Manager of Inspection reporting to the Vice-President of Operations. Die design and maintenance is the responsibility of the Straightening Foreman who reports any poorly functioning die to the Die Supervisor for the necessary corrective steps. The Straightening Foreman is under the jurisdiction of the Plant Manager.

This paper is merely a review of the practice used at the author's plant on a few of the many castings produced. There is no claim that this is the only good method, that the designs mentioned are the ultimate of perfection, or of our own design, as many of them contain features used by others in the industry and some of the features are suggestions of our customers."

DISCUSSION

Chairman: James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman: H. W. Johnson, Wells Mfg. Co., Skokie, Ill. H. C. Stone: I I was interested in the drift the author showed in the bottom of the differential die. Is that a tight fitting die? Charles Schneider: That drift has at various times been made in various ways. At one time it was a one-piece drift. The difficulty with that was that it either goes into the opening too much and cannot be released, or it does not do any good. The most successful one that has been in use is a plumb-bob type of

round up the inside of the pinion bore, and when the pressure is released they collapse and the casting is easily removed.

MR. STONE: Is that drift also used to get the center line of the flange?

drift that has expanding pieces. The expanding pieces serve to

MR. SCHNEIDER: Yes, although it is primarily to round up the stem if it is out-of-round.

Au

i.e.

cas

sol

lo

an sti of ge su fo ch in

bi va sin of Fi an no th

lui

Mr. STONE: The variations you can expect because of a swelled area are also to be faced with the problem of casting variations.

MR. SCHNEIDER: That is a difficult problem. There is no question about it. If the castings have excessive variations in the essential dimensions that you are trying to press, there is very definite hardship in the maintenance of the dies. They cannot be expected to stand up under too severe variations. However when the variations are normal the die has to be able to take care of them.

Asst. Works Mgr., Belle City Malleable Iron Co., Racine, Wis.

² Asst. Chief Inspector, National Malleable & Steel Castings Co., Cleveland.

SOLIDIFICATION OF GRAY IRON IN SAND MOLDS

By
R. P. Dunphy* and W. S. Pellini**

ABSTRACT

Thermal studies of the solidification of gray irons from sand walls have shown that solidification proceeds in a wave-like fashion by the travel of "start" and "end of freeze" waves. Austenite dendrite "start" and "end" waves travel in sequence, i.e. the "start" wave completes its travel to the center of the casting before the "end" wave begins thus creating a semi-solid condition through the entire casting. Eulectic formation proceeds by the travel of "start" and "end" waves in conjunction; i.e., as a narrow band. In each case the "start" waves move through isothermal liquid. The effects of superheat, carbon equivalent and sea coal have been shown to modify the timing and rate of progression of the freezing waves.

Introduction

GRAY IRON has long been recognized as anomalous in its response to risering and solidification control practices which are applicable to other ferrous alloys. It has been suggested that the anomalous behavior of gray iron is related to a complex solidification mechanism involving a wide freezing range and the formation of proeutectic and eutectic constituents of widely different densities. The results of solidification studies of gray and other cast irons generally have been discouraging and have failed to supply reliable data as to the effect of the usual foundry variables of pouring temperature, metal chemistry, section size and mold composition. Bleeding tests have been used exclusively in these studies.

Yearley, Schauss and Martin¹ measured the effect of pouring temperature, sand variables and section size on the solidification rates of 1½ in. and 3 in. square blocks of malleable iron by bleeding the castings at various times after casting. Schwartz² conducted a similar series of pour-out tests with the added feature of thermal measurements at the center of the casting. Fifield and Schaum³ followed a similar procedure and found that bleeding of gray iron castings could not be accomplished after the casting had cooled past the liquidus temperature due to the fact that bridging dendrites entrapped the remaining liquid and prevented this liquid from draining. This finding

suggested that bleeding techniques were not adaptable to measuring rates of solidification of cast irons and cast considerable doubt on the validity of the pour-out technique. Ruddle⁴ in his recent review of solidification concluded that the pour-out method is virtually useless for alloys having a wide solidification range. Ruddle based his conclusions on the results of Schwartz and Fifield and Schaum and also from his own studies of the solidification of nonferrous alloys.

A discussion of the various indirect methods of solidification studies, such as mathematical computation and the electrical analogue used by Paschkis, has been presented in a companion paper⁵ and is therefore not repeated herein.

The successful use of thermal analysis techniques for solidification studies of nonferrous metals by Ruddle and independently by this laboratory in the study of the solidification of steel prompted an extension of this method to gray iron. The present investigation is part of a comprehensive study of the solidification of metals. These studies are being conducted by the NRL with the aim of providing basic information required for the solution of practical solidification problems.

Experimental Procedure

The gray irons used in this investigation (Table 1) were prepared synthetically by induction-melting charges of Armco iron, graphite, ferrosilicon (50% type), and ferromanganese in amounts calculated to give the desired composition. All the irons were inoculated with 0.50 per cent silicon as ferrosilicon (50%) by tapping onto the inoculant. All the heats were superheated to 150 F (measured by immersion couples) above the desired pouring temperature and allowed to cool to the pouring temperature in the ladle. The sand used was a synthetically-bonded sand of A.F.S. Fineness No. 80 containing 3 per cent bentonite, 11/2 per cent dextrine, 1/2 per cent mogul, and 31/2 per cent water. In two cases in which 6 per cent seacoal additions were made the seacoal was added to this base sand.

The test castings measured 30 in, high and 7 x 7 in, in cross-section. Details of the casting and of the

NGS

at

ate em nd

ine

lie? een

he too he of

ire

of

he

ng

no

^{*} Metallurgist and ** Head, Metal Processing Branch, Metallurgy Division, Naval Research Laboratory, Washington, D. C.

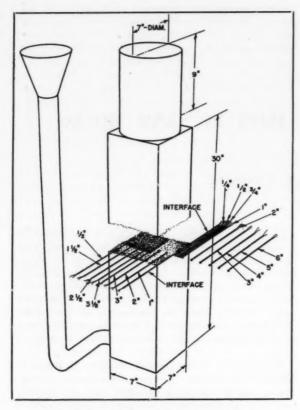


Fig. 1-Details of casting and location of thermocouples.

gating and risering methods are shown in Fig. 1. The sand was molded in a 24 in. square flask which provided a depth of 81/2 in. of sand on all four faces of the castings.

The course of solidification of the test castings was followed by observing the temperature changes through the thickness of the casting at its vertical midpoint. A series of Pt-Pt Rh 13% thermocouples were cantilevered from the mold wall to the center of the casting according to the arrangement shown in Fig. 1. The couples were enclosed in alundum insulators and sheathed in 6 mm o.d. 3 mm i.d. fused silica protection tubes. The metal interface temperatures were recorded by placing a sheathed thermocouple at the interface so that one-half of the pro-



Fig. 2-Photo of general test set-up.

tection tube was in the metal and the other half in the sand. An additional bare couple was placed so that the wires were entirely in the sand and the bare bead at the casting surface. The bead was coated with a thin layer of silica flour mold wash. This couple has been designated as the mold interface couple.

The course of heating of the mold wall was followed by means of chromel-alumel thermocouples placed in the same horizontal plane as the couples which measured metal temperatures. The spacing arrangement is indicated in Fig. 1. The metal temperatures were recorded on a multipoint, automatic recording potentiometer, 1500 to 3000 F scale with an accuracy of 0.25 per cent of full scale and sensitivity of 0.1 per cent. The sand temperatures were recorded on a similar instrument, 0 to 2500 F scale. The temperature recording sequence was arranged so that readings from individual couples were obtained approximately every 8 sec. The general arrangement of the mold and recording instruments is shown in Fig. 2.

Interpretation of Thermal Data

Detailed cooling curves for a series of tests which show the thermal effects of varying carbon equivalent are presented in Fig. 3 and 4. Such detailed plots are not presented for the other tests of this investigation inasmuch as they differed only in minor details. The data from one of the cooling curves (C-13) have

TABLE 1

Heat No.	С	Mn	Si	P	s	Ni	Cu	Мо	Cr	Ti	v	Carb. Equiv.	Liq. Temp. F	Pour. Temp. F	Temp.
C-1	2.99	0.56	1.73	0.005	0.041	0.09	0.09	0.01	0.05	0.01		3.57	2275	2600	2105
C-2	3.63	0.54	1.50	0.007	0.037	0.05	0.06	0.01	0.06	0.01		4.13	2134	2500	2114
C-7	2.93	0.63	1.70	0.015	0.033	0.05	0.10	0.01	0.05			3.50	2280	2500	2115
C-8	3.39	0.51	1.41	0.019	0.034	0.02	0.03	0.01	0.10			3.86	2215	2300	2110
C-9	3.74	0.42	1.05	0.011	0.033	0.05	0.08	0.01	0.05			4.09	2165	2295	2112
C-10	3.91	0.50	1.33	0.014	0.033	0.04	0.07	0.01	0.04			4.45		2300	2110
C-12	3.66	0.52	1.44	0.017	0.041	0.02	0.04	0.01	0.02			4.14	2158	2300	2119
C-13	3.04	0.53	1.39	0.014	0.046	0.02	0.05	0.01	0.03			3.50	2286	2500	2109
C-14	3.30	0.55	1.69	0.017	0.035	0.02	0.04	0.01	0.02			3.86	2215	2500	2105

n so re h

es es es

ic

h

i-

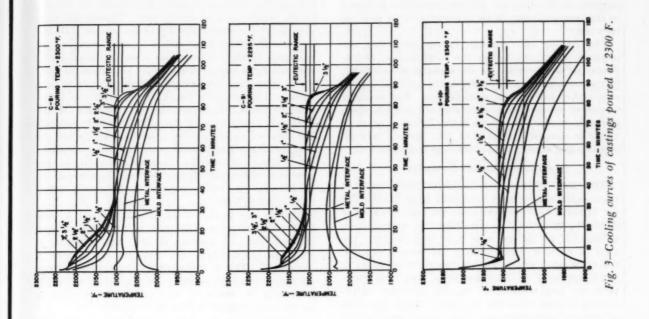
e

e. d

r-

h

ts



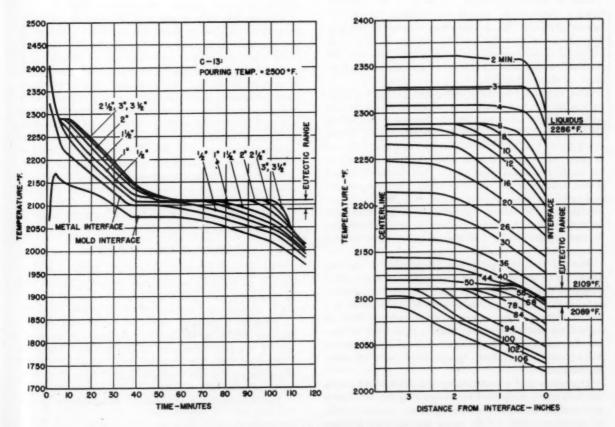


Fig. 4-Cooling and gradient curves of casting C-13 poured at 2500 F.

been replotted as thermal gradients in Fig. 4 for the purpose of detailing the general procedure followed in the interpretation of the thermal data of all tests

reported.

The cooling and gradient curves show that the sequence of events during the solidification of hypoeutectic irons consists of the removal of superheat, formation and growth of dendrites, and the beginning and end of eutectic solidification. The sequence is shown by the times at which the various space positions in the casting reach the liquidus (start of austenite dendrite formation), the eutectic start-of-freezing and eutectic end-of-freezing temperatures.

Austenite Dendrite Formation

The start of austenite dendrite formation was taken as the time when a significant drop (2 F) below the liquidus hold was observed; the end of austenite dendrite formation, as the time an approach to within 2F of the eutectic hold was reached; the start of eutectic formation, as the time a drop to 2 F below the eutectic hold occurred. The choice of a 2F approach or deviation from the "hold" temperatures was necessitated by the difficulty of observing a point of merger for curves which approach or deviate asymptotically. The errors involved are small and of no practical significance. The end of eutectic formation was taken as a "technical solidus" temperature set at 20 F below the eutectic hold. The choice of this temperature in favor of an absolute solidus temperature was made because of the inherent difficulty of determining thermally the disappearance of the very last films of liquid. The 20 F range of eutectic solidification was based on the best available information for irons of the silicon contents used in these studies.6

The method of taking the point of final solidification as the temperature at which delayed cooling was no longer observed on the inverse rate cooling curve of the central couple (the method used in the companion investigation of steels) ⁵ was not followed because of microstructural and thermal indications of marked undercooling of the last pools of liquid which solidify at the center section. The temperature indicated by this method would thus not be representative of the solidus temperature of the remainder of the casting. Actually, the temperature differences in end points involved in the choice of these various temperatures are small.

For all practical purposes the choice of the end point based on the end of the eutectic solidification range is justifiable as the point at which technical solidity is attained; i.e., as a point which at most will be in error to the extent that remaining films of liquid are not recognized. Such a procedure is commonly resorted to in the study of reaction kinetics. S-curve studies similarly ignore last traces of transformation.

Discussion and visualizations of solidification events are very much simplified by recognizing the wave-form progression through the casting of the various cycles of transformation which occur during solidification.

The first event to be noted from the thermal curves of Fig. 3 and 4 is the rapid elimination of

superheat during which time the thermal gradients through the casting except in the immediate vicinity of the surfaces are relatively mild. During the time superheat is removed the liquid in contact with the mold wall is cooled below the liquidus temperature resulting in the formation of austenite dendrites at the casting surfaces. The initial formation of dendrites represents the initiation of the "dendrite start wave." The necessity for the removal of the latent heat of fusion accompanying the formation of the dendrites produces a thermal block* which prevents the interior of the casting from cooling below the liquidus temperature until reached by the "dendrite start" wave. Essentially isothermal conditions are therefore developed throughout the liquid portion of the casting. This condition is frequently called the liquidus "hold."

During the period of superheat removal a solid skin is in no sense developed at the mold surface. Actually only a fraction of the total amount of dendritic solid which can form at the surface is developed. This is indicated by the gradient curves of Fig. 4 which show that at 6 min when all of the superheat is dissipated, the surface temperature is

as yet only slightly below the liquidus.

Following the removal of superheat a wave-like progression of initial freezing moves through the casting. It is noted that initial freezing extends to the center of the casting in approximately 11 min at which time the surface temperature indicates that approximately half of the total amount of austenite dendrites which may form has actually solidified at the surface. It is not until approximately 36 min have elapsed that the formation of the dendrites is completed at the surface. This point represents the development of the "dendrite end" wave which in turn progresses gradually through the casting. It is thus evident that a mushy condition is developed through the entire casting consisting of a lacework of austenite dendrites in liquid. Solidification is progressive in a sense that the amount of solid is higher nearer the surface. A solid skin is, however, not developed during the greater part of the austenite dendrite formation period. The mechanism of solidification during this period is similar to that shown for the solidification of steels against sand walls in the companion investigation5.

Growth of Austenite Dendrites

The growth of the austenite dendrites results in carbon enrichment of the interdendritic liquid and both growth and enrichment continue until the eutectic temperature is reached. It is noted from the 36 to 50-min gradient curves of Fig. 4 that eutectic freezing occurs at the surface immediately after the end of dendrite formation at the surface. Thus, for a short distance the "dendrite end" wave is followed immediately by the "eutectic start" wave and both waves moving out togeth.".

The solidification of eutectic which occurs at po-

th

^{*}Ref. (5) gives a detailed description of the mechanism by which a thermal block is developed.

DS

nts ity ne he

at

nt

he its he

on ed

id ce. of

is

res

he

is ke

st-

he at

at

at in is

he in

is ed rk is

is er,

ite

divn

in

in

nd

u-

he

ic

he

or

ed

th

0

by

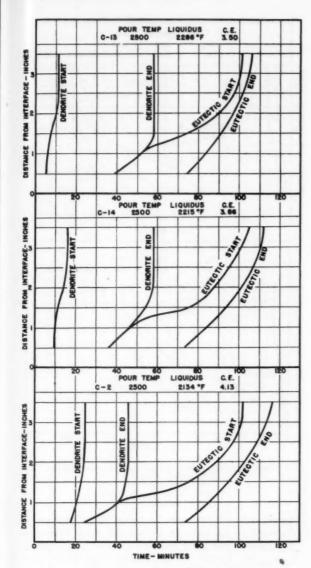


Fig. 5—Progression of solidification for castings poured at 2500 F.

sitions close to the surface following the passing of the "eutectic start" wave provides a condition of thermal block, since the temperature at the casting surface cannot drop below the eutectic range (2109 to 2089 F) until all of the eutectic liquid at this point is solidified. The time delay required for solidification of the near-surface eutectic slows down the initially rapid progression of the "eutectic start" wave drastically.

The thermal block occasioned by eutectic reaction cannot, however, prevent the completion of dendrite formation since this reaction occurs at temperatures above the range of eutectic formation. Thus, the "dendrite end" wave continues to completion rendering the central portion of the casting isothermal at the eutectic reaction temperature. During this interval the "dendrite end" wave therefore moves away from the "start of eutectic" wave to complete its cycle

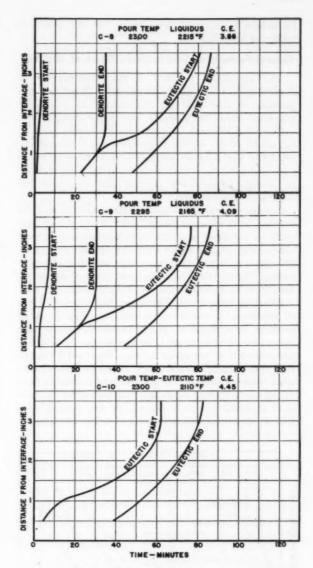


Fig. 6—Progression of solidification for castings poured at 2300 F.

to the center of the casting. The travel of the "end of dendrite" wave at this stage is rapid because of the advanced state of dendrite formation throughout the casting.

The advanced state of dendrite formation is indicated by the mildness of the gradients which exist through the central region of the casting at the time (50 min) of separation of the "dendrite end" and "eutectic start" waves.

Following the completion of dendrite formation eutectic solidification gradually proceeds through the casting. Eutectic formation does not occur over the entire casting as in the case of dendrite formation but proceeds instead as a solidification band; i.e. with travel of "start" and "end" waves in conjunction. The width of this band is approximately 3/4 in. The narrow temperature range of eutectic formation in contrast to the broad temperature range of dendrite

formation is believed responsible for the different modes of solidification.

General Features of Solidification-Effect of Variables

The significant features of the solidification of irons which have been discussed in detail in the previous section may be represented in a concise graphical form by plotting the progression of the various solidification waves from the mold wall to the center of the casting. Figures 5 and 6 present the results of various tests plotted in this fashion. These tests include a variety of irons ranging from eutectic to strongly hypoeutectic poured from high and low superheat temperatures.

Certain general solidification characteristics of these various irons deserve recognition. It is noted that each of the four separate solidification waves have remarkably similar shape and rate of travel in all of the tests. It is apparent from this that the effects of carbon equivalent and superheat are primarily reflected in time displacements of the waves.

Effect of Superheat—The time displacement effect of superheat is readily apparent from Fig. 7 which presents superimposed plots for four hypoeutectic irons poured at four levels of superheat. Casting C-8 poured with a minimum of superheat (85 F) shows that the "dendrite start" wave develops immediately after pouring without significant delay. Casting C-14 of the same carbon equivalent, as indicated by identical liquidus temperatures, which was poured with 285 F superheat, shows a time delay of approxi-

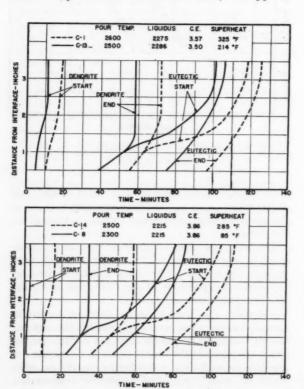


Fig. 7-Progression of solidification as influenced by superheat.

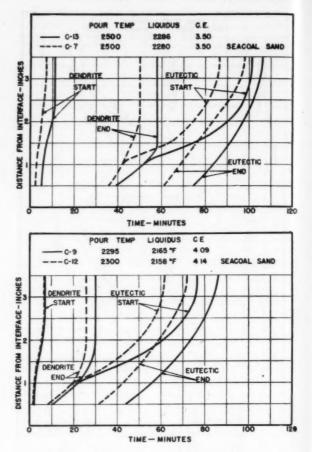


Fig. 8-Effect of sea coal on progression of solidification.

mately 10 min for the "dendrite start" wave. As discussed in the previous section this delay represents the time required or the dissipation of superheat. Castings C-13 and C-1 representing hypoeutectic irons of slightly lower carbon equivalents poured with 214 and 325 F superheat respectively indicate similar delay effects.

The effect of the initial delay of the "dendrite start" wave is reflected in time displacements of each of the subsequent solidification waves. The time displacement of the following waves is, however, 2 to 3 times greater than the initial time displacement of the first solidification wave indicating a general increase in the total times involved in the travel of the waves across the casting. Thus, the final solidification time is increased considerably more than would be expected from a single initial delay in the start of solidification.

The effect of superheat in introducing an initial delay to the start of solidification of irons is exactly analogous to the effect of superheat on the solidification of steels⁵. The effect of increasing the time for completion of solidification is likewise shown for steel. Basically, these related effects may be traced to the initial heating of the mold occasioned by the flow of superheat, so that solidification proceeds not

LDS

As its at. ns 14 ay ite ch ne 2 nt nhe on be of al ly fi-OF ed

ot

with an initially cold mold wall, but with a mold wall preheated by superheat.

Effect of Carbon Equivalent—The relative solidification characteristics of irons of various carbon equivalents poured from commonly used (2500 F) and low (2300 F) temperature levels are shown in Fig. 5 and 6.

Analysis of these data requires a consideration of two interacting effects: (1) the effect of superheat level which has been shown to introduce an initial delay to the start of solidification and then to slow down completion of solidification and (2) the effect of varying the relative amounts of the proeutectic constituents of the casting by variations of composition.

It is noted that the time interval between dendrite "start" and "end" curves is reduced as the total amount of the dendrite phase is decreased. The time required for eutectic solidification at any given position, however, increases with decrease in dendrite solidification time. The total time elapsed between start of solidification and final freezing at any given position in the casting does not vary significantly with variation in composition. Thus the decreased time required for the dendrite formation of near-eutectic irons is almost exactly made up by somewhat longer times for eutectic solidification. The effect on final solidification times resulting from pouring irons of various carbon equivalents from a fixed temperature is therefore primarily due to superheat delay. At low pouring temperature levels (the 2300 F series) the effect of superheat occasioned by variations of carbon equivalent is small and it is noted that complete

solidification is reached in approximately the same time by the eutectic and the two hypoeutectic irons. At commonly used pouring temperature levels (the 2500 F series) the effect of superheat due to variations of carbon equivalent becomes more pronounced, i.e., an increase in final solidification time follows an increase in carbon equivalent. The total difference is, however, not large; being in the order of 10 per cent. Comparison of the 2300 F versus the 2500 F series, shows, however, that change of pouring temperature level has a marked effect. The various compositions poured at 2300 F solidify in approximately 85 min while similar range of irons poured at 2500 F require approximately 110 min to solidify, an increase of 25 per cent.

Effect of Sea Coal—The effect of sea coal is shown in Fig. 8 by comparing castings C-9 and C-12 poured at 2300 F and C-13 and C-7 poured at 2500 F. The relative positions of the solidification waves indicate that sea coal increases the rate of progression of solidification over the entire course of solidification. Comparison of the castings poured with high superheat shows superheat is eliminated rapidly in the presence of sea coal; i.e. the time delay of the start of freeze wave occasioned by superheat is approximately halved by sea coal. Comparison of the times for the formation of dendrites and eutectic solidification indicates that sea coal is effective during the entire solidification cycle in decreasing the time of solidification at all positions in the castings. The cumulative effect is a decrease in final solidification time in the order of 15 per cent. The increased rate

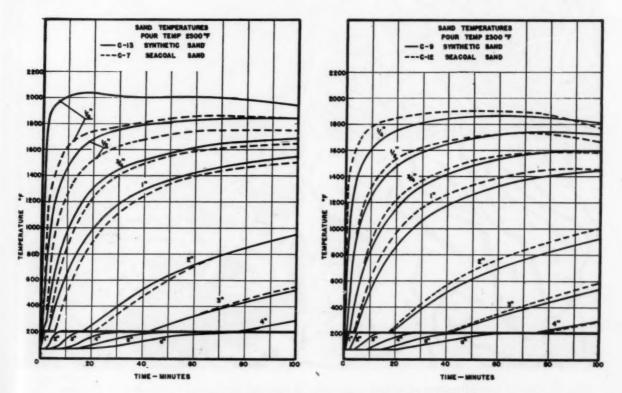


Fig. 9-Thermal course of mold walls during solidification.

of progression of solidification occasioned by sea coal must necessarily indicate increased rate of heat removal by the sea coal sand.

Thermal Course of Mold Walls

Temperature-time curves for the mold walls of castings poured at 2500 F and 2300 F in the base synthetic sand and the sea coal containing sand are presented in Fig. 9. Analysis of these data shows that sea coal sands consistently maintain temperatures lower than the base sand when high pouring temperatures are employed and an opposite effect when the pouring temperature is low. In view of the extremely limited information available on the nature of distillation and mold gas reactions occurring in the presence of sea coal it is impossible to rationalize the significance of these data in terms of the contribution of sea coal to the thermal properties of the mold. Direct evaluation of relative heat transfer rates from thermal gradient curves is not possible when the heat transfer medium undergoes complex endothermic or exothermic reactions consequent to the flow of heat by conduction. The behavior of sea coal during the solidification of castings is evidently complicated and meriting of further research.

The general features of the thermal course of the base synthetic sand during the solidification of the various irons are similar to the features which have been noted for this same sand during the solidification of steels⁵. The temperatures at corresponding distances from the interface at the same times are lower for the irons as should be expected due to the lower pouring temperature and range of solidification of irons. The depth to which the sand is heated to the 212 F water-to-steam hold at the time of complete solidification, is, however, greater for the irons. This is believed to be due primarily to the much greater times required for complete solidification. The 6 in. sand position for the four castings of Fig. 9 reaches 212 F considerably in advance of the time of final solidification of these castings (85 to 105 min). While no data were obtained for positions closer to the flask it is evident that the 212 F hold extends beyond the 6 in. position and likely approaches the flask at the time of complete solidification,

iı

fı

ii

fre

slo

Macro- and Microstructures—Sections from the casting at a plane immediately above the thermocouples were hot acid etched to determine macrostructural

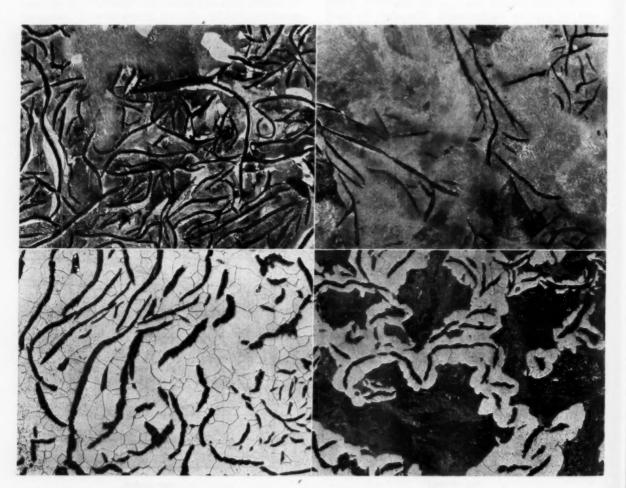


Fig. 10-Typical microstructures of surface and center of two test castings, C-9 (left) and C-13 (right). Top photos-surface: Bottom photos-center. Mag. 100x.

DS

he

he

80

fi-

ng

re

he

211

to

te

is

er

29

al

le

1.

10

10

al

characteristics. In general high superheat produced coarse dendritic structures, whereas low superheat, and the addition of sea coal produced finer structures. Ordinarily the dendrites near the surface were fine, closely spaced and approached a columnar orientation. While those in the center were fewer in number, larger and randomly oriented. As the composition approached the eutectic the dendrite pattern is lost.

The microstructure varied from Type A graphite in the castings of high carbon equivalent to Type D in the low carbon irons. The change in flake size from surface to center indicated some undercooling in the geometric center of the castings. The matrix structures were for the most part pearlitic with increasing amounts of ferrite from the surface to the center. Typical microstructures of the surface and center of one of the castings are presented in Fig. 10.

Acknowledgment

The authors wish to express their appreciation for the many helpful discussions with Dr. Marzke, Superintendent of the Division and the assistance extended to them by H. F. Bishop, F. A. Brandt, T. W. Snodgrass and E. T. Myskowski. The assistance of the Metallographic section of the Laboratory in preparing the photomicrographs is likewise appreciated.

Bibliography

1. B. C. Yearley, R. P. Schauss, P. A. Martin, A.F.S. Transactions, vol. 52, pp. 483-498 (1944).

 H. A. Schwartz, A.F.S. Transactions, vol. 55, pp. 66-67 (1947).

3. J. E. Fifield, J. H. Schaum, A.F.S. Transactions, vol 56, pp. 382-388 (1948).

4. R. W. Ruddle, "The Solidification of Castings," Institute of Metals: Monograph and Report Series No. 7 (1950).

5. H. F. Bishop, F. A. Brandt, W. S. Pellini, "Solidification of Steel Against Sand and Chill Walls," A.F.S. Transactions, vol. 59 (1951). Preprint No. 51-21.

6. E. S. Greiner, J. S. Marsh, B. Stoughton: Alloys of Iron and Silicon, Alloys of Iron Monographs, McGraw-Hill, 1933.

DISCUSSION

Chairman: H. A. SCHWARTZ, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: E. C. Troy, Foundry Engineer, Palmyra, N. J. H. A. Schwartz (Written Discussion): 1 In the paper the authors introduce us to a concept of a "heat block" in the zone where eutectic freezing is going on. This results in a concept of an isothermal plane, for example in Fig. 5, between the region where austenite is all frozen and that where eutectic freezing begins.

This commentator has some hesitancy in believing this to be truly an isothermal surface and hopes that the following comment may not be regarded as purely metaphysical.

Any point within the casting is originally above the eutectic freezing point. Presently there is formed a boundary between it and the surface where the eutectic freezing point is maintained as postulated by the authors. Heat transfer within the region of liquid eutectic is still possible until all temperature gradients vanish, which takes time, in theory infinite time.

In finite time there must remain some small gradient and our point is still, very slightly it is true, above that of the freezing eutectic.

The horizontal lines of Fig. 4 are not actually horizontal but slope minutely downward with increasing time.

The 2° F limit set by the authors as the criterion of the smallest temperature change which can be recognized was purely arbitrary and the smallest to be detected by their, or probably any other, method.

¹ Manager of Research, National Malleable & Steel Castings Co., Cleveland.

The lines of Fig. 4 are thus not curves of very short radius tangentially approaching the eutectic melting point but curves of long radius reaching the eutectic temperature near the end of what appears to be a horizontal line. The authors may have had this in mind when they referred to an asymptotic approach. Obviously a line approaching another asymptotically reaches it only at infinity.

This discussion would have no significance did it not alter Fig. 5. If it be accepted it renders meaningless the lines "Dendrite End." These are merely the arbitrary loci of the time space relation when a given temperature was passed.

Perhaps these omissions would improve the discussion as preventing the need of explaining why there must not be continued slow dendritic freezing until the eutectic begins to freeze.

The author's discussion in the third paragraph of the section "Austenite Dendrite Formation" seems to indicate an appreciation that the eutectic has frozen when the horizontal portion of the curve has ceased to exist and a dip in temperature begins.

MR. DUNPHY: Dr. Schwartz's viewpoint with respect to the existence of an isothermal plane during the liquidus and eutectic "holds" is certainly appreciated. The experimental observation that an isothermal plane does exist is of course dependent on the sensitivity of the thermometry. We recognize that an absolute isothermal condition is approached only gradually and asymptotically. Additional experiments were conducted using an automatic recording potentiometer having increased accuracy and sensitivity (0.4°F) to illustrate this point. The experiment reproduced the isothermal plane phenomenon at the eutectic temperature found with instruments of lesser accuracy. Thus, from a practical standpoint an isothermal condition is in fact developed. Deviations from an absolute isothermal condition in the order of 0.4°F or less are entirely possible.

The 2°F approach or deviation to the "hold" temperature was

The 2°F approach or deviation to the "hold" temperature was chosen arbitrarily and represents a question of absolute vs practical end points. From a physical, or solidification, standpoint this means that the curves indicate essential "completion" or "start" rather than an absolute 100 per cent in a r.anner similar to the classical methods used in "S-curve" studies of steel transformations. Use of the term, "technical end point," in the text emphasizes this fact.

V. Paschkis (Written Discussion): ² This paper is of considerable interest because of the many temperature data on the solidification of gray iron contained therein. Inasmuch as the writer is mainly interested in the possibility of computing solidification of various metals this discussion will emphasize the thermal aspects of the paper.

The authors use, as in some former publication of theirs, the term "thermal block." This term is used, it seems to the writer, for two different things: for a high thermal resistance and for a heat source at one point in the system which delays the heat flow from a different point of the system. Although some results of one "thermal block" are similar to those obtained by the other type of "thermal block," it is well to distinguish between those two types.

Looking at Fig. 3 and 4 (left half of the Fig.) it seems somewhat surprising that there is almost no temperature gradient from the center to the casting to a point $\frac{1}{2}$ in. from the interface and then quite a sharp gradient from there to the interface. Such behavior is hard to explain and raises the question of accuracy of the temperature measurements at and near the interface. Reference is made particularly to the time from 30-40 min in the case of C-8, the time from 25-45 min in case of C-9, the time from 8-12 min in the case of C-10, and the time from 40-60 min in the case of C-13.

The temperature gradient curves as shown in Fig. 4, right, can be explained only by the existence of heat sources in the metal which are active even after the temperature has dropped far below the liquidus range. In the case shown, heat sources must have been active at a distance from 3.5 in. to 2.75 in. from the interface until approximately 102 min when this region had a temperature slightly above 2100 F.

Knowledge of the heat sources (heat of fusion or heat of reaction) would be of interest from two aspects:

 It may give information on metallurgical happenings in the solidifying casting and explain crystal growth which occurs.
 This knowledge could serve as essential basis for later cal-

² Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York.

culations of solidification of gray iron.

The strength of the heat source may well be a function of temperature. Curves as shown in the paper could be used to determine on the electric analog computer the strength of the heat sources as function of temperature, provided the following data are known: thermal properties (conductivity and specific heat) of the solid and the liquid iron and the thermal properties of sand.

In the section on "Effect of Superheat" the term "superheat" is used apparently to indicate the temperature difference between the pouring temperature and the liquidus. But in the first section, on "Experimental Procedure" it is said that the iron was "superheated" and this time the term refers to a temporary raising of temperature with a subsequent cooling prior to pouring. It would be well to state clearly the difference between those two meanings, and preferably to use different terms.

The authors state in their section on "Temperature Distribution in the Mold Wall" (they speak of "thermal course of mold wails") of the influence of sea coal on the solidification pattern and on the temperature distribution in the mold. They surmise, probably correctly, that the difference of thermal behavior between sea coal sand and synthetic sand may be ascribed to chemical reactions in the sea coal sand. It should be noted that the heat of reaction of the sand can be found by means of analog computer experiments in a similar way as is done in the paper "Heat Flow in Moist Sand" published in this volume. In order to determine the heat source by means of the analog computer, it would be necessary to have a set of temperature observations in the sea coal sand under specified conditions.

For the sake of accuracy in the experimental work as well as of simplicity in computation, it would be desirable that the "specified conditions" would be those of application of a constant temperature to the hot surface of a large slab of sand. The outer surface of the sand could be held by a steel plate representing the "flask." Inasmuch as a constant temperature at the hot surface may not be too easy to obtain experimentally, it is conceivable to change the condition to those of placing a heavy hot steel block against the sand slab.

Whatever the conditions are, it is important that both the simplicity of calculation and the accuracy of the original experi-

ment be taken into consideration when planning.

MR. DUNPHY: The term "thermal block" has been used only to indicate a heat source, i.e., heat of fusion, and not a high thermal resistance. The release of fusion heat at the liquidus temperature and/or the eutectic temperature delays the cooling

of the central portions of the casting resulting in the isothermal "holds."

The question of interface temperature accuracy has been realized as a serious experimental difficulty inasmuch as it represents a measurement of temperatures in a plane of "no width" with detecting devices of finite size. The problem of positioning the thermocouple bead in regions of sharp temperature gradient was recognized also and care was accordingly exercised to ensure accuracy.

Dr. Paschkis' reference to heat sources active in the central regions of the casting as being responsible for the eutectic "hold" is not in keeping with our views that the "hold" is a result of the movement of solidification waves through the casting which produces a thermal block. The heat source is, therefore, heat of fusion arising from eutectic solidification in the band of active solidification, not throughout the casting.

C. O. Burgess: Work of this kind can be of great practical importance. For example, the spread between the end of dendritic formation and the beginning of freezing in a given iron and increase in this spread with carbon equivalent or perhaps more properly with superheat, may explain difficulties in getting sound castings in heavy sections using a very high carbon

equivalent gray iron.

The work with seacoal indicates how such a condition might be cured simply by changing sand mixtures. It should be stressed that certain of the work that Pellini and Dunphy have done gives us a new insight and some practical answers to the

problem of obtaining sound castings.

MR. DUNPHY: With regard to Mr. Burgess' comments we certainly hope that these solidification studies will help solve some of the many anomalies in gray iron shrinkage characteristics. The particular point raised by Mr. Burgess concerning the effect of superheat in aggravating shrinkage characteristics of high carbon equivalent irons cannot be answered categorically by simple examination of the timing of the various solidification waves as affected by superheat. The fact that higher superheat results in greater liquid shrinkage is likely the significant feature involved in the occurrence of such porosity. The interesting effect of seacoal in erasing to some extent the effect of superheat and speeding up the solidification process in general certainly has practical application, for example, in establishing directional solidification in castings of intricate design.

³ Technical Director, Gray Iron Founders' Society, Inc., Cleveland.

SOLIDIFICATION OF STEEL AGAINST SAND AND CHILL WALLS

By

H. F. Bishop,* F. A. Brandt* and W. S. Pellini**

ABSTRACT

Thermal studies of the solidification of steel from chill and sand walls have shown that solidification proceeds in wave-like fashion by the travel of "start" and "end of freeze" waves. The thermal characteristics of the mold walls determine the rate and mode of travel of these waves. In the case of the chill walls which remove heat at a high rate the two waves travel in conjunction thus creating a band of freezing which proceeds to completion through isothermal liquid. In the case of sand walls which heat at relatively low rates the process is distinctly different in that the "start of freeze wave" completes its travel through isothermal liquid separately creating a semi-solid condition which extends through the entire casting. The effects of wall thickness and superheat of the metal have been shown to modify the timing and rate of progression of the freezing waves.

Introduction

THE PROBLEM OF CONTROLLING the solidification of castings in order to produce soundness together with high yield is of primary importance to the foundry field. Accordingly much of the research being conducted by foundry groups has been aimed at this problem and important advances have been made. Continued advance in this field is, however, hindered by a lack of basic information concerning the true nature of the solidification of metals. In spite of the considerable amount of research performed we still do not know, except in a qualitative sense, the contribution to solidification of such factors as mold materials, metal characteristics, superheat, mold geometry, etc.

Any solution, even partial, to such a broad and complex problem requires an extensive program of research aimed at isolating in turn each of the several variables for exacting study. The present report presents the results of the first of a series of investigations at NRL aimed at providing basic information prerequisite to the solution, of solidification problems.

Inasmuch as the solidification of castings proceeds from mold walls by a process of heat extraction it is essential that the basic features of wall solidification as a function of the thermal properties of the mold

wall be understood. Thus, the first and most elemental question to be resolved concerns the nature and course of solidification from mold walls of widely different thermal-properties and heat capacity. As a first approach in resolving these questions, studies were made of the solidification of steel from the simple plane walls of four-sided molds. The mold walls investigated included the two extremes utilized in practical foundry operation—chill (iron) and sand. These were varied over a range of thickness; the sand in addition was varied from green to dry. The contribution of superheat was also investigated.

The exact nature of wall solidification may be expected to vary in degree depending on the geometry of the mold. This factor must be considered as an associated variable serving to modify but not change in kind the basic characteristics of wall solidification for any given metal. Studies of the modifications of wall solidification associated with changes in mold geometry must necessarily follow as a second approach to the overall solidification problem. Such studies are underway.

The relative lack of quantitative experimental data on solidification may be ascibed in part to the choice of methods used in past investigations. Four methods have been used; these are.

Indirect Methods (1) Mathematical computation.

(2) Electrical Analogue.

Direct Methods (3) "Pour-Out" or "bleeding."

(4) Temperature measurement.

The indirect methods rely on data of the thermal properties of metal and mold which in many cases are either not available or unreliable. The mathematical method is exceedingly cumbersome and its results are generally suspect because of the simplifying assumptions required. The electrical analogue method has a great advantage of flexibility and its potentialities have been clearly indicated in a series of investigations by Dr. Paschkis under sponsorship of the American Foundrymen's Society.

The "pour-out" method has been the most widely used. It is, however, a tool which gives only approximate information. In the case of many alloys with wide solidification ranges such as irons the method

OLDS

realsents with g the dient nsure ntral told" lt of

heat d of tical den-

iron haps tting

rbon

ight

be

nave

the

we

olve

eris-

the

of

ally

tion

eat

ure

ing

neat

inly

nal

^{*} Metallurgist and ** Head, Metal Processing Branch, Metallurgy Division, Naval Research Laboratory, Washington, D. C.

will not work at all; in cases of alloys with intermediate solidification ranges, the information may be misleading.

Temperature measurements have not been used to any appreciable extent except to record final solidification at the center of castings. This method provides exacting information for all types of metals, with wide or narrow solidification ranges and was adopted in the present program of investigation to follow the entire course of solidification from surface inward.

Experimental Procedure

The test castings used for both the chill and sand wall studies were of square cross-section, 20 in. high, tapering from 71/4 in. square at the top to 63/4 in. square at the bottom. These castings were poured with induction furnace steel having a nominal analysis of 0.60 per cent carbon, 0.50 per cent manganese, and 0.30 per cent silicon with 0.10 per cent aluminum added to the ladle. The chill mold castings were surrounded on the four vertical sides by gray iron walls of uniform thickness. Feeding was provided by risers 7 in. in diameter and 12 in. high. The sand castings were surrounded by sand walls of uniform thickness and molded in special flasks made of 1/4-in. thick steel plate. Feeding was provided by risers 9 in. in diameter and 12 in. high. All castings were bottom-gated through a sprue molded in a separate flask. Figure 1 shows details of a sand mold, thermocouples, and recorders ready for a test. The same general arrangement was used for the chill mold studies.



Fig. 1-Photo of test assembly.

The chills used were gray iron while the sand molds were made of an A.F.S. No. 80 silica sand containing 3 per cent water, 3 per cent bentonite, 0.5 per cent cereal and 1.5 per cent dextrine.

Temperatures were measured across both the casting and the mold in a plane at the midheight of the casting. At this position bottom and riser gradients were essentially eliminated and solidification could be considered undirectional; i.e., as growth of a solid

skin in a direction perpendicular to the vertical mold surface. Steel temperatures were measured by means of Pt-Pt Rh (13%) thermocouples enclosed in 1/8-in. OD, 1/16-in. ID quartz tubes which extended completely through the casting and mold (Fig. 2) parallel to and at varying distances from a reference mold wall. All hot junctions were immersed equally in 31/2 in. of steel and thus temperature inaccuracies due to conductivity losses along the thermocouple wires were minimized. The 1/8-in. on quartz tubes were the smallest which could protect the thermocouples from the erosive action of the metal and even these softened to some extent and were bowed upward by the bouyant force of the liquid steel. X-ray examination of sections containing the tubes showed however that the deformation was never more than an inch and was entirely in the vertical plane so that the beads maintained the correct alignment with respect to the inter-

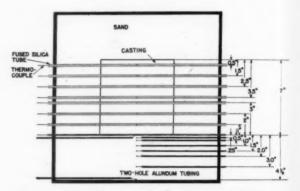


Fig. 2-Sketch showing thermocouple locations.

ti

Chill mold temperatures were measured with chromel-alumel couples flash welded to the bottoms of 3/32-in. diameter holes which were drilled perpendicular to the chill wall to the desired distance from the interface. These were grouped about the plane of the casting couples. Sand mold temperatures, except at the interface, were also measured with chromel-alumel couples. These thermocouples were unprotected and molded horizontally in the sand parallel to the casting surface. Interface temperatures were measured on both the casting side and the mold side of the interface.

The temperature on the casting side of the interface was obtained in the same manner as the other metal temperatures with the quartz tube containing the couple placed adjacent to and extending across the mold surface. In this case a protection tube of only ½16-in. on afforded adequate protection for the couple because of the faster freezing of the metal at the interface. The interface temperature on the mold side in the chill tests was obtained by first drilling a hole through the chill until the point of the drill pierced the inside surface. The bead of the chromelalumel couple was flash welded into this opening, effectively plugging the hole, and then covered with a thin silica wash. No wash was used on the chills ex-

ALLS

nold eans g-in. tely and All . of convere nallthe d to ant secthe was ainter-

ro-

ar

in-

he at nel nd ston

erer ng ss of ne at

11

5,

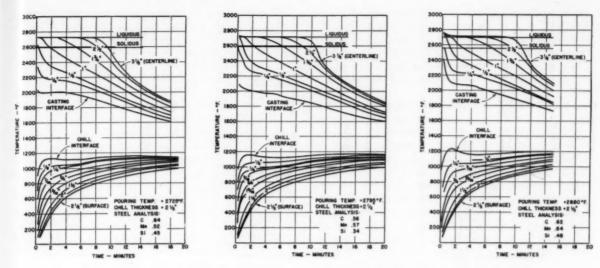


Fig. 3-Superheat series. Temperature cycles at various locations in casting and chill.

cept at this point. The natural oxide coating of the "as-cast" surface of the chill walls served to prevent the casting from fusing to the chill.

The interface temperature of the sand mold was obtained by locating a Pt-Pt Rh (13%) thermocouple junction, again protected only by a silica wash, at the sand interface. For a distance of 3 in. from the junction, this thermocouple was molded parallel to the casting interface so that it would become heated by the surrounding sand and thus minimize conductivity losses along the wires.

The platinum and platinum-rhodium wires were 26 gage—the chromel and alumel wires 28 gage. The casting temperatures were recorded on a 16 point automatic potentiometer recorded 1500 to 3000 F scale with an accuracy of 0.25 per cent of full scale and a sensitivity of 0.1 per cent. The sand temperatures were recorded on a similar instrument 0 to 2500

F scale. These instruments permitted sequential temperature readings from 16 thermocouple stations within 30 sec. Since in the majority of tests there were not over eight couples attached to each instrument, readings were obtained from each thermocouple on 15-second cycles.

Characteristics of Solidification from Chill Walls

Temperature Distribution in Casting and Mold—Fig. 3 and 4 show the temperature cycles at various locations in the casting and chill walls during the solidification process. These data are replotted in Fig. 5 and 6 to show the thermal gradients from the casting centerline to the outer edge of the chill wall at various times after pouring. Figures 3 and 5 illustrate the thermal effects resulting from increasing superheat and Fig. 4 and 6 of increasing mold wall thickness.

The "liquidus" temperatures indicated in Fig. 3 to

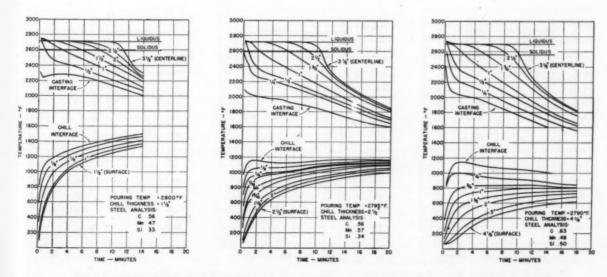


Fig. 4-Chill thickness series. Temperature cycles at various locations in casting and chill.

th

ai tu in

ch

th

of th ga

in

co

in

air

th

ca

ch

an

for

to

ple

co

int

pe ne ver wa caj fac tha

thi

fica the mo

rec

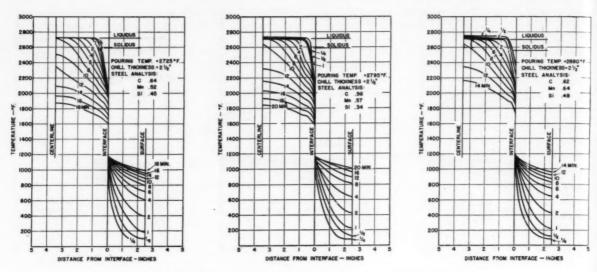


Fig. 5-Superheat series. Thermal gradients in casting and chill.

6 represent the temperatures at which the thermal arrest occurs upon cooling. The "solidus" temperatures represent a "technical solidus" which is the temperature, indicated by inverse-rate cooling curves, where delayed cooling resulting from the liberation of heat of fusion is no longer observed. The "technical solidus" indicates essential solidity, except for possible thin films of interdendritic liquid metal the solidification of which cannot be detected thermally.

Thermal Course of Chill Walls—Two variables were considered in investigating the thermal course of chill walls (1) the relative thermal capacity of the chills resulting from their different thicknesses and (2) variations of the relative heat contents of the metal resulting from superheat. It should be noted that other possible variables were held constant: viz, the composition of the metal undergoing solidification,

and methods of chill application.

It is apparent from Fig. 3 and 5 that the superheat condition has relatively little effect on the thermal course of the chill walls. Marked temperature differences can be observed only at the chill interfaces for the first 2 min after pouring. This is due to the initial temperature difference in the metal in contact with the chills. The higher pouring temperature is reflected in a higher rate of heating and a higher maximum temperature at the interface. The differences are greatly reduced in the later stages of the solidification process.

The thermal capacities of the chill walls, however, show pronounced effects on the entire thermal course of the walls, Fig. 4 and 6. The effects are not marked during the period of initial surface heating of the walls for during this period the walls are heated ap-

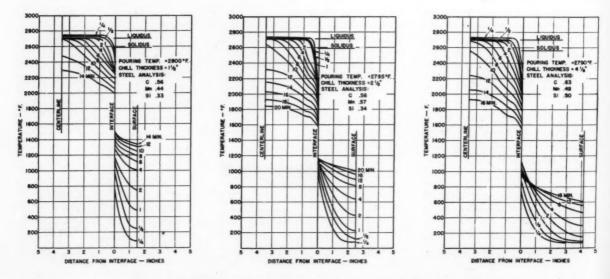


Fig. 6-Chill thickness series. Thermal gradients in casting and chill.

preciably only near the interface. At this stage all of the walls, $1\frac{1}{2}$ to $4\frac{1}{8}$ in. thick, behave essentially as if of the same thickness. However, as solidification proceeds and general heating occurs throughout the walls, the various heat capacities of the mold walls begin to be reflected markedly in the thermal course of the mold walls.

The first indication of the heat capacity effect is shown by the timing and nature of a thermal disturbance at the interface due to the formation of an air gap. In the system with the thinnest wall this disturbance is developed suddenly at 1/2 min after pouring as indicated by a general increase of temperature at the casting side of the interface. The chill side shows a mild inflection in the rate of heating at this time but otherwise continues to rise in temperature. The systems with the intermediate and large size chills undergo a gradual temperature reversal in the chill side at about 11/2 and 2 min respectively. When the air gap forms in these systems, the cooling rates of the casting side of the interfaces are reduced. That these temperature disturbances are the result of air gap formation was verified experimentally by pouring a similar casting against 21/2-in. thick chills with a 45-deg tilt such that the bottom face remained in continuous contact with the chill while the opposite side could shrink away to form an air gap. The chillinterface temperature on the top side fell when the air gap formed, but the chill-interface temperature on the bottom side climbed steadily throughout solidification.

al

er-

or

al

th

6.

ci-

es

fi-

r,

se

d

ne

The fact that an air gap forms earlier with thin chills than with thick chills is ascribed to the more rapid heating of the thin chills which then expand and pull away from the casting. The times of air-gap formation as related to chill thickness are inherent to the manner in which the chills in these tests completely surround the casting.

Following the stage of air gap formation, with its consequent drop in rate of heat transfer across the interface, further indications of the relative heat capacities of the mold walls are provided by the temperature course of the various mold walls. The thinnest chill wall shows insufficient heat capacity to prevent continued rapid heating throughout the entire wall. The intermediate chill wall has sufficient heat capacity to maintain an approximately constant interface temperature while the heaviest is oversufficient in that a falling temperature occurs.

An indication of the relative heat capacities of the three chill wall sizes can be made by comparing outer surface temperatures at a given time. For example 2 min after pouring the surface temperature of the $1\frac{1}{2}$, $2\frac{1}{2}$ and $4\frac{1}{8}$ -in. chills, Fig. 6, are 770 F, 425 F, and 140 F respectively.

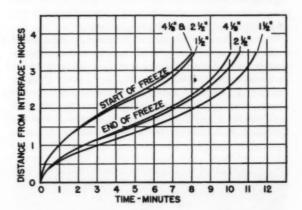
Solidification Course from Chill Walls

For a basic understanding of the process of solidification from mold walls, it is essential to consider the nature of heat transfer from the casting to the mold. This may be deduced from the shape of the thermal gradient curves shown in Fig. 5 and 6. It is recognized that heat from three sources is metered

from the casting to the mold during solidification: (1) specific heat of liquid metal, (2) heat of solidification and (3) specific heat of the solid metal which has already formed. The progression of solidification at any given time is governed by the rates at which these various heat components are metered to the mold.

It will be noted from Fig. 5 and 6 that the usual continuous thermal gradients generally associated with heat flow are disturbed by the formation of a "knee" in the temperature range of transformation from liquid to solid. The abrupt change in gradients at the knee is due to the necessity of removing heat of solidification from this zone before the thermal gradients can move inward. The "knee" zone, thus represents a band of active solidification which forms a thermal block to the central liquid portion of the casting. The initial development of this thermal block occurs early in solidification during the formation of the initial skin on the mold wall.

Since the solidification "knee" zone becomes established at and below the liquidus temperature, it will not act as a thermal block to the specific heat of the liquid metal. Thus, during the formation of the initial skin at the wall surface the superheat of the re-



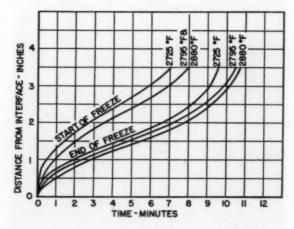


Fig. 7—Progression of solidification. (Top) Chill thickness series; P. T. (Pouring Temp.) 2800 F. (Bottom)
Superheat series; chill thickness 2½ in.

mainder of the liquid is simultaneously metered out at a rapid rate as indicated by the short time required for the gradient curves in the central portion of the casting to reach the liquidus temperature. It will be noted for example that the 180 F superheat in the casting of Fig. 5 poured at 2880 F is lost in little over a minute, and in this time a completely solid skin ½ in. in thickness has also been formed. Since significant cooling of the central portion of the casting below the liquidus temperature cannot occur because of the thermal block, the central portion of the casting gravitates to an isothermal condition at essentially the liquidus temperature and remains as such until reached by the solidification knee.

The passage of the solidification knee is denoted by a temperature drop which is consequent to the removal of heat of solidification. For example, the center of the casting poured in the small size mold, Fig. 6, remains near the liquidus temperature until 10 min after pouring. However, in the interval between 10 and 11 min (interpolated) during which time its solidification is completed, a drop of approximately 125 deg is noted. With the passing of the transformation knee the remaining, now solid, metal on the mold wall side of the "knee" cools further thus contributing specific heat of solid to the total flow of heat into the mold. These basic features of solidification are observed to hold for all the test castings.

The specific contributions of superheat and mold wall thickness to solidification may be observed more readily by replotting the thermal data in terms of the progression of solidification waves or "fronts" through the casting. Figure 7 shows the course of solidification of the castings thus summarized from the thermal data of Fig. 5 and 6. The rates of progression of the forefronts and ends of the "knee" zones are shown as

"start of freeze" and "end of freeze" curves respectively.

The effects of superheat and mold thickness on the course of solidification are shown in both the "start" and "end" curves; the specific effects being similar for both curves but somewhat more pronounced for the "end" curves. Thus, in order to simplify discussion, and because of the greater practical significance of the "end of freeze" curves, further discussion will be restricted to the wave of complete solidification.

It was observed earlier from the gradient curves that superheat is lost rapidly during the stage of initial skin formation. Since the mold has a limited capacity for accepting heat, the presence of superheat necessitates a reduction in the flow of solidification heat. Figure 7 shows that this is reflected in decreased initial rates of solidification. Following the complete release of superheat, the rates of solidification of all castings become equal regardless of initial superheat; i.e. the solidification curves are thereafter displaced by a time interval which is proportional to the amount of initial superheat. This time displacement is retained throughout the remainder of solidification. An increase in superheat from 25 F to 180 F increases solidification time approximately 13 per cent (9.5 to 10.7 min).

The relative heat capacities of the chill walls also affect solidification as shown in Fig. 7. During the first $\frac{1}{2}$ min the solidification rates of the three castings are identical. This should be expected inasmuch as during this period only the mold wall surfaces are heated. After the first $\frac{1}{2}$ min the solidification of the casting with the lowest heat capacity chill wall ($\frac{1}{2}$ in.) becomes slower. The solidification rates of the castings with the two larger chill walls ($\frac{21}{2}$ to $\frac{41}{8}$ in) are identical for the first 4 min after which the

In

m

in

ma for tin

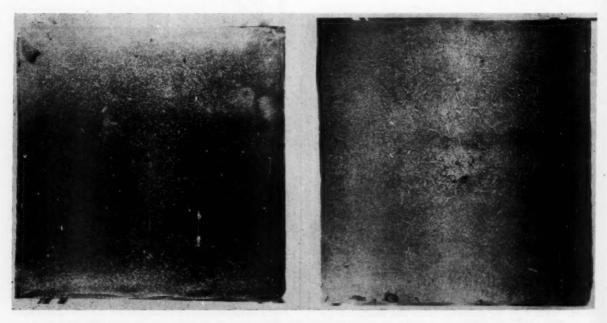


Fig. 8-Representative macrostructure of chill cast- ings. (Left) P. T. 2725 F; (Right) P. T. 2795 to 2800 F.

Photo reduced about 22% in reproduction.

LLS

Dec-

the rt" for the on. of be ves of ted eat on sed ete all at; ed

he

on.

ses

to

lso

he

st-

ch

re

of

all

of

1/8

he

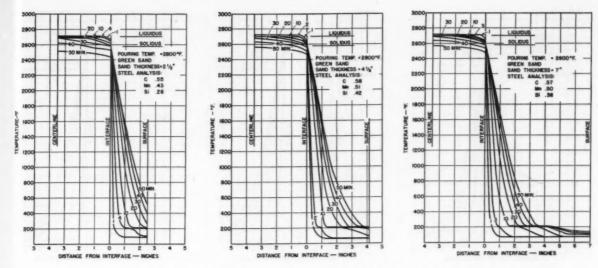


Fig. 9-Green sand mold thickness series. Thermal gradients in casting and mold.

rate for the intermediate 21/2-in. wall becomes somewhat slower.

It is noted, that the time of total solidification is decreased approximately 13 per cent (11.5 to 10 min) by increasing the chill thickness from $1\frac{1}{2}$ to $4\frac{1}{8}$ in. Increasing the chill wall thickness from $1\frac{1}{2}$ to $2\frac{1}{2}$ in. decreased the final solidification time by approximately 9 per cent (11.5 to 10.5 min) while an increase from $2\frac{1}{2}$ in. to $4\frac{1}{8}$ in. further decreased this time only 5 per cent ($10\frac{1}{2}$ to 10 min), thus indicating a point of diminishing returns past approximately $2\frac{1}{2}$ in. thickness. From practical considerations the $2\frac{1}{2}$ in, thick chill may be deemed of sufficient thickness.

While variations of chill thickness or of superheat may have approximately the same effect on the time for final solidification, the mechanisms by which this time difference is developed are completely different. The retarding effect of low heat capacity chills does not begin until solidification has started and then becomes increasingly greater as solidification progresses. The superheat effect retards the solidification process only during the initial state of solidification.

Variations in thickness of chill wall have no significant effect on the macrostructural characteristics of the casting. Increasing superheat, however, produces a marked coarsening effect throughout the casting. Figure 8 shows typical macrostructures resulting from variations in superheat. It is hypothesized that the fine structures present when superheat is low result from the formation of small crystallites during pouring which are distributed throughout the liquid and serve as nuclei or centers of growth as the start of freeze wave moves through the casting. With high superheat such centers are not developed hence the start of freeze wave moves through essentially nuclei-free liquid.

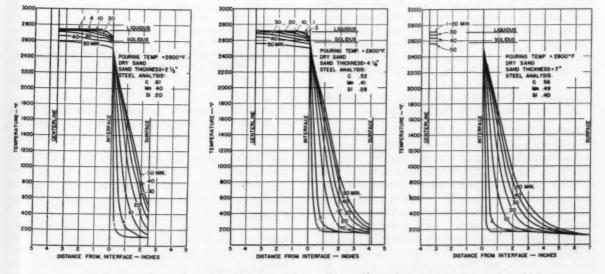


Fig. 10-Dry sand mold thickness series. Thermal gradients in casting and mold.

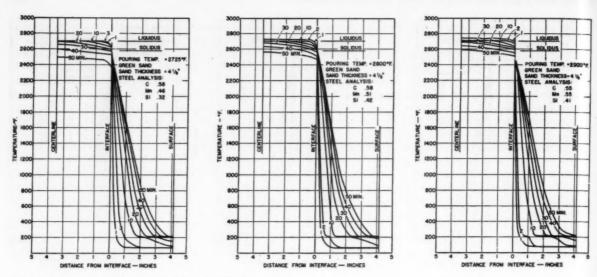


Fig. 11- Superheat series. Thermal gradients in casting and green sand mold.

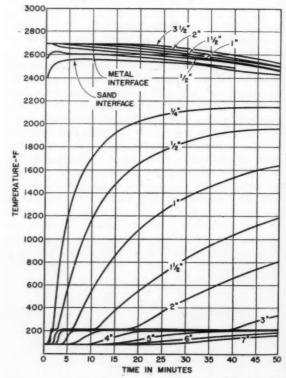
Characteristics of Solidification from Sand Walls

Temperature Distribution in Casting and Mold—In considering transfer of heat from the casting to a sand wall it is necessary to recognize that in practice, sands, unlike chills cover a wide range of materials which may differ considerably in thermal properties. Thus, in addition to thickness or thermal capacity effects, as discussed in the previous section on chill molds, it is necessary to consider the contribution of such factors as grain size, density, water content, binders, etc. For the sake of limiting the scope of the immediate investigation, the studies were confined to the role of wall thickness and water content of a single practical foundry sand.

Figure 9 shows metal and sand temperature gradients for 7 x 7-in. bar castings poured at 2800 F into green sand molds (3 per cent water) having wall thicknesses of 2½, 4½ and 7 in. Figure 10 shows gradients for a similar series with sand molds which had been baked at 450 F for 16 hr to remove water. The effects of superheat are shown by the series presented in Fig. 11. In this series the metal was cast into green sand molds having wall thicknesses of 4½ in

Liquidus and solidus temperatures were obtained in the same manner as for the chill castings.

Thermal Course of Sand Walls – Poor heat transfer properties of sand are reflected in the interface temperatures and in the thermal gradients in the sand walls, Fig. 9, 10 and 11. In contrast to the chill walls the thermal gradients are steep and the interface temperature of the sand approaches that of the solidifying metal. The interface conditions of the metal and mold are more clearly shown in Fig. 12 which is a plot of the temperatures recorded from the individual couples in a typical test. It can be seen that the mold interface temperature quickly approaches the metal interface temperature. Five min-



th

bl

th

at

su

ne

ap

ra

of

rec

lev

Wa

ou

at

the

mi

the

Fig. 12-Temperature cycles at various locations in casting and 7 in. green sand mold. P. T. 2800 F.

utes after pouring the temperature difference is approximately 50 F, and after 30 min the difference has become negligible. No irregularities were noted in interface temperatures which would signify the formation of an air gap.

It is of interest to note, both from the gradient curves of Fig. 9 and the time-temperature curves of Fig. 12, that the presence of water results in a transformation "knee" or thermal block similar to the liquidus to solidus transformation "knee" noted in

277

p-

125

in

r-

nt

of

10

in

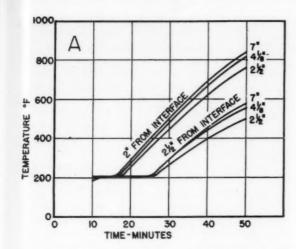


Fig. 13-Effects of flask on sand temperature. (Left)
Green sand. (Right) Dry sand.

the solidification of the steel against chill walls. This thermal block developed by the water to steam transformation at 212 F.

The ultimate depth to which the 212 F thermal block extends into the sand at the time of complete solidification provides a practical means of denoting the thickness of sand wall which contributes to heat absorption during solidification. On this basis, mold walls which exceed this thickness may be considered to have oversufficient heat capacities. The curves for the green sand castings Fig. 9 show that at the time for complete solidification the 212 F "hold" extends approximately 41/2 in. into the 7-in. mold wall. On this basis a 41/2 in. wall thickness would be deemed of sufficient heat capacity for 7 x 7-in. sections. If 212 F is taken arbitrarily as an indication of sufficient thickness for dry sand molds it is noted that a mold wall approximately 5 in. thick (Fig. 10) is required. Figure 9 shows that when green sand walls are less than this sufficient size the 212 F hold reaches the flask prior to complete solidification and the mold flask then enters into the heat absorption cycle of the mold wall.

The flask, being steel, extracts heat at a more rapid rate than the sand and may be considered as a chill with its effectiveness reduced by the sand insulator between it and the casting. The increased amount of heat removed by the flask in the thin-walled mold reduces the sand temperature near the flask to a lower level than it is at comparable locations in the thickwalled molds. This can be seen in Fig. 13 which shows the temperatures of the sand at 2 and 2½ in. from the interface in $2\frac{1}{2}$, $4\frac{1}{8}$ and 7-in. walls at various times.

The metal interface temperatures in these castings at comparable times differ by only a few degrees while the sand temperatures 2 in. from the interface are as much as 50 F lower in the thin-walled mold than in the two thicker molds. Thus when the 21/2-in. mold becomes heated throughout, its gradient through the first 2 in. of sand is greater, and since the rate of heat transfer is proportional to the gradient through the

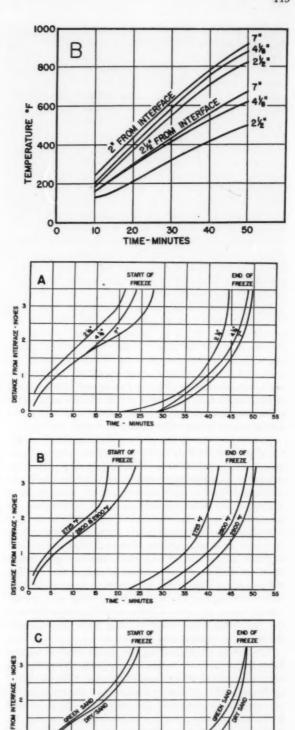


Fig. 14—Progression of solidification. (A) Green sand mold thickness series; P. T. 2800 F. (B) Green sand superheat series. Mold thickness 41/8 in. (C) Comparison green vs dry sand castings. Mold thickness 41/8 in. P. T. 2800 F.

sand, heat is removed from the casting in the smallest mold at a faster rate than it is from the larger molds.

There are no significant differences in the sand gradients in the pouring temperature series.

Course of Solidification from Sand Walls-Gradient curves of Fig. 9 to 11 clearly indicate that the poor heat transfer properties of sand result in a marked modification of the gradients in the solidifying steel as compared to those of chill castings. The gradients in the steel are comparatively mild as dictated by the high temperature condition at the interface, and the transformation "knee" is accordingly broadened so that it encompasses almost the entire cross-section of the casting. As in the chill castings, superheat is lost early in the solidification process and almost the entire cross-section of the casting becomes isothermal at the Injuidus temperature. Only a small amount of solid is formed at the surface during this interval. Superheat is metered to the mold almost to the exclusion of other forms of heat.

The progression of the "start of freeze" and "end of freeze" waves across the 7 x 7-in. sections are shown in Fig. 14. The "start of freeze" curves are similar in shape to those of the chill castings, showing a rapid initial solidification which gradually retards and then accelerates near the center of the casting. As indicated by the "end of freeze" curves, final solidification does not begin at the surface until after solidification has started at the center. At this point a mushy condition exists over the entire casting.

At any location in the casting there is a time interval of at least 20 min between the time when solidification starts and when it ends, and during this interval metal is solidifying upon the dendrite skeletons originated by the "start of free" wave. Solidification is progressive from the surface to the center in the sense that the nearer the metal is to the casting surface the more advanced is its state of solidification. However, it is not until more than 30 min after pouring that any of the castings have become completely solid for a distance of even 1/2 in. from the surface. When such a completely solid skin is formed solidification throughout the rest of the casting is in an advanced stage; i.e., only a little liquid remains between the dendrites. The final freezing wave then travels through the casting at a rapid rate.

The effect of increasing the mold wall thickness from $2\frac{1}{2}$ to 7 in. is shown in Fig. 14A. The effect is one of a general delay in solidification with increased thickness. This is exactly opposite to the effect of increasing the thickness of the chill wall which showed faster solidification with increased thickness of wall. This apparently anomalous result is ascribed to the contribution of the steel flask as a chill which was discussed in the previous section. In the case of the $2\frac{1}{2}$ and $4\frac{1}{8}$ in. green sand walls the water hold reached the flask in approximately 15 and 25 min respectively thus bringing the chill action of the flask into play at these times.

In the case of the 7 in. wall, the water "hold" only extended approximately 4½ in. into the wall at the time of complete solidification thus the flask did

not contribute significantly. Increasing the mold wall thickness from $2\frac{1}{2}$ in. to $4\frac{1}{8}$ in. shows an increase in solidification time of approximately 10 per cent (44 to $48\frac{1}{2}$ min) while a further increase from $4\frac{1}{8}$ to 7 in. shows a relatively small increase of approximately 2 per cent $(48\frac{1}{2}$ to $49\frac{1}{2}$ min). It may be concluded that a $4\frac{1}{8}$ -in. green sand wall is approximately of "sufficient" thickness for a 7-in. x 7-in. section above which the extra sand does not contribute significantly to solidification and below which chill effects of the flask are developed.

The specific effects of superheat are summarized in Fig. 14B for castings poured in 4½-in. molds. As in the case of the chill wall studies the two higher superheat curves fall closely together and are displaced from the lowest superheat curve toward later times. This displacement appears to originate from a delay in the "start of freeze" curve as a result of superheat liberation. This delay is reflected in the "end of freeze" curves. It is observed that after the delay times have elapsed, the rates of travel of the various "end of freeze" curves are approximately the same. Thus the effect of superheat in introducing an early delay in solidification and increasing the solidification interval is basically the same for both chill and sand castings.

The effect of 3 per cent moisture in 41/2-in. sand walls is shown in Fig. 14C. The presence of moisture expedites solidification during the initial stages of solidification and causes the development of the "end of freeze" wave to occur earlier. This effect is apparently lost during the later stages of solidification as indicated by the gradual merging of the green and dry sand "end of solidification" curves which become coincident at final solidification. This peculiar course of events is not completely understood. It is hypothesized that the somewhat more rapid heating noted for the dry sand wall results in bringing the flask into play earlier in the case of the dry sand thus expediting the last stages of solidification. If such is the case the coincidence of final solidification times is fortuitous and specific to the 41/8 in. sand wall, in that this thickness is less "sufficient" for the dry than for the green sand.

fer

ou

Te

the

tra

Fo

he

he

les

fica

wh

tra

the

mo

mo

ft/

gre

the

acc

sar

rat

wa

wa

As in the chill castings, only pouring temperature altered the macrostructure of sand castings. The trend toward coarser structures with increasing superheat can be noted from Fig. 15.

General Discussion

The foregoing thermal analyses have shown that solidification is basically a heat transfer problem. The exactness inherent to the thermal analysis method of investigating solidification together with its adaptability to the simultaneous study of metal and mold conditions provides a means of deducing the nature of the heat transfer processes which occur during solidification.

The total heat content of liquid metal, which was absorbed during the melting and superheating process, includes specific heat of solid, heat of fusion, and specific heat of liquid. Solidification proceeds by the transfer of these heats from the metal to the mold LLS

vall in (44 0 7 clv led of ove tly the in in erced les. lav eat of av ills ne. rly ca-

nd

nd

ol-

of

u-

nd

ne

se

16-

or

to

it-

se

it-

is

ne

re

ıd

at

at

16

of

it-

d

re

g

18

0

d

e

d

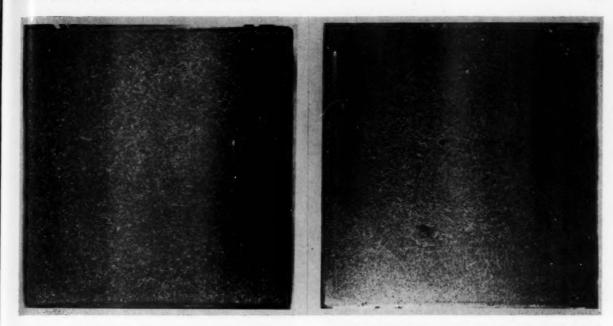


Fig. 15-Representative macrostructures of sand castings. (Left) P. T. 2725 F. (Right) P. T. 2800 to 2900 F. Photo reduced about 22% in reproduction.

walls. The thermal characteristics and heat capacities of the mold walls determine whether all of the transferred heat is absorbed by the mold or if part of it is passed on to the atmosphere by radiation from the outer mold surface. The important thermal characteristics to be considered include the volumetric specific heat and conductivity of the mold material. Total heat capacity, of course, is a direct function of the wall thickness.

For practical ranges of wall thickness, chill walls transfer an appreciable fraction of the heat liberated during solidification to the atmosphere by radiation. Sand walls however absorb essentially all of this heat. For dry sand, the heat is entirely absorbed as specific heat of the sand while green sands remit part of the heat to the vaporization of water. For this reason, less mold thickness becomes heated during the solidification process when the sand is green rather than when dry.

The rate at which heat is removed from the casting is basically determined by the rate at which heat is transferred from the mold interface by diffusion throughout the mold. High conductivity, high volumetric specific heat, and low level of temperature of the mold provide conditions of high rate of heat removal. Gray iron has a volumetric specific heat of more than three times that of sand (90 vs 26 Btu/cu ft/°F) and a conductivity approximately 20 times greater (18.4 vs 0.9 Btu/ft,hr°F) than sand. Thus, the rate at which heat is removed by the iron chills is accordingly much higher than that of sand. In the same sense heated mold walls remove heat at a lower rate than cold walls since the heat capacity of the walls has been partly satisfied.

It was observed that solidification of steel from chill walls, which remove heat at high rates proceeds as a

band of solidification by the simultaneous travel of "start" and "end" of freeze waves. The space separation of these two waves is relatively narrow, hence the solidification process consists basically of the movement of the solidification band from the mold wall into isothermal liquid which is retained essentially at the liquidus temperature. Solidification of the same steel from sand walls, which remove heat at comparatively low rates proceeds in a distinctly different fashion. Solidification is still progressive in the sense that the "start" and "end" of freeze waves travel from the surface to the center of the casting, but a solidification band as such is not developed. The "start of freeze" wave completes its cycle before the "end of freeze" wave starts to follow it. In contrast, solidification in sand castings is therefore more general throughout the casting.

It is important to note that modifications in the thermal characteristics of the mold wall result not only in changes of solidification rates but also in marked modification of the solidification process itself. The full significance of these findings cannot be predicted at this time.

Variations in the range of practical thicknesses of chill and sand walls have been shown to affect solidification in correspondence with the extent to which the rate of heat removal is affected. Increasing the thicknesses of chill walls decreases the general temperature level of the walls, hence increases heat removal rates resulting thereby in increased solidification rates. Increasing the chill thickness above the point required to maintain approximately constant interface temperatures following air gap formation increases solidification rates relatively little compared to increasing the thickness up to this point. Accordingly, this point has been denoted as the point of

"sufficient" chill wall thickness. In the case of sand walls heat removal rates are increased markedly by decrease in sand thickness below the thickness required to prevent chill effects due to the flask. Solidification rates are accordingly increased by decreases in sand thickness below this point of "sufficient" sand thickness. Increasing sand thickness above this point does not modify heat removal rates since the extra sand thickness does not enter significantly into the heat transfer process. Solidification rates are accordingly not affected significantly.

Superheat has been shown to have relatively little effect on the overall thermal course of the mold walls inasmuch as its liberation occurs during the initial heating period. Comparison of the chill wall with the sand wall shows that the rate of superheat liberation is not sufficient to satisfy the heat demand of the chill wall. Accordingly appreciable solidification heat is also liberated at the chill wall during the period of superheat liberation. Sand, however, is very nearly satisfied by superheat alone and very little heat of solidification is liberated along with superheat. Basically, superheat represents excess heat in the system which must be eliminated prior to extensive solidification. Thus, the rate of formation of solid at the wall surface is decreased in the presence of high superheat.

The relative heat removal characteristics of chill and sand walls as indicated by comparison of solidification times and rates are of considerable practical interest. Such comparisons can be made from Fig. 16, which presents solidification curves for steel against "sufficient" chill and sand mold walls.

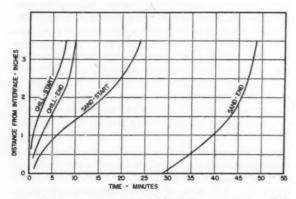


Fig. 16—Comparison of solidification curves for chill and green sand castings. Wall thickness 41/8 in.

P. T. 2800 F.

It is noted that the "start of freeze" solidification rate of steel against a chill surface is as a rough approximation 3½ times as great as it is against a sand surface. The "end of freeze" solidification rates are approximately the same for both sand and chill castings after the first inch of wall growth, but are displaced from each other by a time interval of approximately 40 min. The similarity in "end of freeze" growth rates of the castings solidifying against the two media has no apparent practical significance. The internal conditions within these two castings are en-

tirely different while the "end of freeze" waves travel through them. When the "end of freeze" waves have reached points 1½ in. from the centers of the castings, the chill casting has a completely liquid interior while the liquid in the central portion of the sand casting has been reduced to small interdendritic pools and films.

Considering only the times for complete solidification (10 min for the chill casting and 48 min for the sand casting), the relative chilling powers for the two mold media would be in the ratio of 5 to 1. On a heat balance basis this ratio would be even greater since, during the solidification period, the chill also removes considerable specific heat from the solid metal near the chill surface. This can be noted from Fig. 6 by the extent to which these locations have cooled below the solidus temperature.

The close agreement of these experimental findings with the predictions of Paschkis^{1,2,8} based on electrical analogue studies is particularly noteworthy. Paschkis treats the solidification problems strictly as one of heat transfer with electrical variables of capacitance, resistance, voltage and amperage replacing the casting variables of heat content, conductivity, temperature, etc. Thus the casting heat transfer cycle is synthesized on an electrical circuit. Typical electrical analogue curves obtained by Paschkis are shown in Fig. 17.

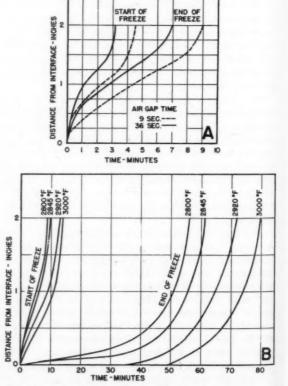


Fig. 17—Typical electrical analogue progression of solidification curves (Paschkis). A—Steel thickness 4 in.; chill thickness 3 in.; P. T. 2845 F; solidification range 2700 to 2600 F. B—Steel thickness 4 in.; sand thickness 10 in.; solidification range 2700 to 2600 F.

acientic

H

m

se

te

aı

fr

er

Q

th

fe

up

sh

Pla 7-i inc tin pla

the inte

2. ACT 3. ing (198 4. by 6 98.

ings
Co
V.
mate
catio
whice
In

meth mole this vant

1 T Unive LLS

el

ve

st-

or

nd

ols

ca-

he

wo

a

ter

lso

tal

. 6

ed

ngs

cal

kis

of

ce,

ist-

ra-

vn-

cal

in

В

5 4

ion and

) F.

The Analogue method correctly predicts that in solidification from chill walls the "start" and "end" of freeze waves proceed to completion in conjunction, producing a limited band of solidification which moves out into free liquid; also that solidification from sand walls is distinctly different in that the "start" and "end" curves proceed to completion in sequence thus creating a mushy condition which extends through the entire casting. The actual shapes and relative locations of the "start" and "end" of freeze curves and the delay effects resulting from superheat are likewise predicted correctly by Paschkis. Quantitative comparisons with the chill castings of this investigation are not possible because of the different test castings, chill practice, and air gap times upon which Paschkis' curves are based.

In the sand castings however, only differences in shapes are involved and comparison may be made according to Chvorinov's rule which states that solidification times for steel castings are proportional to the square of volume to surface area ratios. Comparison on this basis of the time of final solidification of the 7-in. x 7-in. bar casting poured at 2800 F in a sufficient sand mold is in agreement with the solidification time which Paschkis predicts for a 4-in. thick plate poured under like conditions. Scaling of the 7-in. x 7-in. bar to the same S/V ratio of the plate indicates a time of 621/2 min which compares with a time of 57 min reported by Paschkis for the 4-in. plate.

Acknowledgment

The authors wish to express their appreciation for the many helpful discussions with Dr. Marzke, Superintendent of the Division and the assistance extended to them by T. W. Snodgrass and E. T. Myskowski.

Bibliography

1. V. Paschkis, "Theoretical Thermal Studies of Steel Ingot Solidification" Trans., ASM vol. 38, pp. 117-147 (1947).

2. V. Paschkis, "Studies on Solidification of Castings," TRANS-

ACTIONS, A.F.A., vol. 53, pp. 90-101 (1945)

3. V. Paschkis, "Influence of Dry Sand Conductivity on Freezing of Steel Slabs" Transactions, A.F.S. vol. 58, pp. 147-152

4. N. Chvorinov, "Control of the Solidification of Castings by Calculation," Foundry Trade Journal, Aug. 10, 1939, pp. 95-

5. B. Matushka, Arch. Eisenhüttenwesen, vol. 2, p. 405 (1929). 6. B. Matushka, J.I.S.I., vol. 124, p. 361 (1931).

DISCUSSION

Chairman: H. A. SCHWARTZ, National Malleable & Steel Cast-

Co-Chairman: E. C. TROY, Foundry Engineer, Palmyra, N. J. V. PASCHKIS (Written Discussion): 1 The authors compiled

material of great significance to anyone concerned with solidification of metals. In studying the paper a few points occurred

which merit discussion.

University, New York.

In the "Introduction" the electric analogue is listed as one of the available indirect methods and it is stated that the indirect methods rely on data of the thermal properties of metal and mold. This is true; but as far as the analogy method is concerned this obvious disadvantage can be, to some extent, turned to advantage by studying the relative importance of different vari-

¹ Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia

ables, for example of the different properties of metal and mold. Thus it can be found which of them is important for the process in question and need be determined more accurately.

Regarding the selection of the dimensions for the test casting there may be some question if for a cross-section of 7 in. square a height of 20 in. is sufficient to warrant disregarding axial flow. Conditions are slightly improved by the fact that, thermally, the riser and the casting form a unit so that the actual effective height is more than 20 in. From this view point however, it might be preferable to place the thermocouple somewhat above mid-height of the casting.

The most critical items of the authors' investigation are the temperature measurements. Errors in temperature measurements are particularly severe near the interface. Whenever there are temperature gradients in the body to be measured the magnitude of possible error increases. It would be desirable if the authors could elaborate on the accuracy of their measurement. Although not directly applicable to the present tests certain information may be obtained in the paper by H. D. Baker and E. A. Ryder (See Ref. 1 in bibliography appended to this discussion).

When measuring sand temperatures, it would seem desirable to apply small discs to the bead of the couple, because, with fine wires, it is conceivable that the bead rests completely or almost completely in a space between sand grains which, of

course, would result in wrong measurements.

Measurement of the interface temperature in the chill raises the question, if the weldment extends into the chill. The temperature observed on the instrument is obviously that of the last point of contact between the two thermocouple wires.

In the discussion of the temperature-time history of the chill walls the term "thermal capacity of the chills" is used where "thickness" only should be mentioned. The variation was, as the writer understands, only of thickness but not of specific heat. The thermal capacity is the product of volumetric specific heat x thickness.

In the same section the authors comment on the "thermal disturbance at the interface due to an air gap." The observed increase in temperature in the metal is interesting. The writer, without having experimental data available, predicted such rise in a study of ingot solidification (Ref. 2 in bibliography).

In the same connection it is interesting that the authors find a time for formation of the air gap of 2 min for large size chills. In the first report of the Heat Transfer Committee (Ref. 3 in bibliography) the writer found that air gap times for 2-in. slabs were probably 9 sec and increased with the square of the thickness, provided that the chill was increasing proportionally to the thickness of the casting. Thus for a 7-in. slab (no corner effects) the air gap time should be 110 sec for 5.25-in. chill. At early times of solidification the thickness of the chill probably does not enter severely so that the 110 sec probably would hold for thinner chills and seem to check with the present authors' 2 min.

The authors use at two different places the word "sufficient chill" in different connotation. In the section on the "Thermal Course of Chill Walls," it is stated that "intermediate chill wall has sufficient heat capacity." However in the section on "Thermal Course of Sand Walls" the gage for "sufficient heat capacity" is taken differently. It appears to the writer, that the proper definition would be, that a "sufficient" chill or mold thickness is one, beyond which an increase in thickness does not result in a significant change of solidification time. In Fig. 4 however, the center of the casting with over sufficient chill freezes in 10 min against the 10.5 with the medium chill.

Towards the end of the section on the "Thermal Course of Chill Walls" a comparison is made of the surface temperatures of three chills, 2 min after pouring. The values are taken from Fig. 6.

A similar comparison could be made, by noting the time required to reach a given outside surface temperature, e. g. 600 F, with the different chills. The times are 1.4 min, 3.6 min and 15 min. In first approximation the times should increase proportionally to the square of the chill thickness. Thus for the 2.5-in. chill the time should be $(2.5/1.5)^2 \times 1.4 = 3.9$ min. The above value of 3.6 is a good check. For the thickest chill the check is not quite so good (10.5 min computed, as against 15 min observed). But this is proper, because the square law approximation loses its validity for longer times.

In Fig. 7 the curves for a pouring temperature of 2795 F and

2880 F coincide. It would appear from the authors' statements that with increasing pouring temperature the solidification time should increase. Do the authors have any explanation why these curves coincide?

The identical behavior of the chills of different thickness at early times is as thermally expected. At the instant of pouring

all of the chills act as "semi-infinite solid."

In the section on "Thermal Course of Sand Walls" the authors observe that in this case there is no indication of an air gap. The writer's experiments (Ref. 3 in appended bibliography) lead him to the conclusion that even if an air gap should exist when casting against sand, it would be of little significance because the thermal resistance of the air gap is small compared with that of the sand whereas it is large as compared with that of the chill.

In the section on "Course of Solidification From Sand Walls" the authors comment on the delay in solidification with increased chill thickness. The authors correctly relate this observation with the happenings on the cold surface of the "container"

(chill or mold).

Increase of "container" thickness increases the thermal resistance noticeably in case of the sand mold, but only to a negligible amount in case of the chill. On the other hand the heat storage capacity per unit volume is much larger for the chill (77.5 Btu/cu ft. F) than for the sand (28 Btu/cu ft., F). Thus in case of the chill, a large part of the heat extracted from the casting may be stored temporarily in the "container" (chill), whereas in case of sand only a small part can so be stored. An increase in the storage ability by increased thickness is more important in the case of the chill.

The observation on moist sand may eventually be clarified by work on the Analyzer particularly because the computations for moist sand are carried out without a flask and thus may

give an explanation for the observed difference.

In the general discussion a reference is made to a comparison of the authors' findings with those by the writer obtaining a good check of two observations made on different shapes and compared by the method of Chvorinov. (Ref. 4 in bibliography.) Chvorinov makes two statements:

1. That bodies having the same volume/area ratio freeze at

the same time, independent of their shape.

2. That two bodies having different volume/area ratios have a ratio of freezing times which is equal to the square of the ratio of the "shape factor" (the latter being the volume/area ratio).

of the "shape factor" (the latter being the volume/area ratio). The statements by Chvorinov have drawn much attention and therefore a thorough discussion may be in place. This discussion necessitates to analyze the "shape factor" which Chvorinov defines as the ratio volume/area. The shape factor, S, has of course the dimension of length. The shape factor, then, can always be related to a critical dimension, D, by means of a proportionality factor, p. S = D/p.

The critical dimension is defined as the thickness in case of large slabs with no end effects, or the diameter in case of cylinders or of spheres. In case of slabs where the end effects must

be considered, D stands for the smallest dimension.

Hereafter the subscript P will be used for large slabs without end effects, P denoting plate; the subscript S for spheres and C for cylinders of great length (no end effects). Thus $p_P=2$; $p_0=4$; $p_8=6$.

Work on the Analyzer proves that two bodies of the same shape (equal value of p being a necessary but not a sufficient condition) but different size (different value of S) require solidification times proportional to the square of the ratio of their D values. For example, if the D value is doubled then the solidification time is increased to four times its original value. This holds on the following three conditions.

- 1. That the mold size increases proportional to the size of the casting; that means that if D is doubled then the mold thickness is also doubled.
- 2. In case an air gap is formed, the time of formation of the air gap changes with the square of D.
- 3. That the heat flow from the outside of the mold does contribute to solidification; in other words that the mold is so thick that the temperature rise on the outside surface facing the flask is negligible.

If these three conditions are fulfilled this law of increase with the square of D value is rigid.

Chvorinov's statements would require that two bodies having

the same shape factor S but different p values, freeze in the same time. Experiments on the Analyzer show that this law is not rigidly correct, and, moreover, that the approximation to absolute validity depends on the amount of superheat. Comparing solidification times of large slabs with no end effects (D=4, p=2, S=2) with those of spheres (D=12, p=6, S=2) the results are tabulated below. It may be seen that the ratios of solidification times, slab/sphere are nearly one for high degrees of superheat but deviate considerably for low degrees of superheat.

COMPARISON OF SOLIDIFICATION TIMES OF A LARGE SLAB (NO EDGE EFFECTS) WITH A SPHERE, BOTH HAVING THE SAME VALUE

OF VOLUME/AREA			
Pouring Temp.	Time to I	Time to Reach Solidus	
°F	Slab (Sec)	Sphere (sec)	Slab/Sphere
3010	4600	4320	1.06
2900	3900	3920	0.99
2820	3280	3520	0.93
2800	3100	3400	0.91
2740	2375	3280	0.73

Checking all values of Chvorinov's Fig. 19 one may see that he covered *p* values from 2.11 to 6. In view of the findings of the above table there is serious doubt if his disregarding differences in pouring temperature might not obscure deviations from his stated law.

BIBLIOGRAPHY

1. "A Method of Measuring Local Internal Temperatures in Solids" by H. Dean Baker and E. A. Ryder, presented at the annual meeting of the ASME, December 1950.

 "Theoretical Thermal Studies of Steel Ingot Solidification" by Victor Paschkis, Transactions, ASM, vol. 38, pp. 117-147 (1047)

- 3. "Studies on Solidification of Castings" by Victor Paschkis, Transactions, A.F.S., vol. 53 (1945).
- 4. "Theory of the Solidification of Castings" by N. Chvorinov, *Die Giesseri*, British Iron and Steel Institute, Translation No. 117, vol. 27, pp. 117, 201, 222 (1940).

MR. PELLINI: We appreciate the time and effort which Dr. Paschkis obviously has expended in his review of the subject paper. His extensive experience in the field of heat transfer makes his review of great value to the authors.

We certainly agree with Dr. Paschkis concerning the flexibility of the Electrical Analogue. It should be recognized that this tool correctly predicted the features of solidification which have now been determined by direct experiment. In essence Dr. Paschkis has conducted foundry experiments using electrical circuits and has developed information which was later checked on the foundry floor. The potentialities of this approach should now be evident to foundrymen.

Dr. Paschkis is correct in his deduction that axial heat flow is not entirely eliminated with the casting design used. The casting in question represents a rather delicate balancing between two features-axial heat flow which is developed when the bar is not sufficiently long to be semi-infinite and shrinkage developed whenever a semi-infinite section is used. We know from our studies of the axial solidification of bars that semi-infinite conditions are developed at approximately 3T (21 in. for 7-in. bar.) from the casting end. It should be noted further that complete soundness in 7x7-in. bars is developed only to approximately 10 in. from the end. Full details may be found in a companion paper 1 [The Contribution of Riser and Casting End Effects to the Soundness of Cast Steel Bars" by Bishop, Myskowski and Pellini, pp. 171-180 in this volume]. Thus, 10 in. from the end is the closest practical approach to semi-infinite conditions for completely sound 7-in. x 7-in. bars. The couples, accordingly, were placed at 10 in. from the end of the casting.

The absolute accuracy of the thermocouples was approximately 4°F. The sensitivity 1.5°F. Precautions were taken to eliminate conduction errors and to ensure good thermal contact. Disawelded to the thermocouples heads were tried for sand temperature measurements and found not to change the reading of the couples. This may be due to the fact that the clay bond present in the sand provides thermal contact with the beads, or that the

the not their is o in I ing lish

H.

fine

wet

the

incr solid abore cent 21/2 volu reco in. It the

shor

protence ture pera Trance tion are o prov turb

cong pain It se satisf ticul resist

ance in the In apper follow passe crease ance also

sort

grou

conta

is less not a tion fluctu

media might chill"

In

³ Rass.

LLLS

same

bso-

ring

= 4.

= 2)

atios

de-

s of

EDGE

here

that

rom

s in

the

ion"

-147

kis.

nov.

No.

Dr.

ject

ility

tool

now

hkis

and

the

now

w is

ast-

een

bar

rom

nite

-in.

om-

oxi

End

ow-

rom

ndi-

les.

ing.

tely

ate

iscs

the

fines in the sand perform the same function. Ceramic sleeves were used to electrically and thermally insulate the wires from the sides of the drilled hole in the case of the couples which were flash welded to the chills.

Dr. Paschkis is correct also in stating that thermal capacity is the product of volumetric specific heat × thickness hence does not vary directly with thickness. We have spoken of the relative thermal capacities of the chill as being a function of the thickness, that is, increasing with thickness. Thus, we spoke of the thin chill having relatively less thermal capacity—the statement is only qualitative and intended to simplify discussion. We have in fact calculated the actual quantitative thermal capacities using the relationships given by Dr. Paschkis for a report (unpublished) entailing detailed study of the heat transfer processes.

It is recognized that the proper definition of "sufficient" heat capacity should be a thickness of chill or sand beyond which an increase in thickness does not result in a significant change in solidification time. The differences of interpretation revolve about the term "significant." We considered a 30 sec or 5 per cent decrease in solidification time by increasing the chill from 2½ to 4½ in. as not significant considering the great increase in volume involved. Thus, from a practical standpoint it should be recognized that severely diminishing returns are met above 2½ in. thickness.

It should be observed that only the "start of freeze" curves for the 2795 and 2880 F curves coincide. The "end of freeze" curve is delayed by the higher superheat. The same behavior was shown by the sand wall series. We have noted this effect also in gray irons, i.e., the displacement due to superheat is more pronounced for the "end" than "start" curves. The 85 F difference in superheat actually is a difference in pouring temperatures. It should be expected that the higher the pouring temperature the more temperature loss will occur during pouring.

The fact that sand walls did not indicate a thermal disturbance due to air gap formation does not preclude air gap formation. Inasmuch as the interface temperatures of sand and metal are quite close the formation of an air gap would not necessarily provide a condition which would be reflected as a thermal disturbance. The effect is dependent on the thermal properties of the mold material and the interface temperatures which exist at the time of casting-mold separation.

C. M. Adams (Written Discussion): The authors are to be congratulated upon the precision and completeness of their painstakingly gathered data as well as its excellent presentation. It seems quite certain that thermal analysis presents the only satisfactory means of investigating alloy solidification. In particular, the data constitutes an accurate measure of the effective resistance for the case of steel in a cast-iron mold.

The writer feels, however, that the concept of "air gap" formation is being badly overworked. The question of contact resistance for metals poured into metal molds is now being studied in the M. I. T. Foundry Laboratory.

In all six of the thermal analyses shown in Fig. 3 and 4 it appears that the interface temperature difference behaves in the following fashion: The temperature difference starts out large, passes through a minimum, a maximum, and then slowly decreases with time. If, for the sake of argument, the contact resistance is assumed to be fairly constant, the rate of heat flow is also passing through a minimum and a maximum. This is the sort of fluctuation which can be predicted on mathematical grounds, and does not require a discontinuous variation in the contact resistance for explanation. Essentially, a "heat wave" is being propagated into the chill, and the authors' instrumentation has been accurate enough to reproduce it.

It should be emphasized that the magnitude of the fluctuation is less than one tenth of the total temperature drop, and would not appear to reflect anything as drastic as the sudden formation of an "air gap." Also the resistances before and after the fluctuation do not appear to be materially different; that is, the disturbances in the curves are somewhat localized.

In addition, it is interesting to notice the fact that linear freezing rates are greater in the center of the casting than at intermediate points (the slope of the thickness vs time curves). This might well have some hearing on the phenomenon of "inverse chill" in cast iron.

"concept of air gap formation is being badly overworked" is very puzzling. If by this he means that an air gap does not form against chills we strongly disagree with him. There is so much evidence in the literature describing air gap effects that it must be accepted as a fact. Matushka 5 has shown that air gap formation will break an electric circuit through an ingot and mold. Paschkis has predicted it from his analogue studies. We at Naval Research Laboratory have shown in a number of ways that the air gap forms. We have shown that by mechanically maintaining the chill in contact with the solidifying metal the chill temperature will not indicate thermal fluctuations but will be heated to a temperature some 100 F higher than a chill which is allowed to separate from a casting. The only possible conditions where an air gap would not form at the chill interface would be when the metal was sufficiently hot to fuse to the chill, or when the chill was located at the bottom of the casting and gravity would maintain the casting in contact with the Mr. Adams states that, assuming a constant contact resistance

Mr. PELLINI: Mr. Adams' statement to the effect that the

Mr. Adams states that, assuming a constant contact resistance between the chill and solidifying casting the rate of heat flow is also passing through a minimum and a maximum. This also is difficult to believe. The experimental evidence as indicated by Matushka^a and experiments which we have made at Naval Research Laboratory shows an initial rapid rate of heat flow which constantly diminishes as solidification continues.

While the air gap between the intermediate size chill and the casting causes a drop of less than 10 per cent, this is significant when it is remembered that the tendency for the chill interface at this time is to rapidly rise in temperature. Contrary to Mr. Adams' statement, the resistance before and after the air gap fluctuation do appear to be materially different. Before the air gap forms the resistance to heat flow is comparatively slight and permits a rapid temperature rise. After the air gap forms the resistance is increased to the point where the temperature in the intermediate chills falls slightly and then remains essentially constant. In the largest chills the temperature continues to fall after the air gap forms. In the smallest chill which the air gap increases flow resistance across the air gap to such an extent that the casting interface temperature actually recalesces. Such marked changes would not occur unless there was a sudden change in heat flow.

As Mr. Adams points out, linear freezing rates at the center of the casting are greater than at intermediate points. This occurs despite the fact that the rate of heat flow from the ingot is less at this time than at any other time during solidification. This of course is due to the fact that the volume of liquid remaining at the center of the casting near the end of solidification is very small. Actually, volume rate of solidification is constantly diminishing with time. We agree that the rapid linear rates of solidification at the center of the casting could be related to inverse chill in gray iron.

C. E. SIMS: The authors presented an excellent paper which is very enlightening. There is every indication, moreover, that the work has been done with such care that the results may be accepted with considerable confidence.

Several years ago, I studied the freezing characteristics of some 8-in. by 8-in. ingots cast into iron molds with refractory hot tops. Freezing rates were measured by dumping at various intervals and by thermocouples placed at the center where the hot top and ingot join. In these tests, we found that superheat had very little effect on the time required for total freezing, much less than the 10 per cent to 15 per cent reported in these papers.

Superheat definitely retarded the onset of freezing but after it started, the freezing rate was faster, with the result that the end of freezing was about the same. Apparently the only way this can be accounted for is by a difference in the air gap. With very low superheat, a frozen skin will form almost immediately. This will shrink away from the mold leaving a distinct gap. With high superheat, there will be a longer period before freezing starts. A thin skin may form but ferrostatic pressure will keep it in contact with the mold. In the meantime, the mold wall is being heated with the result that, when freezing does begin, there is a very small air gap and more efficient heat transfer from ingot to mold. This produces the faster freezing rate. In another test, 500 lb of SAE 1035 steel was poured into a

² Research Assistant, Massachusetts Institute of Technology, Cambridge, Mass.

³ Assistant Director, Battelle Memorial Institute, Columbus, Ohio.

dry sand mold with walls 4 in. thick, to form a rectangular casting 9 in. by 18 in. in horizontal section and 12 in. high. A platinum platinum-rhodium thermocouple was placed at the geometric center of the casting. After pouring, the center temperature dropped to the liquidus temperature within 2 min and then remained constant until 45 min after pouring. Then during the next 15 min, it fell slowly to the peritectic temperature, indicating some ferrite crystals were forming. In the following 8 min, freezing was completed making a total freezing time at the center of 68 min.

This casting had all the indications of freezing by the progressive shell-type sequence and without a mushy stage. The principal difference between this casting and those described by the authors was the lower carbon content, and that may be sufficient to change the mode of freezing. If castings of medium carbon content do freeze by going through a mushy stage, it is difficult to see how clean, sharp, and empty shrinkage cavities are obtained.

With cast iron, on the other hand, the mushy type of freezing seems quite obvious and that may readily account for the absence of large shrinkage cavities.

J. B. CAINE: A I was somewhat perplexed by Fig. 12 in this paper in comparison with Fig. 4 in the paper "Solidification of Gray Iron in Sand Molds." See pages 425-434 in this volume. It seems to me that the rate of cooling between the liquidus and what is equivalent to the solidus range in iron is much faster than it is in steel. I wonder whether that would not explain some of the differences between feeding distances in iron and steel. It seems that the temperature gradients during solidification are steeper for iron than steel.

Mr. Pellin: Mr. Sims' question regarding the absence of shrink voids on internal "pipe" in cast irons should be considered

from a standpoint of frequency of occurrence. We know that such voids are developed in gray iron but with less frequency than is noted for steel castings. The conditions which lead to the formation of such gross voids are the same for irons or steels—for castings or ingots. The concept of "mass flow" which, while not new has been relatively little appreciated, adequately describes the condition. "Mass flow" refers to movement of mushy metal which may occur to an appreciable extent before a point of coherency is reached. Ordinarily this point is not reached until 60 to 70 per cent solid has formed. Prior to this stage of coherency the solid-liquid mush existing at high points in the casting may move downward to compensate for shrinkage at low points. A void is left behind which we regularly recognize as a pipe or shrinkage cavity.

Steel and iron castings which have been bled by forcing high pressure inert gases through the coherent mass which exists at late stages of solidification have shown the expulsion of the interdendritic liquid which existed at this time. Radiographs of sections of these castings show a fine network of solid having a lacelike appearance. This was direct evidence that a mushy condition existed almost completely through the casting section for both steel and gray iron. Figure 18 shows details of the radio-

graphs obtained.

Mr. Caine's question relative to width of the band of freezing in the case of the iron as compared to that of the steel requires consideration of the respective constitutional diagrams. For a given mold condition (chill or sand) the solidification band thickness is determined primarily by the temperature difference between the liquidus and solidus. Thus, gray iron eutectic solidification,by virtue of the narrow temperature range encompassed, results in a narrow width of solidification band. For the same metal, dendrite solidification proceeds over a broad band since a much broader range of temperatures are encompassed between the liquidus and solidus points. Essentially, the same conditions exist for austenite dendrite solidification of the steel, hence similar solidification bands are developed.

⁴ Foundry Consultant, Wyoming, Ohio.

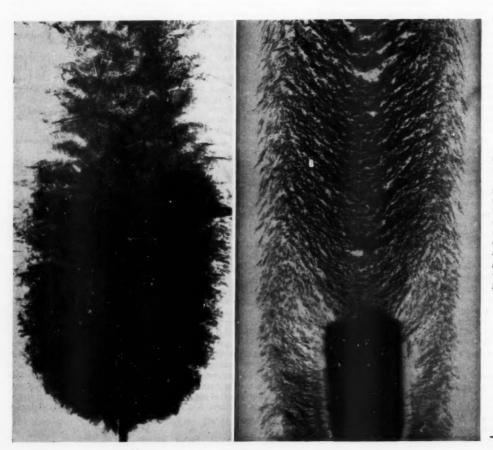


Fig. 18 – Radiographs of ½-in. thick slices from center of steel (Right) and gray iron (Left) castings which were bled by forcing high pressure inert gas through a permeable core. The thickness of solid skin and the region of intermixed liquid and solid are as predicted by thermal measurements.

* M

and

tent

an

but

proj

defo

tion

ties

ture

com a va mole ever sure with defo defo to se

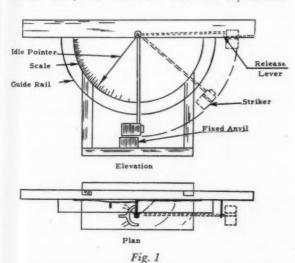
51-30

TESTING OF SAND UNDER IMPACT

By

Wm. H. Moore*

IT MIGHT BE SAID that the field of sand testing is already overcrowded with routine and special tests and that all we really need is grain analysis, clay content, permeability, moisture, compressive strength and an occasional refractoriness test. This may be true but even these tests fail to indicate completely the properties of each and every sand. Introduction of the deformation test has imparted some valuable information, particularly with regard to the tempering qualities of a sand and it is now easy to see when the moisture is in excess of that required by the bonding material. Also, the product of deformation and green compressive strength gives the toughness or resilience, a valuable aid to overcome such defects as cuts, scabs, mold drops and rat-tails. The deformation test, however has a disadvantage in that it is difficult to measure, particularly in the case of synthetic sands bonded with fireclay and cereal binders. These sands tend to deform rather than break, and the end point in both deformation and green strength is sometimes difficult to see. Also, these sands are often used dry and the



^{*} Meehanite Metal Corp., Cleveland Heights, Ohio.

deformation cannot conveniently be measured in the dry state.

In order to provide a new measure of this obviously important property of toughness, a simple piece of apparatus was constructed utilizing the principle of the familiar Izod impact test. This apparatus in its simplest form is shown in Fig. 1.

The release point of the pendulum is raised sufficiently above the horizontal so that the free-swinging pendulum will overcome frictional resistance and come to rest in a completely horizontal position.

In all initial tests using this apparatus, a standard A.F.S. 2-in. diam sand specimen was used and a striking energy of 11,550 gram-centimeters was obtained according to the weight and striking height of the pendulum chosen. As this apparatus was later abandoned, it is not thought pertinent to record its construction in any great detail.

The tests conducted with this simple apparatus were quite illuminating and paved the way for extension of this method of testing. Table 1 shows results obtained with a weakly-bonded synthetic sand as the moisture content is varied.

TABLE 1

Moisture %	Green Compression, psi	Kg/cm Green Impact
3.3	3.7	6.15
4.0	5.4	7.35
4.2	6.1	7.30
5.2	5.0	7.75
5.5	6.4	8.15
6.0	5.9	8.30
6.4	6.0	7.85
8.2	5.6	8.90
8.7	6.0	9.30
10.0	4.8	7.60

It is apparent from these results that the point of maximum compressive strength and maximum impact strength do not fall at the same moisture content. The point of maximum impact strength falls at a higher moisture content.

By taking a new natural-bonded sand and varying

hat het for not bes tal coicy av

A or gh at

er-

dior

ng es

nd

ce

li-d.

ne

ce

n

ce

n

S is e

H

0

m

T

m

at

of

m

us

te

W

de

m

be

th

co

th ab di

de

me

wa rai

holine ine the and the bod

par

Ten

40 (

equ

the degree of temper, the results shown in Table 2 are obtained.

TABLE 2

-					
Moisture %	P. No.	Green Compressive, psi	Deformation, in.	Green Impact Kg-/cm	Toughness
5.2	42	18.0	0.009	6.45	167
6.1	42	18.0	0.008	7.60	148
7.0	42	17.2	0.009	7.42	155
7.6	40	15.5	0.010	7.85	155
8.8	30	12.4	0.040	8.29	496
11.8	16	12.2	0.044	9.25	537
12.6	6	11.0	0.045	9.08	495

The column headed Toughness represents the product of the green compressive strength and the deformation. These results show that the point of maximum toughness is co-incident with the point of maximum impact strength. It may be stated therefore that the impact test is a reliable means of measuring the toughness of a molding sand and hereafter the strength obtained by impact testing will be referred to as the toughness. This assumption is further confirmed by the results obtained in a synthetic mix and recorded in Table 3.

TABLE 3

Moisture %	P. No.	Green Compressive, psi	Deformation, in.	Green Impact Kg-/cm	Toughness
5.0	49	7.2	0.012	5.20	86
6.4	46	8.6	0.014	5.30	120
7.2	44	7.9	0.021	6.60	166
8.4	42	6.7	0.007	5.85	47
9.5	40	7.4	0.008	6.00	59
12.5	9	5.6	0.007	5.84	39

Several tests were conducted on baked samples and typical results are shown in Table 4.

TABLE 4

Туре	Green Compressive Strength, psi	Dry Compressive Strength, psi		Dry Impact Strength kg-cm
Green Sand Facing	7.9	28.0	6.60	8.00
New Sand	17.2	27.0	7.42	6.60
Dry Sand Facings (Cereal Bonded)	9.0	88.0	9.25	18.50

The results are illuminating. In the case of the dry sand facing, the dry compressive strength is 9.5 times the green compressive strength whereas the dry impact strength is only twice the green impact strength. Baking a sand therefore does not increase the toughness as drastically as the compressive strength. This is not difficult to understand if the function of a binder as a wedge between sand grains is considered.

The conclusions from these initial tests may be briefly summarized:

Conclusions

1. Impact testing is a direct indication of the toughness of a molding sand.

2. The impact strength increases with increase in moisture content to a definite maximum. This maximum is usually coincident with the point of maximum resilience as measured by the deformation test and is not coincident with the point of maximum green strength.

3. The impact strength is only improved by drying in the case of sand containing a dry sand bond. In the case of ordinary green sand, the impact strength is only increased 1½ times by drying but in cereal-bonded sands, the impact strength is doubled by drying.

As a result of this work, it was decided to apply the method of measurement to the high temperature behavior of sands to obtain toughness measurements which hitherto had not been available. The form of the apparatus used was slightly modified as described below.

Until reliable indications of the worth of impact testing under temperature conditions could be established, it was decided to use a simple pilot apparatus. This is shown diagrammatically in Fig. 2. The movement of the pendulum was followed by means of light reflection from a mirror onto a greatly magnified scale and the furnace used for heating was entirely separated from the impact tester.

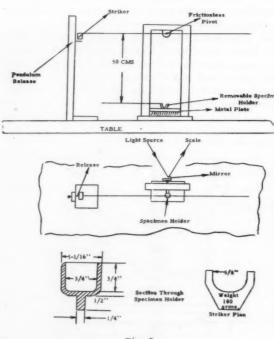


Fig. 2

Details of the testing method are somewhat as follows:

 The striking weight is 100 grams with a striking height of 50 centimeters, thus developing a total striking energy of 5000 gram-centimeters.

2. The indicator is in the form of a plane mirror

PACT

e in

axi-

axi-

test

um

dry-

ond.

igth

eal-

dry-

ply

ure

ents

of

bed

act ab-

ve-

of

ni-

en-

1

ng k-

or

fixed to a scale arranged at a suitable distance to give magnification of eight times the actual spindle movement. Although the patch of light moves rapidly over the scale, it is easy to observe its position as it momentarily comes to rest before changing direction. The scale is calibrated by means of a master chart marked out in centimeter drop divisions and graduated directly in hundreds of gram-centimeters.

3. The specimen size is of 3/4-in. diameter instead of the standard 11/8-in. specimen used for hot strength measurements. The reason for this change was to suit the available furnace tube. It was thought that using such a small specimen for a heterogeneous material as sand might give erratic results but instead, it was found that good reproducibility combined with greater sensitivity was obtained.

Figure 2 is a sketch showing the apparatus with details of striker and specimen holder which fits in a metal base plate provided with a 1/4-in. location hole.

Furnace

This consisted of a 11/8-in. diameter silica tube 6 in. in length and surrounded by a combustion chamber. Two blow lamp flames passed tangentially into the furnace chamber and the temperature was recorded by means of a base metal pyrometer, the thermocouple being placed inside the silica tube just above the sand specimen. Temperature control was difficult but tests from 300 up to 800 C (572 to 1472 F) were conducted.

Specimens were rammed by means of a double compression method which insured a constant rammed density.

Procedure for Hot Tests

The sand specimen is placed inside the metal specimen holder, and the whole is pushed vertically upwards into the furnace tube, which has already been raised to the desired temperature. The specimen and holder are held for the desired soaking time, then rapidly lowered and transferred to the testing machine when the pendulum is released. It was found that the specimen could be transferred from the furnace and tested in from 5 to 10 sec. In this period of time there could be very little temperature loss in the sand body, so that to all intents and purposes the test is conducted at the furnace temperature.

The reproducibility of tests conducted on this apparatus may be gaged from results in Table 5.

TABLE 5
Oil core sand with 2% iron oxide

Temperature of Test	Readings	Average Residual Energy	Energy Re- quired to Frac- ture in gm-cm
40 C (104 F)	38, 38, 38, 38, 37 37, 37, 37, 38, 38 37, 37, 38, 37, 37 38, 38	37.5	1250

Where the samples were tested hot the results were equally reproducible as may be seen in Table 6 and 7.

TABLE 6
Oil core sand baked in foundry oven

Temperature of Test	Readings	Average Re- sidual Energy	Energy Re- quired to Frac- ture in gm-cm
Green	37, 37, 37,	37.0	1300
40 C (104 F)	27, 28, 29,	28.0	2200
300 C (572 F)	34, 34,	34.0	1600
400 C (752 F)	33,	33.0	1700
500 C (932 F)	38,	38.0	1200
600 C (1112 F)	40, 40, 39	39.6	1040

TABLE 7

Dry sand facing with 2% iron oxide baked in foundry oven

Temperature of Test	Readings	Average Residual Energy	Energy Re- quired in Frac- ture in gm-cm
40 C (104 F)	30, 30, 28, 28,	29.0	2100
500 C (932 F)	32, 28,	30.0	2000
600 C (1112 F)	32, 31, 33,	32.0	1800
700 C (1292 F)	32, 33, 35,	33.3	1670
800 C (1472 F)	38, 36, 35,	36.3	1370

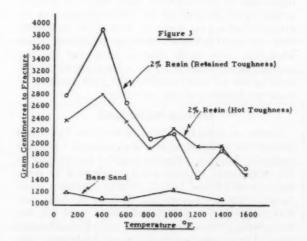
At this stage a nichrome-wound furnace was constructed and more complete tests were run under a variety of heating conditions. The soaking period was limited to 5 min as this was demonstrated to be sufficient for heating a \(^3\s/_4\)-in. diam specimen throughout its mass.

The results of all the experiments conducted are shown in graphical form only, in order to conserve space. The scales are chosen proportional to the accuracy of the measurements and each point is the average of three tests, unless erratic results indicated the need for more tests. This was seldom the case.

In each case the results obtained on the base sand without benefit of binders are given for comparison. Retained strength tests were made after allowing the sample to cool to room temperature before breaking.

Resin-Bonded Sands

Figure 3 indicates the hot toughness of the base sand bonded with 2 per cent pine resin. In this curve



an initial increase at 400 F follows melting of the resin bond and flowing around the sand grains. In

a

d

si

a

13

te

n

d

SI

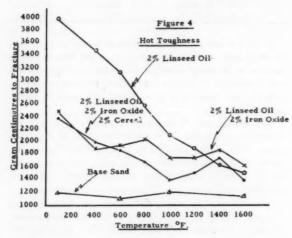
T

Units

the retained strength curve, the same increase is evident, only to a greater degree. This is due to hardening of the flowed resin on cooling to room temperature. After 400 F the toughness decreases steadily to 800 F. At 1000 F there is a peak hot strength value followed by a smaller secondary peak value at 1400 F. In other words, the strength falls more rapidly from 1000 F to 1200 F than it does from 1200 F to 1400 F indicating an opposing influence at work between 1200 F and 1400 F. In the retained strength curve this secondary peak is more pronounced. The peak values at 1000 F and 1400 F are undoubtedly due to coking of the resin bond which apparently takes place in two stages.

Oil-Bonded Sands

The behavior of the base bonded with raw linseed oil is shown in Fig. 4. The baked strength is shown

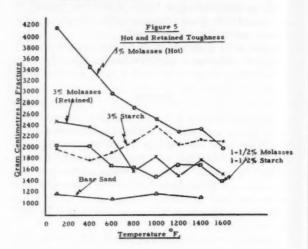


at 4000 gm-cm. The hot toughness is gradually low-ered to 600 F, after which it falls more rapidly. How-ever, even at 1600 F, the sand still has a fair degree of toughness which is in all probability due to the clay content of the mixture. Iron oxide appears to contribute to hot toughness from 800 F on whereas cereal appears to decrease the hot toughness with the exception of a peak value at 1400 F. It is possible that the cereal promotes reducing conditions which affect the iron oxide thereby increasing the hot strength beyond 1000 F. This is counteracted by collapse of the oil and cereal beyond 1400 resulting in a general lowering of the hot strength.

Molasses-Bonded Sands

Studies of these have been included in Fig. 5 and indicate the tremendous bonding power of 3 per cent molasses. There is a regular drop in hot strength as the temperature is raised and indications of an arrest value at 1000 F and 1400 F. This is confirmed by retained strength experiments. This curve shows definite peak values at 1200 F and 1400 F, probably due to a coking propensity, as in the case of resin-bonded sand. The retained strength remains moderately high over the whole temperature range, and so does the hot toughness. Combined arrowroot starch is pre-

dominantly the same up to 1000 F after which it assumes the characteristics of the starch curve. The range up to 1000 F due to molasses is somewhat lessened by the increase in strength due to the starch. Beyond 1000 F the increase due to starch is more pronounced than the loss due to molasses, so that the overall effect is an increase in toughness. Molasses as a bond imparts a good degree of strength through the 0 to 1600 F temperature range and does not collapse readily on cooling to room temperature. Starch gives an increase in hot toughness between 600 F and 1200 F. Combining starch and molasses appears to capture the characteristics of both.



These results may all be summarized briefly.

Summary

1. Impact testing of sands at elevated temperatures is a reliable means of indicating their strength at these temperatures. The test is not subject to any rate of loading defect.

2. Resin-bonded new sands containing clay show a fair degree of hot and retained toughness in the temperature range 0 to 1600 F. They exhibit peak values according to the behavior of the resin on heating.

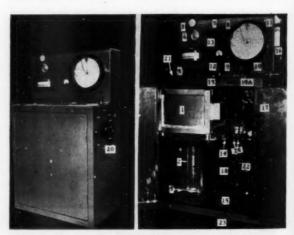


Fig. 6

PACT

t as-The

lesrch.

ore

the s as

the

pse

ives

200

ure

sses h

th

)W

ak

on

3. Oil-bonded new sands begin losing strength at and above 800 F.

4. Arrowroot starch is unusual in that it promotes a high hot toughness. This is, undoubtedly, due to its property of melting when heated.

5. Molasses gives good baked toughness and the hot toughness and retained toughness remains moderately high over the temperature range 0 to 1600 F.

The work already done with this method of testing was considered sufficiently accurate and reproducible to warrant the construction of a better designed and more substantial piece of equipment. The apparatus is described in Fig. 6 and was built for both toughness testing and gas evolution measurement. As this particular work is concerned only with toughness testing, the gas evolution equipment will not be considered.

Description of Apparatus

The sand to be tested is weighed on the balance (15) (Fig. 6) and the specimen is prepared in the double compression split specimen tube (14). The specimen measures 3/4 in. in diameter and 2 in. long.

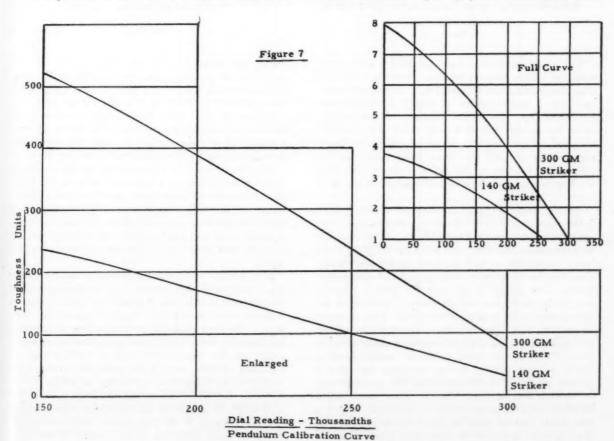
To run a test, the furnace (1) is raised to the desired temperature as indicated by the pyrometer (13). The current input to four globar heating elements (3/4-in. diam) is varied through the powerstat transformer (17) by rotating the handle (20) and reading the ammeter (12).

The specimen is inserted in a stainless steel or re-

fractory holder (not shown) and placed on the refractory post (18) which rests on a support (3) which rides on the guides (2). The support is raised to place the specimen in the heating zone at the center of the furnace and is held there for the desired soaking period. The support is then raised further so the specimen is located between the guide rollers (4) when the support is at the top of its travel. The switch (13A) is thrown to activate the electromagnet (8) which releases the pendulum and striker (5) and breaks the specimen. The swing through of the pendulum is measured by means of the eccentric cam (6) and the dial-gage (7). This dial-gage is fitted with a braking device which holds the indicator at the top of its travel and the whole is calibrated initially from the pendulum position. The calibration curve so constructed and shown in Fig. 7 allows a direct conversion from thousandths of an inch, shown on the dialgage, to toughness units.

It should be noted on the calibration curve given in Fig. 7 that the strength or toughness has been recorded in Toughness Units instead of gram-centimeters of energy required for fracture. Under this nomenclature a toughness unit equals 10 gram-centimeters of energy.

All the points recorded on the graphs in the data following are the average of three or more determinations. This was done to conform with the testing procedure developed during pilot investigations. In order to conserve space graphical results have been

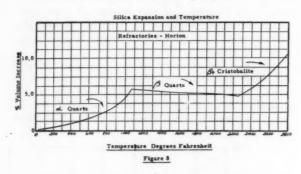


given rather than tabulations of the exact experimental data.

Effect of Heat on Molding or Core Sand

When a sand surface is subjected to heat by radiation or conduction, several things may be expected to happen as sand is a composite material. If, however, we consider only the silica grain and the clay grade in a sand at this time we can considerably simplify the whole problem.

Silica and Heat-Pure silica when heated undergoes a number of allotropic changes accompanied by changes in expansion co-efficient and by volume changes at the temperature of allotropic change. The degree of change in a sand will vary according to the relative purity and amount of silica in the sand. The important change from the foundryman's point of view is illustrated in Fig. 8 which shows the volume change during heating of pure quartz.



The a to \$\beta\$ conversion is around 1062 F and is preceded by a large volume increase. At 1062 F there is a sudden contraction followed by a slow contraction until the \$\beta\$ quartz begins to change to cristobalite. In a sand body this 1062 F change is important because the outside layers of sand reach the inversion temperature and begin contracting while the inside layers which are still at a lower temperature are expanding. This sets up disrupting forces in the sand leading to expansion cracks and fracture of the sand

Another extremely important factor in a sand body is the relative ability of the silica grains to move one against the other when expanding or contracting due to temperature changes. If a grain has no place to go when expanding, it will displace an adjacent grain, thus leading to ultimate displacement of the surface grains. The ability of the grains to move is dependent on the pore space and closeness of packing of the grains due to the degree of ramming. This closeness of packing with a given ram is largely influenced by the grain shape and the uniformity of grain distribution. Thus, an extremely uniform rounded grain is likely to pack closely during ramming and give a greater tendency to rupture when the grains change in volume due to heat. On the other hand, a less uniform angular grain would be expected to give a less dense packing with the same degree of ramming leaving more space for the individual grains to adjust

their position when expanding.

The relative ability to pack during ramming may be measured by the flowability test and a comparison between flowability and expansion supports the above assumptions. This is shown in Fig. 9 and is the result of experiments conducted by H. W. Dietert. The results show that as the flowability (or the ability to pack) increases, the expansion on heating increases. This fact can only be explained by consideration of the void spaces between the grains. It may also safely be assumed that with the same grain shape and distribution the hardest rammed sand body will exhibit the greatest expansion.

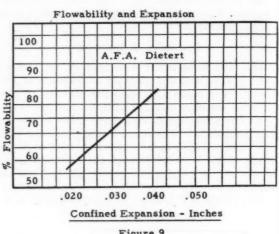


Figure 9

Clay and Heat-Due to the large number of clay types used in molding sand, the effect of heat in clay can only be dealt with in a general way. Specific results can only be confined to a single clay type.

In general, clay, when heated, loses water and exhibits a contraction in volume. The degree of contraction and the temperature at which the maximum contraction takes place varies with the clay type.

The loss of water is accompanied by the evolution of steam which sets up a frictional force or a definite pressure inside the sand body. Further than this the steam which travels inwards from the heated portion or face of the sand can condense in the interior layers of cooler sand and exert a profound influence on the strength and subsequent behavior of these interior layers.

It is possible by having the correct balance between silica grain and clay amount, and by having a good distribution of this clay through the sand, to have a resultant sand body which neither expands nor contracts. The importance of non-expanding or contracting sand has been highlighted by expansion measurements on the dilatometer and all defects such as buckling, rat-tails, veining and penetration have been attributed to differential expansion in sands.

While not under-estimating the importance of a well balanced sand from the expansion point of view, it is the purpose of the experiments described in this paper to focus attention on an equally, if not more

N

in

ei

de

di

SU

fo

tia

je

fa

if

SU

SU

fla

po

gi

in rea

a t

CT

ay

on

ve

re-

he

to

es.

of

ely

is-

iic

X

nm

11

te

16

m

rs

ie

or

d

a nte-

n

a

important, phase of sand behavior under heat influence viz. the degree of hot toughness and the rate of development of hot toughness.

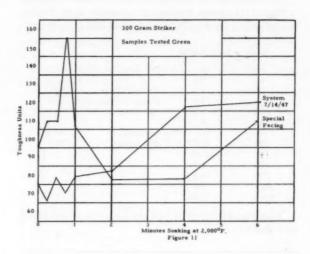
It was felt that if a sand had sufficient hot strength during the period when expansion forces and gas pressure forces were at a maximum, to resist the spalling forces developed without rupture, it would be essentially a nonbuckling sand. It might, however, be subject to veining or light rat-tails when the expansion forces were high enough to cause cracking of the mold surface.

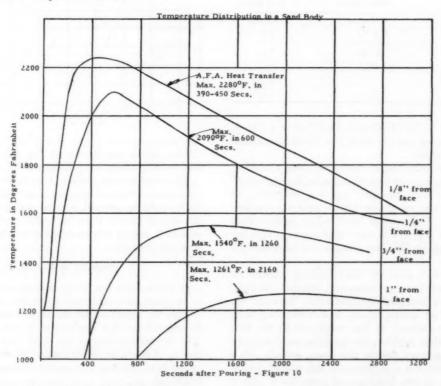
It was further felt that the critical period would fall during pouring or after gate solidification because if the ferrostatic pressure acts fully against the mold surface it should help offset any disrupting forces. In support of this argument we have the fact that large flat surfaces subject to heating by radiation during pouring are more prone to defects and that the rate of pour is extremely important. Pouring slow often gives a lot of buckling which could be eliminated by pouring fast.

Temperature of Test

In selecting the most informative temperature of test, the rate of heating of an average sand body was considered. This is brought out forcefully in Fig. 10, representing the temperatures obtained in a sand body at various distances from a 2 x 6 x 6-in. steel casting poured at a temperature of 2850 F.

experimental difficulties, it was decided to use the 2000 F range for hot toughness tests. Another reason for the selection of this temperature is that it is generally agreed that above 2000 F sand loses strength extremely rapidly. It is extremely doubtful from all the evidence we have that anything more than the extreme surface of a mold gets heated above 2000 F in a period of less than say 2 min.





It should be noted that the maximum temperature in the sand ½ in. from the metal poured at 2850 F is only 2280 F, and that it takes 390 to 450 sec to reach this temperature. Because of this and because a test temperature of 2500 F would involve greater

Buckling in Green Sand

To more directly attack the problem of buckling and allied defects it was decided to investigate the basic differences between a sand known to buckle in the foundry and one which did not buckle on the same castings. Because of the obvious importance of the first minute of shock heating, tests were run at 15 sec, 30 sec, 45 sec and 1 min of shock immersion and then followed with tests at 2 min, 4 min and 6 min, to allow full heating of the test samples throughout their diameter. Tests were conducted on green unbaked samples.

All results are an average of three or more readings. Figure 11 shows results on a system sand having the following basic grain distribution and properties:—

Mesh	System Sand	*	
6	0		
12	0.5		
20	1.1	Permeability	99
30	1.3	Green Compression	10.0
40	3.7	Moisture	3.7
50	10.2		
70	30.2	Ventability	87.0
100	25.5	Bondability	85.0
140	11.0		
200	3.1		
270	1.4		
Pan	0.4		
Clay	11.60		

This Fig. 11 refers to two samples, the mixes being as follows:-

	System at July 14 400-qt Batch	Special Facing 82-qt Batch
System Sand	1 hopper	4 pails
Ottawa No. 50	B pails	2 pails
Minco Bond	51/2 qt	2 qt
Green Bond	23/4 qt	2 qt
Permibond	1 qt	
Ajax	•	1 qt
Sea Coal		4 qt
Fuel Oil		1 qt

The special facing eliminated buckling on a 3 to 4in. section casting and on other jobs where the system sand gave trouble.

In a conventional hot strength test using a soaking period of 2 min or longer this special facing would have a lower hot strength (or hot toughness as measured on the toughness tester) than the regular system sand. This would lead us to believe that the lower hot strength is more advantageous to overcome buckling, or that hot strength has no influence on buckling.

Hot toughness tests on short soaking periods to simulate conditions encountered during pouring, however, show a remarkable difference between the two sands. The special facing or non-buckling sand gives a sharp increase in hot toughness in the first 45 sec of heating, whereas the buckling sand remains at a low toughness during this same period. This increase in the hot toughness of the special facing is put down to the influence of corn flour (Ajax) and the increase is evidently sufficient to prevent rupture of the mold surface by the expansion and gas pressure forces.

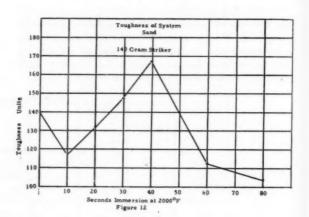
The behavior of the two sands were confirmed by ramming a 5 in. x 7 in. cope and placing these copes

3/4 in. above a pool of metal thus allowing the radiant heat from the metal to play on the cope surface. The system sand cope dropped after 30 sec of heating whereas the special facing held up until the iron was solid and afterwards. By ramming soft spots on these copes they could be made to drop on heat exposure. This again illustrates the importance of hot strength because the expansion forces and gas pressure forces would presumably be less in the softly rammed areas.

To improve the system sand in this foundry, corn flour and Western bentonite were introduced into the system sand. Hot toughness tests confirm the improvement in freedom from buckling found in the foundry.

The mix used was:-

System Sand	1 hopper
Ottawa 50	8 pails (96 qt)
Green Bond	8 qt
Corn Flour (Ajax)	6 qt
Sea Coal	3 qt
Fuel Oil	3 qt
This is a 400-quart	batch.



The results are shown in Fig. 12 and were run with a 140-gram striker and up to 80 sec of immersion.

The toughness as a whole is high and the desirable increase in 30 sec of heating is prevalent. The drop off from 30 sec onward is quite marked and indicates the desirability of pouring molds made in this sand within 40 sec or less to avoid any defects due to the strength drop-off.

Buckling in Dry Sand

Buckling does not occur as frequently in dry sand as in green sand although from the expansion point of view we would expect the same if not more trouble as some of the clay contraction is lost during drying. The dry sand surface of the mold does have a greater degree of strength and is further protected by a layer of wash so we would expect a greater resistance to buckling.

When buckling does occur in dry sand it is usually very severe and can cause a lot of trouble. A dry sand in the buckling condition and in the non-buckling condition gives the hot strength curves shown in Fig. 13. All tests were made on baked samples

Toughness Units

wi na an

Sy: Ta Oi Vo

to La as Ti

in qı th

fir be vo

re

it ho tha po

po of ru he

ng

on

on

eat

ras

tly

rn

to

mhe

n.

le

d

d

11

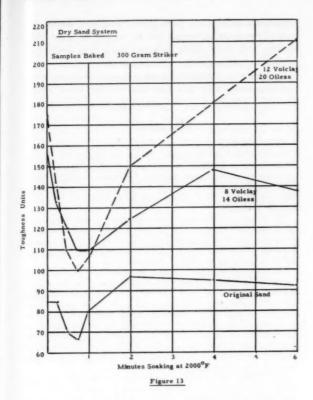
le

g.

1.

1

0



with the 300-gram striker and the sand tested is a natural-bonded system sand containing oilless-binder and a volclay addition to improve hot toughness.

The mixes under consideration are as follows:

	Buckling	Non-Buckling
System Sand, 1b	1900	1900
Taggart Gravel, lb	600	600
Oiless Binder, qt	20	20
Volclay, qt		12

The volclay addition was to increase toughness and to help consolidate the burnt out clay in the system. Later the amount of volclay and oiless binder as well as the new sand was reduced without loss in strength. The change-over of this sand eliminated the buckling troubles which had caused quite a few scrapped castings.

In Fig. 13 we may see that the toughness drops quite considerably on shock immersion, but that with the volclay addition it remains at a high enough value apparently to keep away from defects. Later results show a smaller drop in strength during this first minute as the system sand has approached a better condition by continued mulling in of the volclay.

Increasing Hot Toughness of Green Sand

In dealing with buckling in a green system sand, it was shown that Ajax corn flour increased the rapid hot strength quite markedly. There is no doubt but that many other binders could achieve the same purpose, but time and space does not allow a description of all the possibilities. A few tests, however, were run on a regular green sand to study the effect of

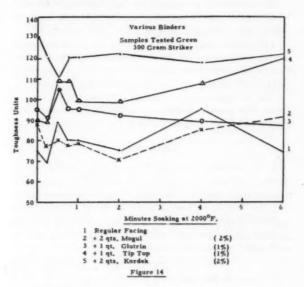
some binders. It would seem that the type of material which melts readily and retains its original form or flows around the sand grains would impart a degree of hot toughness.

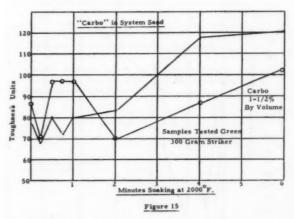
The basic sand used for the tests described is:-

Heap Sand	27	shovels
Taggart O	3	shovels
Michigan City	1	shovel
Oiless Binder	3	quarts

Thus, all tests will be influenced to some extent by the oiless binder and the behavior of the binders tested would not necessarily be the same in a different base sand.

Results with several cereal binders in this base sand are shown in Fig. 14.





Incidentally, the base sand used in these tests is liable to buckling when castings are poured slowly or where the sand is worked slightly on the wet side. A kordek binder addition of 2 quarts which is approximately 2 per cent by volume would give better buckling resistance in this sand. The 1 per cent glutrin binder also helps considerably, but of course glutrin should be watched from the moisture pickup point of view.

ot in

co

ca

ra

sa

CO

of

m

of

It should be noticed that the majority of these socalled "cushioning agents" which are presumed to stop buckling because of a decrease in expansion due to increased void spaces, also impart a degree of hot toughness to the sand during the critical first minute. This may be seen for example in Fig. 15. It is felt that this increase in strength is the factor that really stops buckling because very little cushioning action results from some of these binders if the change in permeability with ramming may be regarded as an index of cushioning.

Dilatometer Confirmation

In view of the somewhat unusual curves obtained it was decided to confirm some tests on a dilatometer.

This involved considerable experimental difficulty with the 15 and 30 seconds soaking periods, but the results were checked several times and confirmed.

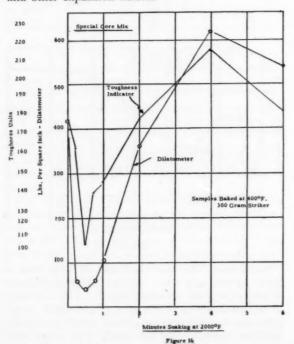
The dilatometer curves is shown on the same graph as the toughness indicator curve in Fig. 16 and the degree of correlation is excellent in that curves show peaks and valleys at the same immersion times.

The strength drop of from 420 psi to 40 psi, or from 172 gram-centimeters to 102 gram-centimeters in the first 30 seconds of heating is quite amazing.

Conclusions

1. A method of testing has been developed which gives a reliable measure of the toughness of a molding or core sand in the high temperature condition.

2. It has been found that the drop in toughness usually obtained in the first minute of shock heating is intimately associated with the formation of buckles and other expansion defects.



3. By increasing the hot toughness during the first 1 min of shock heating it is possible to overcome

expansion defects in both green and dry sands.

4. The peculiar behavior of a special core sand during a 5 min shock heating cycle has been confirmed by dilatometer measurements.

5. Many so-called cushioning agents increase rapid hot toughness and this fact may be the chief reason for their efficiency in overcoming expansion defects.

DISCUSSION

Chairman: C. A. SANDERS, American Colloid Co., Chicago.

Co-Chairman: H. W. DIETERT, Harry K. Dietert Co., Detroit. CO-CHAIRMAN DIETERT: Would you say that a 11/6-in. by 2-in. specimen would be a good specimen for impact testing? If so we could then standardize more or less on a 11/8-in. specimen for high temperature tests. Would you be agreeable to the 11/8-in. by 2-in. specimen?

MR. MOORE: In the interest of uniformity it would be well to use a standard size specimen. However, I feel there is some loss in sensitivity as you go to the larger specimens. I would presume, although I have not tried it, there would be some loss in sensi-

tivity in going to the 11/8-in. specimen.

Co-Chairman Dietert: Figure 16 showed a correlation between hot strength and hot impact toughness. The shape of the curves is identical for both methods. The hot impact toughness curve, I assume, was made using the 3/4-in. diam specimen and the hot strength curve was made using the 11/8-in. specimen. Using Fig. 16 as an index, I believe that one might assume that the sensitivity is about alike on the two sizes of specimens. From your remarks it might be in order to state that the smaller the specimen diameter the more sensitive the test due to quick heating, but your graphs do not confirm this.

MR. MOORE: We cannot draw any conclusions from the graph simply because in one case we are using a toughness test and in the other case a different method of strength testing entirely. We are not measuring the same property in those two tests. We have shown that the general trend is in the same direction, but as to the height of drop, we cannot assume that the size of the specimen has anything to do with it. We are more correct in presuming that the type of strength is more significant.

J. A. RASSENFOSS: 1 I would like to bring that up again. Your answer to Mr. Dietert's question said two things. It said that the indications of the two types of elevated temperature tests were alike and that they were different, of which, of course only one conclusion can be true. What I am getting at is this: Do we measure anything new with this test? Can there be any difference in the property you are measuring when the data curves are almost an exact parallel? Why should we introduce a new method of testing to an already formidable array if we do not get something new out of it?

Mr. Moore: Toughness is definitely a different type of strength from green compressive strength or whatever compressive strength we are talking about. I am sure we are measuring a different property. It is the same as making an impact test on a metal in order to arrive at a certain index of toughness in the metal. In sand we must apply the same line of reasoning.

By making tests in the green condition they are more reproducible because we do not have the effect of heating. We have found that impact strength is different from the green compressive strength and a relatively low toughness and vice versa. There is a definite difference in these and the properties are useful. Green compressive strength is no index of moldability. Sand is under tensile and shear forces and hence, with a respectively high compressive strength you can have drops and fall-out, but if you have a tough sand you will never get that condition. You can measure with this test about the same thing as with the deformation gage and the compressive test combined.

It is a value we previously arrived at by two methods and this represents a single method of arriving at that same answer. I think it has been shown in the past that the deformation property of a sand is quite important. I am introducing a new method of measuring the same thing.

CO-CHAIRMAN DIETERT: I have one point which might be well to clarify. We have A.F.S. toughness value now. I wonder if Mr. Moore would be agreeable to calling this what it is, impact. Then

¹ Research Metallurgist, American Steel Foundries, E. Chicago, Ind.

CT

nd

on-

oid

on

cts.

oit.

in.

we

for

to

loss

me.

nsi

he-

the

ess

and

en.

hat

om

the

eat-

aph

in

We

but

the

in

our

the

ere

one

we

fer-

rves

new

not

sive

g a

on the oroave omrsa. are lity. oecout. ion. the this opnew well Mr. hen our literature would be more correct, that is, when we speak of impact we call it impact. Thus we would have green impact, dry impact, etc. I wonder if Mr. Moore would agree to such terms in place of using "toughness," which is misleading.

MR. MOORE: I think that the use of the word "impact

strength" is probably more correct. However, it was because I did not want to introduce anything new and I wanted to express this in terms of some property we already had that I used the word "toughness" in the first place. But "impact strength" is a better way of designating it.

J. B. CAINE: 2 I wonder if Mr. Moore would elaborate on his correlation between this test and actual castings. Correlation is mentioned in a number of cases, but usually is dismissed with one line. A number of very definite statements have been made and have not been limited to iron.

MR. MOORE: It is difficult to give these details at the moment and they are not in the paper. The metal that was used in these castings was iron. However, I thought we were dealing with sand rather than metal. What we actually did, was this. We took a sand which gave buckling in the foundry under specific molding conditions and tested it. It gave a curve showing a drop in impact strength in the first minute of shock heating. We then took another sand which did not buckle under the same casting conditions and tested it. In this case the curve showed an increase in impact strength during the first minute of shock heating.

MR. CAINE: Did the sand produce buckling in every casting in that foundry?

Mr. Moore: No, only in that particular casting

Mr. CAINE: Can you describe that casting?

Mr. Moore: The castings were essentially flat type castings weighing around 100 lb and having sections up to 3 in., usually with flat copes made in green sand.

MR. CAINE: How was it gated? MR. MOORE: I realize that the method of gating and the speed of pouring and many other factors affect buckling. The object of this test was to see whether there was any method of measurement which we could do in the laboratory which would distinguish which sand, under the condition of casting, would give buckling and which would not give buckling. This measurement proved to show a difference and since that time, by careful ob-

servation we have been able to set a tentative value of somewhere around 100 toughness units being sufficient to prevent buckling.

I will agree if the gating is changed, or the type of casting, you might run a much lower value and still do a good job.

Mr. CAINE: What you are advancing, Mr. Moore, is a control test, which to you is superior to any other control test you know of. Is that right? This is not an evaluation test for sand. The more I think of it, the more important I think it is to differentiate between these two types of tests. If you say this is a control test on that particular sand then I agree with you. However, I am very skeptical that this test can be used as an evaluation test.

MR. MOORE: You are correct. It is hoped that other foundrymen will start to use it and we will be able to develop these values that are necessary rather than use it just as an evaluation

MR. CAINE: Mr. Moore, did any of your sand glaze, either in the test or on the casting, the way a steel sand glazes when in contact with molten steel?

Mr. Moore: I did not observe any glazing of the sand.

C. E. Wenninger: 8 Apparently three different types of hightemperature testing equipment and methods are now available to the industry. One is the standard Dietert dilatometer; the second is the Ross-Meehan dilatometer presented in this paper; the third is the dip-test as developed and advocated by Mr. Caine. Essentially, all are aimed at the same objective; i.e., determining the probable behavior of a sand or sand mixture at the pouring temperature under which it will be used and under the particular conditions pertaining to its use.

Mr. Moore has compared and presented data in the form of curves that show that similar curves are obtained from both types of dilatometers—and Mr. Rassenfoss has asked why should we introduce a new method of testing if we do not obtain something new out of it. I would like to carry that a little further and state that I cannot see where Mr. Moore's method has anything over the Caine dip-test when it comes to predicting the probable casting behavior of a sand. In work at the University of Kentucky we were making casting tests of eight different sands from eight different parts of the country. To test the usefulness of the dip-test, bonded samples of each sand mixture were sealed in jars and sent to Mr. Caine with only code letters on the jars to identify them. Mr. Caine was in Cincinnati; we were in Lexington. In every case Mr. Caine and his associates were able to use the dip-test to accurately predict the probable casting behavior of the respective sands before we ever made the test castings. It is because of this experience that I, too, question whether Mr. Moore's method brings forth something new that cannot be obtained from an existing method.

E. C. TROY: 4 I think this new test would have advantages if it can be accomplished with less skill and time requirements. I would like to ask the author how he would compare the time and skill required here with that of the dilatometer method.

MR. MOORE: I have found this is considerably faster and easier to manipulate than the dilatometer test. I have also taken an average of three curves in plotting these curves, because the values change so much that the curves may appear erratic, but an average is always quite reproducible.

Mr. Troy: I assume that on short time tests you have found it is an easier test to manipulate.

Mr. Moore: Very definitely. This is an instantaneous test, so the properties are not changing during the time you are applying the breaking force. I think on this basis that it is the only way you can do a satisfactory, rapid test.

⁸ Foundry Consultant, Wyoming, Ohio.

³ Director, Foundry Research and Development Section, University of Kentucky, Dept. of Mining & Metallurgical Engr., Lexington, Ky. Foundry Consultant, Palmyra, N. J.

ABSORBING THE TECHNICAL TRAINEE

By

Collins L. Carter*

DURING THE PAST 35 YEARS, American industry has operated under three different sets of conditions—peace, war and preparation for war. American industry has met successfully the widely varied economic and military demands imposed by these conditions.

This ability to produce anything, and everything, in staggering quantities, at top speed and under always changing and always growing plans, is what we have come to call "American know-how." It is a compound of brains, the cool judgment of mature experience, the imagination and courage of youth and the tireless energy of American men and women defending and maintaining a standard of living—a way of living—never enjoyed by any other people.

The foundry industry has not lagged behind in the development of "know-how." For instance, forward-looking management has for a long time recognized the vital importance of the management group in any successful operation. But, for a long time, foundrymen were like the people of whom Mark Twain wrote, who "complained about the weather, but hardly ever did anything about it."

Need Management Continuity

It is well known that labor, equipment, and materials are of little value without management thought and direction. Also, it is well known that the membership of any management group is subject to losses as inevitable as death and taxes. Illness, accident, the toll taken by the passing years, the greener grass in some distant field of work, the unexpected and unexplainable sudden failure of a man to measure up to the demands of his job—all these continually threaten the continuity of management policy and practice.

When these losses occur there should be, and frequently is, a man in the organization trained and ready to take his place. But this man's promotion means that his former place must be filled. How and by whom? Not by raiding some other concern within the industry, a procedure frowned upon by polite foundry society—

especially by those raided. The foundry industry answered with another question. Why not recruit replacements from trainees within the respective plants?

A successful organization, built to withstand the ravages of time and to keep abreast of modern techniques, cannot afford to be without a trainee program. For instance, when suddenly forced to bring in an untrained supervisor from the outside or step up an inexperienced man from within, serious personnel problems can result—even to the extent of strikes. Often such dislocations are much more costly than the annual expense of any trainee program. Furthermore, obsolescence in foundry methods and equipment point up the necessity for injection of new blood, technically trained to wrestle with such problems. These facts are equally true whether the firm is small or large.

How many trainees should be added each year? Most firms have one member of management to approximately 15 employees and historical figures show that management mortality is in the neighborhood of 10 to 12 per cent per year. As an example, a company employing 600 people might have approximately 40 members in the direct management team. If 10 per cent were

Frank Tobakos, whose father worked in a foundry for 36 years, molding during his trainee days at Albion Malleable Iron Co. He started at Albion Malleable, where he is a foreman, after graduation from Rensselaer Polytechnic Institute in 1949.



0

m

u

of

CO

u

pl

of

he

ve

sp

De

de

th

ar

ar

th

SC

ne

ea

President, Albion Malleable Iron Co., Albion, Mich.

lost each year, it would mean, by allowing for some shrinkage within the trainee group itself, an addition of 6 to 7 trainees each year. This generalization may vary with different companies dependent upon factors such as the average age of individuals in the management group, etc.

The two main sources for recruitment of excellent

material for trainee programs are:

1. Mechanical, industrial, electrical, and metallurgical graduates of engineering colleges throughout the country, particularly those who have had some summer work experience and the opportunity to attend engineering schools which are members of the Foundry Educational Foundation. In a relatively short period of three years the Foundation has made available to our industry some 300 young graduate engineers, who have been trained in foundry techniques.

2. Qualified individuals selected from the company's

own work force.

n-

e-

V-

es,

n-

d

r

ns

h

X.

le

d

ly

ıt

0

e

It is the practice in the author's company to balance the trainee program with an equal number of college graduates and practical shop men. In close association, the college graduate trained in theory, but without much actual work experience can learn much from the practical shop man. Conversely, the shop man can pick up much knowledge from the college trained engineer. This combination is also an inspiration to capable young men who are excellent employees but who have

not had a college education.

Do not expect the graduate engineer to come to the foundry after June graduation looking for a job. Statistics show that for the next few years there will be relatively few graduate engineers who have had the advantages of the Foundry Educational Foundation or of summer foundry work experience. The Foundry Educational Foundation periodically places in the hands of its members a list of prospective graduates. Personal contact with the students and placement officers of the universities is generally made several months before graduation. Often the students make company financed plant visitations. At these plant interviews, some companies require prospective employees to take a battery of psychological tests which measure mental ability, social intelligence, and certain personality traits.

Training Programs Differ

In the foundry industry there seems to be a wide variation in the type and duration of trainee programs; however, each program has as its goal the practical development of the trainee to undertake supervisory responsibility. Various trainee programs extend over a period of six months to two years. In some cases, the degree of previous experience of the individual governs

the length of this period.

Some big companies where larger groups of trainees are employed, have a program embracing rather rigid and definite schedules. For instance, in one company the practical experience consists of 20 weeks spent in the core, melting, and foundry departments and in scrap elimination (four weeks each) and in the annealing and the inspection departments (two weeks each). Twice each week this practical training is augmented by a two-hour class of theoretical training in: job of first line management, organization structural

policies and procedures, department functions and inter-relationships, management skill in working with people, management skill in job training and job improvement.

In smaller companies a more flexible program can be employed. One company which has only five or six trainees at a time sets up training programs which gen-

erally adhere to the following schedule:

 Initial indoctrination in all departments, including administrative for overall conception of the business. 2 weeks.

2. Assignment to various operating departments for work experience. 12-36 weeks.

3. Reorientation. One week.

- 4. Engineering department with assignment of specific projects to get experience in organization of effort and to develop a feeling of accomplishment. 12-36 weeks.
- 5. Reassignment to certain operating departments for minor supervisory experience. 3-12 weeks.

6. Permanent assignment to one of five divisions of

the company.

Total time for the program is 8 to 22 months. The time spent in each assignment is dependent upon the previous experience, ability, and progress of the individual trainee.

In this same plant, a Committee on Trainees is responsible for all trainees and training programs. The committee consists of the manufacturing manager, works manager, chief engineer, and personnel director. For every trainee a rating report is prepared weekly by his supervisor. The committee periodically reviews these reports. Also, each week the trainee turns in a written report to the committee. The trainee's report may include any observation or comment whatsoever that he wishes to make in respect to his work.

Stress Actual Work Experience

It should be noted that in the curricula of these two training programs actual work experience is heavily emphasized. The value of actual work experience lies in teaching the trainee how to use his hands. Regardless of the degree of mechanization in this country, at one stage or another everything we have is made with somebody's hands. Further, if the trainee knows what work is, he can later determine who much work it is reasonable to expect from a man and the trainee will command more respect from the people whom he will presumably someday supervise.

An essential phase of training which must not be overlooked is learning how to get along with people by leading them. The Simon Legree has no place in today's American industry. The trainee must learn that driving will not yield the required results. He must also recognize the importance of intelligent cooperation with all. These points should be stressed through-

out any curriculum.

Under all conditions, it is extremely important that a trainee placed on any phase of the program, not be forgotten by management. If the individual is progressing satisfactorily, he should be recognized and encouraged. Conversely, if it is noted that a trainee is not progressing, additional attention should be given his shortcomings. However, a danger must be pointed

following from the cia character the blockele ties

and

ach of in tha

per cia ter im

Eng lurg T Illir part

out that too much recognition and too much attention can have an adverse affect on the trainee or on the attitudes of others toward him, particularly on the part of those employees who have not had formal training and the advantages of a technical engineering education. "Hot shot" attitudes can be and must be prevented. In justice, if it is apparent that a trainee is not suited for the foundry industry or does not fit in the organization, it is only fair that management advise him accordingly, and the sooner the better for all concerned. He should be given an opportunity to seek employment elsewhere.

Another important factor in the success of any trainee program is the working agreement and practical acceptance of this program by the union. If some trainees have been picked from the ranks of the union, reaction to the program is much more favorable. It is recommended that an understanding be reached previously to the installation of a trainee program. This understanding should be included in the labor con-

tract. The trainee must have some production work experience. The greatest stumbling block is the amount of replacement of direct or indirect production workers who are members of the union. One company had written into its labor contract that the number of trainees on production work at any one time could not exceed one per cent of the factory payroll.

At the completion of the training program, every trainee has for his goal a permanent assignment as a supervisor or technician in the division of the business in which he is best suited. This assignment is often influenced by openings available at the time. Often, due to emergencies, the permanent assignments are made before the training program is completed. In any case, after permanent assignment the responsible executive of the division should see that a certain amount of training is continued. With such an adequate and well-rounded training program, the company will have a thoroughly trained supervisor whose abilities will assure him of consideration for further promotion.

REFINING SECONDARY COPPER ALLOYS

By

Marvin Glassenberg,* L. F. Mondolfo** and A. H. Hesse***

Introduction

THE OBJECTIVES of these experiments are twofold: (1), to determine the rates of removal of common impurities from secondary copper alloys by means of air injection; (2), to investigate the effect of commercial oxygen on these rates of removal.

The use of injection of air to remove impurities from secondary copper alloys is common in commercial practice. The normal procedure is to melt a charge with a composition approximating that of the desired alloy. When the charge is molten, compressed air is blown into it to oxidize the impurities. The resulting compounds rise to the surface where they form a slag which is periodically removed. The blowing of air is continued until the undesirable elements are removed. After removal of the impurities, the necessary alloying elements are added to bring the melt to the desired chemical composition and then the melt is poured into ingot molds.

The substitution of commercial oxygen in place of air for more rapid refining is theoretically sound, since a faster rate of oxidation of impurities should be achieved, thus reducing the operating time. The use of oxygen injection in the refining of steel and also in various combustion processes, strongly indicates that oxygen can also be successfully employed for refining copper alloys.

nt k d of ot

a SS nie

le

e,

/e 1-

11

Previous Studies in the Field

Although air injection for refining secondary copper alloys has been successfully utilized in commercial practice for many years, very little written material on the chemical and metallurgical aspects of impurity removal has been published. "Trade secrets" and the fact that air injection "works," probably hindered publication of this information.

No published information on oxygen injection in brass refining was found. Some papers1, 2, 3 discussing oxygen injection in the refining of ferrous metals have been published, but these applications are of little value in copper refining.

Experimental Technique

Equipment and Melting Practice—The metal used for all the experiments came from a homogeneous melt which had been produced as one heat by melting scrap in a rotary furnace and casting it into ingots weighing approximately 20 lb. The chemical analysis of the heat appears in Table 1. One hundred and fifty pounds of metal was used for each experi-

TABLE 1-ANALYSIS OF ORIGINAL MELT

	Element	Percentage
	Copper	81.42
	Tin	3.39
	Lead	
	Nickel	0.76
	Sulphur	0.12
	Antimony	
	Aluminum	
	Iron	
	Silicon	
	Manganese	
	Phosphorus	
	Zinc (by difference)	
Zinc		

A No. 60 silicon carbide crucible, having a capacity of 180 lb, was used as a reservoir for the metal. Starting with a cold oil-fired furnace, the brass required an average of 60 min to reach 2100 F.

After the metal was brought to a temperature of 2100 F, a sample was taken and approximately 13 oz of glass were added to the bath to create a cover. About 10 min were required to melt the glass after which another sample was taken and the subsequent refining begun. Chemical analysis indicated that the addition of the glass and the 10-min lapse had little effect on the composition of the alloy.

^{*} Graduate student, ** Associate Professor of Metallurgical Engineering, Illinois Institute of Technology, and *** Metallurgical Engineer, R. Lavin & Sons, Inc., Chicago.

This paper is abstracted from a thesis submitted to the Illinois Institute of Technology by Marvin Glassenberg in partial fulfillment of the requirements for the degree of Master

Method of Sampling—Sampling of the metal was obtained by immersing a preheated ladle, 3 in. in diameter, into the bottom of the crucible and stirring as the ladle was brought to the surface. Iron contamination was held to a minimum by pouring the sample within 20 sec after the immersion of the ladle.

The metal was poured into a cast iron mold to form a 2½-in. disc weighing approximately ½ lb. Drillings were taken all over one face of the disc in order to eliminate errors due to segregation in the sample. These drillings were used for chemical analysis. The undrilled face was machined and conditioned for a spectrographic analysis of the sample.

Temperature Control—In order to eliminate the temperature variable, each experiment was conducted at a temperature of 2100 F. No automatic device was available for maintaining a constant temperature in the furnace; therefore, manual control of combustion was necessary. The temperature was constantly recorded with an automatic potentiometer, which showed a maximum deviation of \pm 25 F.

The continuous recording of temperature necessitated the use of a protective tube for the chromelalumel thermocouple which had to be in contact with the metal during the complete experiment. A silicon carbide tube proved successful.

Injection Technique—Figure 1 illustrates the apparatus used in measuring and injecting the air or oxygen into the bath of metal.

The oxygen or air was drawn from a single commercial cylinder equipped with a three-stage regulator to maintain a steady flow. Commercially dry air and commercial oxygen were used.

The cylinder rested on a beam balance having a capacity of 200 lb with \(^{1}\sqrt{4}\)-oz sensitivity. When a

sample of metal was taken, recordings of the following data were made: (a), time; (b), weight of gas injected; (c), temperature of metal bath. The complete procedure required about 1 min.

The gas was conducted from the cylinder to the immersion lance by means of a rubber hose and a steel pipe. A mercury manometer was inserted in the 7½-ft rubber hose line to estimate the flow of the gas and indicate leakage after the refining operation was started. The hose line also included a thermometer to indicate the temperature of the flowing air or oxygen. A quick acting valve and copper tube were inserted between the rubber hose and the 5 ft steel pipe to which the lance was fastened. The depth of the lance in the molten bath was controlled by an adjustable vertical stand to which the horizontal steel pipe was attached.

Difficulty was encountered in obtaining material for the lance that would withstand the critical conditions of the experiments. A total of 11 experiments was performed before a suitable material was found and a satisfactory lance designed.

Steel pipe coated with ceramic material failed because the pipe was consumed in a short time. A graphite tube underwent rapid deterioration and, in addition, chemical analysis of successive samples indicated that the graphite inhibited oxidation.

A sillimanite tube indicated suitable properties, but lack of proper means of attaching the tube to the lance assembly limited further use of this material.

A quartz tube disintegrated after a short period of submergence and absorption of brass into the quartz material was noticed.

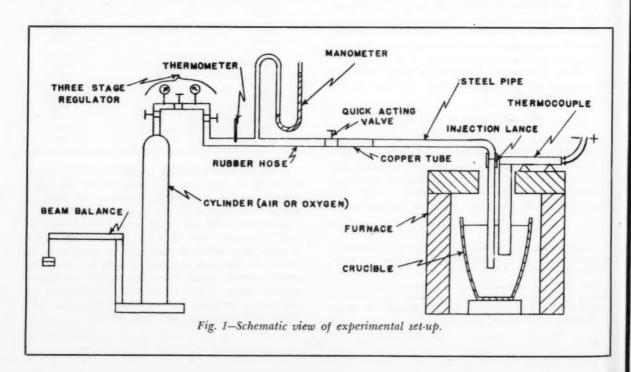
Finally, a porcelain material was found which showed satisfactory chemical and physical properties for air injection. After several preliminary tests with

20

Fig exp I det

of t tion zirc cau

duc



OYS

OWgas mthe da in of eraering ibe ft oth an eel for dints nd beohdited

ies, to ial. of rtz ich ies ith

TABLE 2-FIRST EXPERIMENT WITH AIR (A-1)

	wei	Total ght of ir in-	Mean -						Chemi	cal Com	positio	n, Per Ce	ent									
	Time jected min oz		temp.	Copper	Lead	Nickel	Tin	Sulphur	Anti- mony	Alumi- num	Iron	Silicon	Manga- nese	Phos- phorus	Zinc by diff.	Zinc analyze						
00			2145	81.80*	3.34	0.77*	3.54	0.14	0.12	0.07*	1.22	0.335*	0.08*	0.12*	8.46	8.68						
0			2110	81.84	3.32	0.77	3.53	0.14	0.12	0.06	1.15	0.335	0.073	0.115	8.55	8.60						
.1	4	1	2112	81.95*	3.32*	0.77*	3.50*	0.14*	0.12	0.05*	1.15*	0.325*	0.068*	0.11*	8.50							
2	11	2	2110	82.08	3.30	0.76	3.48	0.14	0.12	0.04*	1.15	0.317	0.065	0.109	8.44							
3	16	3	2095	82.14*	3.32*	0.77*	3.50*	0.14*	0.12	0.03*	1.15*	0.313*	0.062*	0.10*	8.36							
4	22	4	2100	82 24	3.29	0.76	3.53	0.14	0.12	0.02*	1.14	0.293*	0.059*	0.112	8.30							
5	27	5	2100	82.33*	3.32*	0.77*	3.50*	0.14*	0.12	0.01*	1.14*	0.289	0.056	0.092	8.23							
6	33	6	2105	82.52	3.35	0.76	3.53	0.14	0.12	**	1.13	0.28*	0.051*	0.108	8.02	7.85						
7	38	7	2090	82.53*	3.32*	0.77*	3.50*	0.14*	0.12		1.12*	0.273	0.049	0.091	7.99							
8	43	8	2110	82.61	3.29	0.75	3.53	0.14	0.12		1.15	0.257*	0.043*	0.09*	8.02	7.72						
10	49	9	2115	82.86	3.33	0.75	3.48	0.14	0.12		1.09	0.234	0.038	0.088	7.87							
11	54	10	2090	82.83	3.31	0.77	3.53	0.14*	0.12		1.07	0.231*	0.035*	0.085*	7.88							
12	59	11	2105	82.93*	3.32*	0.77*	3.50*	0.14*	0.12		1.11	0.210	0.031	0.084	7.79							
13	65	12	2120	83.10	3.32*	0.77*	3.53	0.14	0.12		1.08	0.208*	0.029*	0.081	7.62							
14	70	13	2090	83.09	3.32*	0.77	3.53	0.14	0.12		1.03	0.194	0.025	0.091	7.69	6.67						
15	76	14	2095	83.22*	3.32*	0.78*	3.50*	0.14*	0.12		1.01	0.187*	0.023*	0.077	7.62							
16	81	15	2100	83.32*	3.32*	0.78*	3.50*	0.14*	0.12		1.03	0.175	0.021	0.079	7.51							
17	86	16	2100	83.42	3.33	0.77	3.48	0.14	0.12		0.98	0.165*	0.02*	0.077*	7.50	* *						
18	92	17	2105	83.52*	3.32*	0.78*	3.50*	0.14*	0.12		0.98	0.159	0.018	0.075	7.39							
19	99	18	2110	83.50	3.32*	0.79	3.48	0.14	0.12		0.94	0.144*	0.015*	0.073*	7.48							

TABLE 3-SECOND EXPERIMENT WITH AIR (A-2)

	We	Total eight of air in	Mean -						Chemi	ical Com	positio	n, Per Ce	nt		1	3
Sample	Time	jected	temp.	Copper	Lead	Nickel	Tin	Sulphur	Anti- mony	Alumi- num	Iron	Silicon	Manga- nese	Phos- phorus	Zinc by diff.	Zinc
00			2125	81.81	3.27	0.80	3.50*	0.13	0.11*	0.070*	1.24	1.352*	0.091*	0.116*	8.51	8.57
0			2135	81.80	3.34	0.80*	3.53	0.13	0.11*	0.061	1.20	0.352	0.089	0.114	8.48	8.72
1	7	0.75	2115	81.87*	3.32*	0.80*	3.53*	0.13*	0.11*	0.058	1.20	0.357	0.086	0.115	8.42	
2	13	1.25	2080	81.93	3.32	0.80	3.53*	0.13*	0.11*	0.050*	1.22	0.343*	0.083*	0.110*	8.37	8.33
3	23	2.25	2125	82.25*	3.32*	0.80*	3.53*	0.13*	0.11*	0.046	1.20	0.335	0.080	0.105	8.09	
4	33	3.50	2080	82.36	3.28	0.80*	3.53*	0.13*	0.11*	0.030*	1.14	0.320	0.073	0.104*	7.85	7.95
5	43	5.00	2110	82.61*	3.32*	0.80*	3.53*	0.13	0.12	0.026	1.19	0.290	0.061	0.100	7.82	
6	54	6.50	2135	82.93	3.34	0.80*	3.53*	0.13*	0.12*	0.010	1.12	0.273	0.051	0.093	7.51	7.31
7	64	8.25	2080	83.14	3.32*	0.80*	3.53*	0.13*	0.12*	0.010	1.09	0.261	0.041	0.093	7.47	7.60
8	79	10.50	2110	83.19	3.32*	0.80*	3.53*	0.13*	0.12*		1.07	0.241	0.034	0.086	7.48	7.45
10	96	13.00	2110	83.45	3.32	0.80	3.53	0.13*	0.12*	* *	1.01	0.202	0.020	0.079	7.34	7.21
11	115	16.00	2105	83.59	3.32*	0.80*	3.53*	0.13	0.12*		0.98	0.177	0.020	0.074	7.26	7.17
12	132	18.75	2120	83.77	3.33	0.80	3.53*	0.13*	0.12*		0.90	0.145	0.014	0.069	7.20	
13	146	20.75	2110	84.05	3.32*	0.81*	3.53*	0.13	0.12		0.91	0.111	0.009*	0.059	6.95	6.90
14	159	22.75	2105	84.33	3.32*	0.81*	3.53*	0.13*	0.12*		0.77	0.067	0.006	0.049	6.86	6.72
15	172	24.75	2110	84.71*	3.32*	0.81*	3.53*	0.13	0.12*		0.72	0.056	0.005*	0.042	6.55	
16	185	26.75	2100	84.97	3.30	0.82	3.55	0.13*	0.12*		0.59	0.022	0.003	0.024	6.47	6.27
17	202	29.50	2105	85.41	3.32*	0.82*	3.53*	0.13*	0.12		0.43	0.003		0.007	6.23	6.07
18	217	32.00	2115	85.85	3.34	0.82*	3.53*	0.13	0.12	* *	0.28	0.002		0.004	5.92	
19	236	35.00	2090	86.56	3.30	0.82	3.53*	0.12*	0.12*	**	0.31	0.001*	**	0.004*	5.24	5.14
			removed) 86.57 ed by int	3.33 ernolati	0.82*	3.53	0.12	0.12		0.22	0.001		0.004	5.28	

porcelain, the construction of the lance as shown in Fig. 2 was found to be suitable and employed in all experiments with air.

In the first experiment with oxygen, an accelerated deterioration of the porcelain tube and a rapid rise of temperature were noted. The same rapid deterioration took place with silicon carbide, sillimanite, and zirconia tubes. This rapid deterioration was probably caused by the strong exothermic reactions which produced very high temperatures in the immediate vicinity of the tube causing it to fuse and disintegrate. In

addition, the silicon carbide tube reacted with the oxygen thereby inhibiting the refining, as previously noticed with the graphite tube. However, by using a fast rate of injection, sufficient data were obtained in some of the runs before the experiment was stopped by the failure of the tube.

Air injection resulted in the inert gases bubbling to the surface of the metal bath, causing the metal to spatter out of the crucible. The rate of flow was therefore adjusted to cause minimum loss through spattering. In the case of oxygen injection, the liquid

TABLE 4-THIRD EXPERIMENT WITH AIR (A-3)

	W	Total eight of air in-	Mean -						Chemi	cal Com	position	n, Per Ce	ent			
		jected	temp.	Copper	Lead	Nickel	Tin	Sulphur		Alumi- num	Iron	Silicon	Manga- nese	Phos- phorus	Zinc by diff.	Zinc analyzed
00			2160	81.75	3.28	0.77*	3.53	0.14	0.12*	0.07*	1.24	0.360	0.082	0.118	8.54	8.25
0			2115	81.94	3.32	0.77*	3.53	0.13	0.12	0.06*	1.26	0.352	0.080	0.117	8.32	8.27
1	8	1.00	2115	82.02	3.32*	0.77*	3.53*	0.13*	0.12*	0.05*	1.22	0.332	0.078	0.114	8.32	
2	25	2.25	2110	82.23	3.32	0.77*	3.53*	0.13*	0.12*	0.04*	1.24	0.346	0.065	0.113	8.09	8.13
3	35	3.75	2075	82.42	3.32*	0.77*	3.53*	0.13*	0.12*	0.03*	1.22*	0.311	0.059	0.108	7.98	
4	49	5.25	2115	82.61	3.31	0.77*	3.53*	0.13	0.12*	0.02*	1.19	0.291	0.053	0.101	7.88	7.90
5	64	7.25	2095	82.70	3.32*	0.77*	3.53*	0.13*	0.12*	0.01*	1.15	0.282	0.044	0.094	7.85	7.80
6	78	9.25	2110	82.92	3.32	0.77*	3.53*	0.13*	0.12*		1.13	0.264	0.040	0.092	7.68	
7	91	11.25	2105	83.02	3.32*	0.77*	3.53*	0.13*	0.12		1.19	0.232	0.028	0.088	7.57	
8	104	13.25	2080	83.24	3.32*	0.77*	3.53*	0.13*	0.12*		1.09	0.224	0.023	0.085	7.47	
10	123	16.00	2105	83.53	3.32*	0.77*	3.53*	0.13*	0.12*		1.12	0.202	0.020	0.083	7.18	* *
11	140	18.50	2105	83.74	3.34	0.77*	3.48	0.13*	0.12*		1.03	0.175	0.013*	0.077	7.12	7.05
12	158	21.00	2100	84.01	3.32*	0.77*	3.53*	0.12*	0.12*		1.02	0.136	0.010	0.070	6.89	6.66
13	176	23.50	2085	84.34	3.29	0.78*	3.53*	0.12	0.12		0.93	0.108	0.008*	0.058	6.71	6.69
14	193	26.00	2100	84.58	3.32*	0.78*	3.53*	0.12*	0.12*		0.85	0.087	0.006	0.048	6.56	6.56
15	210	28.50	2100	84.83	3.32	0.78*	3.53*	0.12*	0.12*		0.73	0.060	0.004*	0.042	6.46	6.27
16	227	31.00	2090	85.06	3.32*	0.78*	3.55	0.13	0.12*		0.72	0.040	0.003	0.035	6.24	6.24
17	244	33.50	2105	85.17	3.33	0.78*	3.53*	0.12*	0.12*		0.71	0.022	0.0005	0.030	6.19	6.15
18	259	35.50	2090	85.46	3.32	0.78*	3.53*	0.12*	0.12*		0.64	0.008		0.020	6.00	5.87
19	274	38.00	2080	85.51	3.33	0.79	3.53*	0.12	0.12		0.54	0.003		0.012	6.04	5.88
				ned by into		on.										

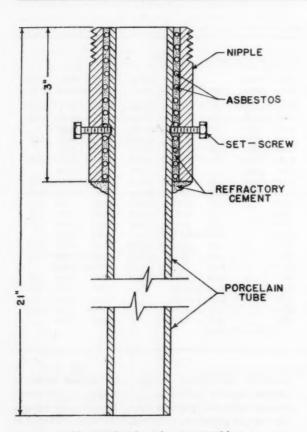


Fig. 2-Sketch of lance assembly.

metal was quiescent and no bubbling was discerned, thus permitting much faster rates. The rate of injection, however, had to be controlled because too rapid a flow would have resulted in a sharp temperature rise.

Of the two oxygen experiments which were successful, the first was completed with two seperate porcelain tubes at a relatively rapid rate of injection. The second experiment was accomplished with a long sillimanite tube at a slower rate. The tube was lowered into the bath as the end deteriorated.

Analysis—The spectrograph was used to identify the elements in the alloy and standard wet procedures were used for quantitative determination of the elements found with the spectrograph, namely, copper, tin, lead, nickel, iron, manganese, silicon, phosphorus, sulphur, antimony, zinc and aluminum. The spectrographic method was also used quantitatively for aluminum to check the values obtained by the wet method.

Results

The results of three experiments, designated as A-1, A-2 and A-3, involving the use of the air injection technique, were chemically analyzed. Similar experiments, designated as O-1 and O-2, involving oxygen injection were also chemically analyzed. The results for each experiment are tabulated in Tables 2 to 6, inclusive, which list the following data: concentrations of the elements; total weight of air or oxygen injected; time; and the mean temperature between successive samples.

The data in Tables 2 to 6, which includes three experiments with air and two experiments with oxygen, were averaged. To avoid the difficulty created by the variance in initial concentrations of the various elements, the following formula was applied:

Percentage Retained =
$$\frac{\text{Concentration}}{\text{Initial Concentration}} \times 100$$

The results, presented graphically in Fig. 3, show the sequence of removal for each element and the in-

Me

cre res phe esse

eve off Ver wei mo

line rate also taii fere

by tion

pro

ed

he lied

he res leer, us, romd.

on

ri-

en

lts

6,

a-

en

en

ee

V-

ed

us

00

he

n-

TABLE 5-FIRST EXPERIMENT WITH OXYGEN (O-1)

	W	Total eight of air in-	Mean -						Chem	ical Com	positio	n, Per Ce	ent			
Sample		jected oz	temp. ∘F	Copper	Lead	Nickel	Tin	Sulphur	Anti- mony	Alumi- num	Iron	Silicon	Manga- nese	Phos- phorus	Zinc by diff.	Zinc analyze
00	* *		2105	81.70	3.31	0.77*	3.52	0.12	0.12	0.07*	1.17	0.353	0.087	0.120	8.66	8.56
0			2105	81.80	3.35	0.77*	3.55	0.12	0.12	0.06*	1.16	0.352	0.086	0.115	8.52	
1	4	2.50	2105	82.10	3.31	0.77*	3.55*	0.12*	0.12*	0.05*	1.14	0.332	0.079	0.109	8.32	
2	7	5.25	2085	82.31	3.33	0.77	3.55*	0.12*	0.12*	0.04*	1.09	0.314	0.071	0.097	8.19	8.15
3	11	8.75	2110	82.60	3.35	0.77*	3.57	0.12	0.12	0.03*	1.06	0.281	0.061	0.091	7.95	7.95
4	15	11.75	2120	82.77	3.33	0.77*	3.55*	0.12*	0.12*	0.02*	1.03	0.245	0.050*	0.077	7.92	
5	19	14.75	2110	83.12	3.38	0.77*	3.55*	0.12*	0.12*	0.01*	0.92	0.210	0.039	0.068	7.69	7.64
6	23	18.25	2100	83.47	3.34	0.77*	3.55*	0.12	0.11		0.82	0.165	0.028	0.055	7.57	7.62
000**	28	21.50	2120	83.89	3.36	0.78*	3.55*	0.12*	0.12*		0.77	0.125	0.024	0.045	7.22	7.28
7	32	25.50	2080	84.35	3.36	0.78	3.55	0.12*	0.12*		0.63	0.084	0.011	0.033	6.96	6.96
8	41	29.00	2055	84.71	3.43	0.78*	3.55*	0.12*	0.12		0.54	0.050	0.005	0.024	6.67	6.89
10	54	32.25	2090	85.03	3.39	0.79	3.55*	0.12*	0.12*		0.47	0.030	0.001	0.019	6.48	6.68
11	71	36.00	2100	85.51	3.42	0.79*	3.55*	0.12*	0.12		0.39	0.018	0.001*	0.017*	6.06	6.04
12 (A	fters	lag was	removed	1) 85.44	3.35	0.79*	3.52	0.12	0.12		0.38	0.022	0.002	0.017	6.24	5.94
* Value	s wh	ich wer	e obtair	ned by in	terpolat	ion.										
				re this san												
				oz per mi												

TABLE 6-SECOND EXPERIMENT WITH OXYGEN (O-2)

	We	Total eight of air in-	Mean -						Chemi	cal Com	positio	n, Per Ce	nt			
Sample	Time	jected	temp.	Copper	Lead	Nickel	Tin	Sulphur	Anti- mony	Alumi- num	Iron	Silicon	Manga- nese	Phos- phorus	Zinc by diff.	Zinc analyze
00			2100	81.79	3.31	0.75*	3.56*	0.12*	0.11*	0.07*	1.21	0.360	0.087*	0.018*	8.51	8.27
0		* *	2120	81.87	3.34	0.75	3.56	0.12	0.11	0.06*	1.20	0.350	0.085	0.117	8.44	8.34
1	6	1.00	2105	82.03	3.28	0.75*	3.56*	0.12*	0.11*	0.05*	1.21	0.350	0.083	0.115	8.34	
2	11	2.00	2115	82.09	3.30	0.75*	3.56*	0.12*	0.11*	0.04*	1.19	0.332	0.080	0.111	8.32	8.29
3	16	3.00	2115	82.21	3.32	0.75*	3.56*	0.12*	0.11*	0.03*	1.19	0.318	0.078	0.105	8.21	
4	22	4.50	2090	82.31*	3.32*	0.75*	3.56*	0.12*	0.11*	0.02*	1.13	0.305*	0.074	0.103*	8.20	
5	27	6.00	2115	82.47	3.29	0.75*	3.56*	0.12*	0.11*	0.01*	1.11	0.286	0.068	0.097	8.12	8.11
6	33	8.25	2105	82.66*	3.32*	0.75*	3.56*	0.12*	0.11*		1.07	0.268*	0.062*	0.091*	7.99	
7	38	10.25	2095	82.77	3.39	0.75*	3.56	0.12	0.11		1.01	0.246	0.052	0.080	7.91	7.70
8	42	12.25	2100	83.02*	3.32*	0.75*	3.56*	0.12*	0.11*		1.08	0.229*	0.050*	0.079*	7.68	
10	45	13.75	2090	83.14	3.33	0.76*	3.56*	0.12*	0.11*		0.98	0.216	0.046	0.072	7.66	7.63
11	54	15.75	2070	83.36*	3.32*	0.76*	3.56*	0.12*	0.12*		0.91	0.203	0.041*	0.068*	7.54	
12	59	18.00	2125	83.49	3.32	0.76*	3.56*	0.12*	0.12*		0.89	0.179	0.036	0.063	7.46	7.35
13	63	20.00	2110	83.63	3.31	0.76*	3.56*	0.12*	0.12*		0.87	0.156*	0.034	0.060	7.38	7.33
14	65	20.50	2085	83.76	3.29	0.76	3.56*	0.12*	0.12*		0.81	0.151	0.030	0.057	7.34	7.29
Value	s whi	ich wer		ed) ned by into oz per mir		ion.	3.61	0.12	0.12		0.77	**	**	**	**	• •

crease of copper content for air and oxygen injection,

The rates of removal of manganese, silicon, and phosphorus, for both air and oxygen, Fig. 3, were essentially linear in the early stages of refining. However, the rates of removal of these elements leveled off as these elements approached low concentrations. Very little iron was removed until about 16 oz of air were injected; subsequently, the element was removed at a constant rate. With oxygen, the iron was removed in the early stages of refining, resulting in a linear curve over most of the refining range. The rate of removal of zinc for both air and oxygen was also linear. The zinc values used in Fig. 3 are obtained from the zinc column listed under "by difference" in Tables 2 to 6. This column was tabulated by subtracting from 100 per cent the total concentration of the 11 remaining elements in the sample. This procedure necessitated the estimation, based on previous and following samples, of some elements which were not analyzed in every sample. The injection of either air or oxygen, resulted in a constant increase in copper content.

Removal of manganese with air was very rapid until about 35 per cent of the original concentration remained; thereafter, the rate of removal leveled off until the manganese was completely removed after about 32 oz of air were injected. The rate of removal of manganese with oxygen was slower than with air although the same weights of oxygen and air were necessary to achieve complete removal.

The rate of removal of silicon with air varied linearly and the element was completely removed after about 34 oz of air were injected. The rate of removal of silicon with oxygen was a little slower than with air and approximately four per cent of the original concentration remained after 36 oz of oxygen were injected.

M

Sig

OX

ou

lai

OX

ph

fin

up

for

mı

use

wh

sec

tive

oxi

gar

not

me

litt

eve

can

con

seco

7

dee

use

this

S. S

ana

and

Melt

Elect

Ch

Bear

F.

ment 0.15

MR

The rate of removal of phosphorus with oxygen was constant, having approximately 13 per cent of its original concentration present after 36 oz of oxygen were injected. The rate of removal of phosphorus with air was slower than with oxygen in the early stages of refining. However, after appearing to adjust itself, the element was removed at a relatively rapid rate, retaining only 3 per cent of its original concentration after 38 oz of air were injected.

The rate of removal of iron with the introduction of oxygen was essentially linear over the entire refining range. The rate of removal of iron was relatively slow until approximately 16 oz of air were injected; subsequently the rate increased rapidly and exceeded that of oxygen by a small amount.

The rate of removal of zinc with oxygen and air varied linearly, although the use of the latter gas resulted in a faster rate of removal. The increase in copper content was constant and found to be practically the same with either air or oxygen injection.

Nickel, lead, tin, antimony, and sulphur in this particular alloy were found to be unaffected by introduction of either air or oxygen, and therefore, no graphical representation was prepared. For aluminum, the results obtained from the wet and the spectrographic methods did not coincide. An approximate representation on the removal with air from the alloy is presented.

Discussion of Results

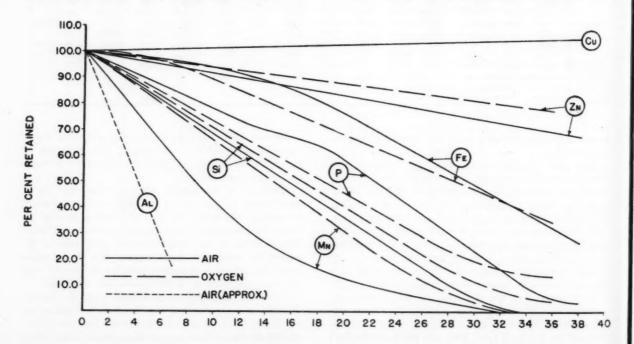
The inevitable loss of zinc by oxidation varies with temperature, atmospheric conditions, time of refining, and other factors. It is believed that the rate of removal of some impurities, such as iron, is dependent upon the concentration of zinc. Thus, commercial operations with air indicate that iron is more rapidly removed after the zinc concentration is reduced below 6 per cent. Observing Fig. 3, it was found that iron removal increased after the zinc concentration was reduced below 7 per cent. However, upon viewing the results of the experiments with oxygen, it is apparent that the iron removal was constant even when the zinc concentration was relatively high.

Figure 3, representing air and oxygen, indicates the sequence of removal of the following elements: aluminum, manganese, silicon, phosphorus, iron, and zinc, and the percentage increase of copper during refining. The sequence of removal of the above mentioned elements was found to be consistent when using either air or oxygen.

A small saving of zinc and consistent removal of iron resulted from the use of oxygen in place of air. There is good reason to believe that if the alloy contained a larger percentage of iron and zinc, the efficacy of using oxygen in removing iron would be exhibited more clearly.

Considering the fact that air contains about 23 per cent oxygen by weight, the results show that oxygen is considerably less effective than air. Although data from these experiments are insufficient to explain this phenomenon, several probable reasons can be offered: (1) the possible presence of other gases in the compressed air acting as a catalyst (water or nitrogen); (2) agitation of the bath while using air (inert gases such as nitrogen bubbling through the bath); (3) loss of oxygen in the injection process due to lack of suitable material for the lance assembly.

The second reason is believed to have considerable



WEIGHT OF GAS INJECTED, OUNCES

Fig. 3-Comparison of use of air with that of oxygen.

YS

al

ly

W

e.

as

19

p.

en

es

nd

e-

n-

en

of

ir.

n-

of.

X-

er

n

ta

in

in

0-

rt

1);

to

le

40

significance since the violent bubbling caused the oxygen in the compressed air to disseminate throughout the bath. The absence of agitation with pure oxygen thus may have prevented full contact and only local refining in the vicinity of the immersion lance may have taken place. This reasoning is supported by the curves appearing in Fig. 3: (a) with oxygen, rates of removal of manganese, silicon, and phosphorus were closely aligned with each other; (b) final removal of phosphorus and silicon was slowed up when using oxygen; (c) although manganese was found to be very active, its rate of removal was very much slower with the use of oxygen than with the use of air.

The third reason does not seem probable. Had oxygen injection been efficient, its superiority would have been revealed in the early stages of refining where the gas was less likely to escape.

Conclusions

1. The curves in Fig. 3 represent the refining of a secondary copper alloy using air and oxygen, respec-

2. The sequence of removal of the elements by oxidation was found as follows: aluminum, manganese, silicon, phosphorus, iron and zinc. It may be noted, however, that although the rates of removal of each element may be different, all of these elements are being removed simultaneously.

3. The use of pure oxygen in place of air showed little value in the refining of typical brass alloy. However, there are indications that the use of oxygen can result in the efficient removal of iron where high concentrations of zinc are present.

Further investigation is required before conclusive evidence on the value of the use of oxygen in refining secondary copper alloys can be determined.

The authors take this opportunity to express their deep appreciation to R. Lavin & Sons, Inc., for the use of their metallurgical and research facilities in connection with the experimental work reported in

Thanks are also due to Mr. F. D. Keller and Dr. S. Selig for their help and guidance in the chemical analyses and to Messrs. W. B. George, R. Cochran and J. Wilkus for their helpful suggestions.

Bibliography

- 1. W. B. Amess, "Eliminating Carbon With Oxygen," Steel, vol. 122, Feb. 2, 1948, p. 120.
- 2. G. E. Bellew, "Oxygen Injection Applied In Steel Foundry
- Melting," AMERICAN FOUNDRYMAN, vol. 16, no. 5, p. 24.
 3. J. H. Berryman and J. M. Crockett, "Oxygen In The Electric Furnace," Iron Age, vol. 162, Aug. 5, 1948, p. 72.

DISCUSSION

- Chairman: W. B. Scott, American Brake Shoe Co., National Bearing Div., Meadville, Pa.
- Co-Chairman: H. G. Schwaß, Bunting Brass & Bronze Co., Toledo, Ohio.
- F. L. RIDDELL: 1 Why were not the air and oxygen injected into the furnace at constant rates of flow? In the oxygen experiments you had a flow of 0.3 and 0.5 oz per min as against about 0.15 oz per min of air.
 - MR. GLASSENBERG: Unfortunately, there were no means avail-

- able to maintain a constant flow of gas for each experiment. As mentioned under "Injection Technique" in this paper, the rate of flow of air was adjusted to cause minimum loss through spattering. In the case of oxygen injection, flow was controlled so as to maintain a constant temperature in the bath. However, since the comparison between air and oxygen was done on a weight basis, time was of no significance.
- MR. RIDDELL: The point I am trying to bring out is that in the analysis of the material vs time, the higher the rate of injection, the faster the removal. You plotted your curve on the basis of weight injection, time not being mentioned.
- Mr. GLASSENBERG: This is not true. Rate of injection with air appears to have relatively no bearing on the rate of removal of impurities when data from the three experiments are plotted on a weight basis. The only reason why experiment A-1 (0.182 oz per min) indicates that the rate of removal of the elements exceeded the rate of removal of the elements in experiment A-3 (0.139 oz per min), say after 90 min., is that A-1 had more ounces of air injected than A-3 at the same time interval. It was this extra amount of air injected in the given time interval that caused additional refining in A-1.
- Observing the two experiments with oxygen, the rates of flow were 0.506 and 0.315 oz per min, respectively. Nevertheless, when the data from each experiment are plotted on a weight basis, the results are identical, clearly showing that rate of injection was of no significance.
- Mr. RIDDELL: It may be advantageous from the cost basis to use oxygen.
- MR. GLASSENBERG: It is beyond the scope of this experiment to present cost data. It seems likely that a more efficient and economical operation may be obtained by the use of a mixture of other gases which, when compared on a weight basis, will remove impurities more rapidly than shown by air or oxygen in Fig. 3
- R. W. Parsons: 2 The 0.119 per cent phosphorus in this analysis seems to be quite high. I cannot conceive of any scrap which would give this amount. Is this correct?
- Mr. HESSE: This particular analysis was prepared especially for these experiments and I do not remember the exact quantity involved. I believe it was around 5000 or 6000 lb. Impurities or undesirable elements were put there on purpose so a better idea of what was happening in refining with either oxygen or air could be obtained.
- G. P. HALLIWELL: 8 I believe that the heat of formation of phosphorus is much higher than that of manganese. What is puzzling me is why is the manganese eliminated so much faster than the phosphorus? Both are there in the same amount, so the factor of mass action is the same.
- Mr. Glassenberg: There is actually no standard to base this on. Heat of formation in standard tables are given for simple reactions. The oxidation reactions that are taking place in the bath are probably very complicated. However, the experiments definitely show that manganese, whether using air or oxygen, was removed faster than phosphorus.
- S. Greenfield: 4 Mr. Riddell asked a question pertinent to using oxygen as a general cause of saving time in a plant. I think you will experience such a slag condition by using oxygen that the time you may save in your use of oxygen, you will lose in skimming off your furnace.
- G. H. CLAMER: 5 During the past year we have conducted some experiments on enriching the air in the reverberatory furnace with oxygen. We have gotten nowhere, because, as the authors stated in this paper, in their laboratory experiments they had a great deal of difficulty with the nozzle. We have been unable to find any material that will stand up under the conditions required to introduce the oxygen.
- This paper is of value because of its negative results. It would seem hardly worthwhile to conduct experiments further in considering the use of oxygen in a commercial way.
- The one important point, however, as the authors brought out, is the relative relationship of iron and zinc in the rate of elimination.

Foundry Metallurgist, H. Kramer & Co., Chicago.
 Technical Director, The Ohio Brass Co., Mansfield, Ohio.

^a Director of Research, H. Kramer & Co., Chicago.

Samuel Greenfield Co., Inc., Buffalo.

⁵ Ajax Division, H. Kramer & Co., Philadelphia.

SOME EFFECTS OF TEMPERATURE AND MELTING VARIABLES ON CHEMICAL COMPOSITION AND STRUCTURE OF GRAY IRONS*

By

E. A. Lange** and R. W. Heine***

ABSTRACT

This paper presents results of a study of some fundamental melting phenomena that direct the course of composition and structural changes in gray cast iron. Variables studied affecting chemical composition were air oxidation of the iron in various refractories, the effect of temperature on composition changes, influence of slags, iron oxide, rusty scrap, coke, and the results of melting under a CO atmosphere. Silicon oxidation was shown to be promoted at low temperatures and prevented at high temperatures by oxidizing atmospheres.

Carbon losses were shown to be practically nil at low temperature and to increase rapidly with temperature. Silica reduction was shown to be normal chemical reaction in molten cast iron at temperatures above 2671 F. The effects of iron oxide in causing oxidation were shown to be entirely different from atmospheric oxidation. Structural changes were shown to be related to oxidizing melting conditions with chilling tendencies being promoted but dependent on temperature. The principles of temperature dependent chemical reactions causing composition and structural changes are discussed.

Engineering control of Gray iron compositions and structures is always associated with the melting practice followed in the foundry. Gray cast irons may be structurally and chemically altered in several ways by temperature, fuels, refractories, slags, atmospheres, cooling rates, inoculants, the charge, furnace operation, and other melting variables. The endproduct is an integrated result of the interactions of the aforementioned variables. Some of these factors have been successfully investigated and reported in the literature while others remain to be studied. It is the intent of this paper to present the results of a study of certain fundamental melting phenomena which direct the course of composition and structural changes in gray cast iron.

Chemical Composition

Composition changes in melting gray cast iron are normally considered from a total viewpoint as losses or gains. Losses of silicon and manganese result from oxidation. Gains may occur from contact of the iron with the fuel in the case of carbon and sulphur. Such changes are viewed as the normal course of events and are rectified by adjustments in the charge or melting practice. While knowledge of overall changes in composition is satisfactory for some practical purposes, it does not clarify the nature and direction of the chemical processes involved nor does it provide explanation when the anticipated changes do not occur. Accordingly, the authors determined first to study the nature and direction of composition changes in molten cast irons to determine the principles involved under a variety of melting conditions.

Air Oxidation

Since this was a laboratory investigation induction melting was employed. The molten iron was contained in a silica crucible because of the silicious nature of many refractories and slags. However, effects of other refractory crucibles were studied and these are taken up later. Eight-pound heats of iron were melted down in the induction furnace and raised to a pre-selected temperature and held at this temperature in contact with air to permit oxidation of the iron to progress. The metal used for these tests was taken from scrap castings made of cupola gray iron in a commercial foundry. The object of these tests was to study the effects of temperature on change in carbon and silicon percentage in the iron under the oxidizing conditions attained by induction melting. The results of oxidation of the iron when held at various temperatures in contact with air are shown in

From Fig. 1 the importance of temperature in effecting the direction of composition changes with respect to carbon and silicon may be seen. Oxidation of both of these elements is seen to be greatly influenced by the temperature at which the oxidation occurs. At low temperatures (2375 F) Fig. 1 reveals that carbon is not oxidized from the metal whereas silicon is. The protection of carbon may be related to two factors. First of all, the ability of silicon to react with oxygen decreases as temperature increases. On the other hand,

tur sub lov ger

tec

per

ture

of c

tion

the

cast

belo

mov

pear

tem

slag

thes

vidi

this

term

mag

ture

Thi

action

0

Ab tracted from a thesis submitted by E. A. Lange in partial fulfillment of requirements for a Master of Science degree at the University of Wisconsin.

^{**} Graduate student, Department of Mining and Metallurgy, University of Wisconsin.

^{•••} Assistant Professor of Metallurgical Engineering, Department of Mining and Metallurgy, University of Wisconsin.

S

m

on

ch

or

es

Ir-

of de

dy

in n-

n n-

its

se

re

to ane

as

n

its

in

ne

g.

at

in

ef.

e

of

d

1

n

ie

S

n

d,

47

carbon reacts more strongly with oxygen as temperature increases (see Ref. 1 for the theoretical principles substantiating these statements). Thus silicon, at the lower temperature, combines more strongly with oxygen than does carbon and therefore may assert a protective action on the carbon.

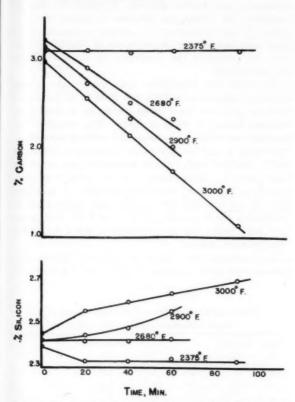


Fig. 1—Curves showing changes of carbon and silicon percentages in molten cast iron held at the temperatures indicated in a silica crucible in contact with air.

Another factor which may prevent the oxidation of carbon at the low temperatures exists in the formation of a scum or dry slag which covers the surface of the metal. This scum is a characteristic of normal cast iron compositions when they exist at temperatures below about 2550 F. The scum cannot be removed by mechanical means. If attempts are made to remove it, it re-forms immediately. The scum thus appears to be a reaction product which forms when the molten cast iron reacts with oxygen in air at temperatures below 2550 F and may be considered a slag phase which normally is generated by the iron at these lower temperatures. Reference 1 is cited as providing the reasons and experimental evidence for this statement. The nature of the scum has been determined. It consists of a mixture of a-crystobalite, magnetite, and a small amount of fayalite, the mixture containing about 50 to 60 per cent iron oxide.1 This scum may act as a barrier in preventing the reaction between oxygen in air and the carbon in the iron.

On heating, the scum disappears at about 2550 F

and the surface of the melt becomes clear. It is postulated that carbon reduces the oxides which make up the scum and thus causes them to disappear from the surface of the melt. When the scum has been removed by heating the iron it becomes possible for the atmosphere to oxidize the molten iron and hence oxidation of the carbon proceeds readily. The curves for 2680, 2900, 3000 F shown in Fig. 1, indicate the course of oxidation of the carbon at these temperatures. Since the carbon appears to oxidize at a steady rate, straight lines are obtained on a graph. Concurrent with the disappearance of the scum from the surface of the metal, the changes in silicon percentage of the melt alter in their direction. As temperature increases it is seen that silicon is no longer oxidized from the iron. There is, in fact, a silicon pickup due to reduction of silica in the silica crucible by the carbon in the iron.

1.
$$SiO_2$$
 (solid) $+ 2 C = Si + 2 CO$ (gas).*

The above reaction results in the evolution of CO bubbles and pitting of the crucibles, especially at elevated temperatures.

As in the case of other chemical reactions, this

When underlined, the symbol signifies that the element is dissolved in molten iron.

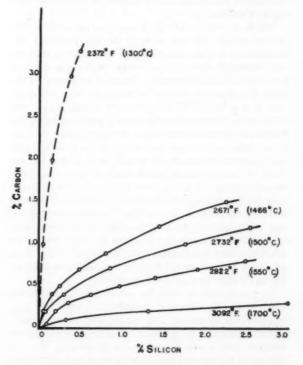


Fig. 2-Graph showing calculated equilibrium concentrations of percentage carbon and silicon for the reaction $SiO_2(s) + 2C = Si + 2CO(g)$ in molten ironcarbon-silicon alloys contained in a silica crucible under one atmosphere pressure of CO. Solid curves indicate temperatures where silica reduction will occur spontaneously if an excess of carbon is present.

process is temperature and concentration dependent. The temperature at which this reaction proceeds spontaneously to the right was calculated as being 2671 F.1 The latter temperature is evidently in agreement with the course of reactions as shown in Fig. 1. The effect of concentration is evidently quite great since, from Fig. 1, it may be seen that carbon can be oxidized down to 1 per cent with continued increase in silicon pickup at 3000 F; i.e. Reaction 1 continues to go to the right even with less than 1 per cent C in the melt.

The equilibrium relationships between carbon and silicon where Equation 1 would progress neither to the right nor to the left have been calculated for various temperatures.1 These equilibrium compositions are shown plotted as curves in Fig. 2.

Figure 2 presents graphically the importance of oxidation-reduction reactions in controlling the direction of chemical composition changes in molten cast irons. The reader is referred to Ref. 1 for a detailed analysis of these principles. For the present purposes, it may be pointed out that Fig. 2 indicates the influence of concentration and temperature on composition changes within the limitations of the assumptions made in calculations. An excess of silicon over an equilibrium concentration would permit oxidation of that silicon to occur at the temperature involved. An excess of carbon over the equilibrium concentration would permit the oxidation of carbon, the reduction of silica, and result in silicon pickup according to Reaction 1.

While the curves of Fig. 2 apply to theoretical conditions which do not exactly fit the practical circumstances of cast iron melting, nevertheless they reveal the basic chemical equilibria involved, and therefore the direction taken by chemical processes in the iron. For example, it can be seen that once a normal cast iron is heated to temperatures above 2671 F the silica reduction reaction becomes a predominating influence in directing the course of composition changes. If no other factors were operating in the normal melting temperature ranges, it would be expected on the basis of this laboratory work that all gray cast irons would increase in silicon and decrease in carbon content during the upper temperature ranges of normal melting operations. From Fig. 2 it appears that all cast irons contain more than sufficient concentration of carbon to insure the progress of silica reduction when the iron is contained in silica refractories. This has been proven under foundry operating conditions for malleable cast iron compositions.2

In conclusion it may be stated that normal cast irons are of a chemical composition which is nonequilibrium in nature and which will change chemical composition in a direction governed mainly by temperature; lower temperatures promoting the oxidation of silicon and the protection of carbon; higher temperatures promoting the oxidation of carbon and silicon pickup.

Influence of Furnace Atmosphere

In considering the composition changes of carbon and silicon which have been discussed in the preceding paragraphs, the effects of the melting atmosphere

should be more fully examined. The composition changes obtained in Fig. 1 were the result of action between the molten metal, the air over the metal surface, and the refractory. At temperatures where the oxidation of carbon progresses, the composition of the air over the metal becomes enriched in carbon monoxide and carbon dioxide. Changes in this atmosphere by convection permits oxidation to progress. If an atmosphere of carbon monoxide were held over the surface of the metal it should be possible to eliminate the oxidation of carbon by the air atmosphere.

To study this possibility, tests were made at two different temperatures, 2375 and 2800 F. Eight-pound heats of gray iron were again melted in an induction furnace having a silica crucible. The crucible was covered and sealed and provided with means of introducing a carbon monoxide atmosphere over the melt. Purified tank carbon monoxide was used for this purpose. Temperature of the metal was measured by Pt-Pt 10% Rh thermocouple. A sufficient flow of gas through the furnace was maintained to produce a slight positive pressure of CO. The metal was held under the gas atmosphere for 1 hr at the temperatures stated.

Gas samples were taken at intervals during this period and their analysis is given in Table 1. In

TABLE 1-COMPOSITION CHANGES UNDER CO ATMOSPHERE IN SILICA CRUCIBLE

Temp.	Analysis			Time, Min	1	
F	%	0	15	30	45	60
2375	C	3.24	3.24	3.24		3.21
	Si	2.26	2.26	2.22	2.25	2.25
	CO ₂	8.88	* *	6.03		6.00
	CO	91.12		93.97		94.00
2800	C	3.22	3.18	3.15	3.11	3.10
	Si	2.23	2.25	2.30	2.32	2.33
	CO ₂	2.78		2.27		2.00
	CO	97.22		97.73		98.00

Table 1 it is seen that the composition of the atmosphere over the melt gradually settled to a fairly steady value. During the 1-hr period samples were withdrawn from the metal at 15-min intervals. Composition changes which occurred are presented in Table 1 and Fig. 3. From Fig. 3 it can be seen that the results obtained at 2375 F under a CO atmosphere are little different from those obtained in air. However, at 2800 F a marked difference is noted.

The oxidation of carbon is markedly inhibited in the presence of a CO atmosphere. There is, however, a loss in carbon and a gain in silicon as might be expected at this temperature from Equation 1. This observation points out that carbon losses at high temperatures originate from two sources. The minor loss occurs by the reaction of carbon with silica in the furnace lining. Such carbon loss is accompanied by a chemically equivalent increase in silicon percentage in the melt. This fact is demonstrated in Table 1 and Fig. 3 by a 0.12 per cent decrease in carpe me

die

cia stra bo tha ma cor

Th clea oth rea

tur

whi vate OCC inc wit atm

I out mo al 1 ONS

lon

ion

etal

ere

ion

on

at-

oro-

ere

2005

the

wo

ind

ion

was

in-

the for

iea-

ent

to

etal

the

his

In

60

3.21

6.00

4.00

3.10

2.33

2.00 8.00

at-

ere m-

in

hat

106-

air.

in

er,

be his

igh

in

ied

er-

in

ar-

bon and a 0.10 per cent increase in silicon in 1 hr at 2800 F.

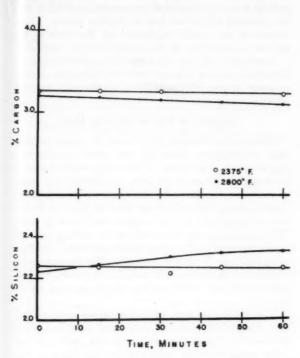


Fig. 3—Curves showing changes of carbon and silicon percentages in molten cast iron held under a carbon monoxide-rich atmosphere at the temperatures indicated.

While this composition change is not exactly chemcially equivalent, it is sufficiently accurate to demonstrate the point in question. The latter type of carbon loss may be described as indirect oxidation, i.e., that lost by reaction between carbon and silica. The major oxidation loss of carbon which may occur in contact with more oxidizing atmospheres than CO, for example as in the data of Fig. 1, occurs by direct oxidation of carbon by the reaction

2.
$$C + O = CO(g)$$
.

This reaction occurs when the surface of the metal is clear and free of the scum described previously. In other words, the direct oxidation of carbon progresses readily in an oxidizing atmosphere above the temperatures where the dry slag disappears from the surface of the metal.

The furnace atmosphere thus influences the rate at which carbon oxidizes from the molten iron at elevated temperatures. Minimum oxidation of carbon occurs under a CO atmosphere while oxidation loss increases with more oxidizing atmospheres and also with increasing rate of movement of the oxidizing atmosphere over the surface of the metal.¹

In connection with the dry slag, it should be pointed out that this slag persisted even under the carbon monoxide atmosphere so that it appears to be a normal reaction product between oxygen and molten cast

iron at low temperatures. Thus, at low temperatures the metal is virtually in equilibrium with a slag phase. At higher temperatures, above 2550 F, the slag phase disappears, and the metal co-exists with a gas phase, oxidation progressing in a manner limited by the composition of the gas. However, the presence of a slag phase generated by the iron at any temperature is a function of the composition of that iron. It has been shown that a slag may be present or absent at any particular temperature depending on the silicon content of the iron.1 When an excess of silicon is present over that which is in apparent equilibrium for Reaction 1 (see Fig. 2), the excess is oxidizable and results in the formation of the persistent dry slag referred to above. Composition changes concomitant with the presence of the dry slag, i.e., excess silicon, are oxidation of silicon and protection of carbon in the iron.

Effect of Coke

Coke is commonly regarded as a source of carbon and sulphur pickup in cast irons. However, in the light of the predominating influence of silica reduction, Equation 1, it appears that coke may have a further effect in promoting silicon pickup at high temperatures. An experiment was performed to test this possibility. An 8-lb heat of gray cast iron was melted in a silica crucible and held under a 1-in. layer of coke at temperatures of 2700 F and 2950 F, for a period of 1 hr. Composition changes under

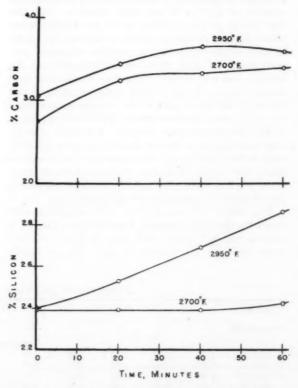


Fig. 4—Curves showing changes of carbon and silicon percentages in molten cast iron held under a layer of coke at the temperature indicated.

aı

in

% SILICON

Fi

car

he

5)

wh

be

cer

up

in

in

dec

ove

iro an

be over at of turn

trai iroi the tho

these circumstances are presented graphically in Fig. 4. From Fig. 4 it can be seen that silica reduction is promoted by the presence of solid carbon in the form of coke, even at the low temperature of 2700 F.

It can be seen by comparing Fig. 1 and 4 that the rate of silicon pickup in the presence of coke is greatly accelerated over that which occurs in its absence. Furthermore, because of the solubility of carbon in cast iron the carbon content of the metal does not decrease but rather increases to a certain level of concentration during the period of silicon pickup. Thus, it appears that conditions similar to those occurring in the well of a cupola, where coke and iron are in direct contact, favor processes of silica reduction. This is to be expected in view of the chemical principles involved.

Effects of Refractory Type

It was considered possible that the type of refractory employed for containing molten iron might change the direction of oxidation or reduction reactions which cause changes in the chemical composition of molten iron. Accordingly, a series of tests were run in other types of crucibles. The magnesia crucible was utilized for one series wherein heats were made at 2375, 2680, 2900 and 3150 F. The composition changes with respect to carbon and silicon which occurred, using the magnesia crucible, are tabulated in Table 2.

It can be seen from Table 2 that the same general trend of composition changes persist in the magnesia crucible as well as in the silica crucible. The fact that silicon pickup occurred at the two higher temperatures in the magnesia crucible is difficult to explain in view of the fact that the crucible refractory contains almost no silica. However, it is thought that this pickup was due to the reduction of silicious slags which develop during melting down of the iron and adhere to the upper part of the crucible walls in contact with iron. Carbon changes proceeded exactly as in the case of the silica crucible. The same observation with respect to the dry scum forming at temperatures below 2550 F and disappearing at temperatures above that apply. It is thus concluded that the magnesia crucible does not alter the fundamental oxidation or reduction reactions which are involved.

The effects of clay-graphite type crucibles were tested by the same method. Composition changes of molten iron in contact with this refractory are also given in Table 2. Again it is seen that the funda-

mental reactions are the same as in the case of the silica crucible with slight modification. Definite silicon pickup over a 1-hr period was noted at 2680 F and the amount of carbon loss is slightly lower. These variations are readily explained on the basis of the normal composition of clay-graphite type crucibles.

Apparently the type of refractory crucible used for induction melting does not alter the direction of composition trends although their magnitude is influenced.

Influence of Type of Slag over Melt

Composition changes which may be caused by slags were investigated. Since the iron oxide content of a slag may act as a source of oxidation with respect to carbon and silicon in the iron, an attempt was made to evaluate the effect of FeO content is two different cupola slags. One slag was selected because it was known to be high in FeO content and black in color. This particular slag was produced by adding iron ore to the regular cupola charge and it contained 12.62 per cent FeO and 37.64 per cent SiO₂, the balance being considered calcium oxide, aluminum oxide, magnesium oxide and alkali oxides usually present in the slags.

The other slag was considered to be a normal cupola slag, dark green in color and containing 5.38 per cent FeO and 45.54 per cent SiO₂, with the balance considered the oxides mentioned before. The latter slag was obtained from a commercial foundry. Gray cast iron scrap was melted in a silica crucible and covered with a half-inch layer of the slag material. A series of heats was made at different temperatures for each slag. The change in composition of the metal and slag over a 1-hr period was determined. Some slag composition changes are shown in Table 3. while the change in chemical analysis of the metal

TABLE 3-SLAG ANALYSIS

Slag	Initial Composition	After 1 hr contact with metal at 2375 F	After 1 hr contact at 2900 F
1. Iron oxide rich	1		
cupola slag (b	lack)		
%FeO	12.62		6.05
%SiO2	37.64		49.00
Normal cupola (green)	slag		
%FeO	5.38	13.94	4.53
%SiO2	45.54	44.6	58.94

TABLE 2-AIR OXIDATION DURING INDUCTION MELTING

Time, Min.	0)	1	5	9	30	4	15	6	0	9	90
Composition	C	Si	C	Si	C	Si	C	Si	C	Si	C	Si
					Magn	esia Crucil	ole					
2375 F	3.40	2.42	3.37	2.39	3.35	2.35	3.35	2.35	3.35	2.30		
2680 F	3.18	2.40	2.93	2.37	2.67	2.34	2.42	2.40	2.16	2.38		
2900 F	2.96	2.34	2.71	2.38	2.41	2.40	2.12	2.45	1.835	2.45		
3150 F	2.98	2.32	2.68	2.36	2.36	2.38	2.08	2.39	1.76	2 44		
					Clav-Gr	aphite Cru	cible					
2375 F	3.26	2.36			3.24	2.29			3.25	2.33	3.27	2.28
2680 F	3.26	2.42			2.78	2.44			2.33	2.48		

INS

the

on

nd

ese

he

for

m-

lu-

igs

a

to

de

nt

as

or.

re

er

ng

eshe

u-

38

al.

he

le

al.

es

he

d.

3.

al

hr

F

is illustrated in Fig. 5 and 6, and Table 6. In Fig. 5 and 6 it is seen that there is a pronounced difference in the effects produced by the two slags.

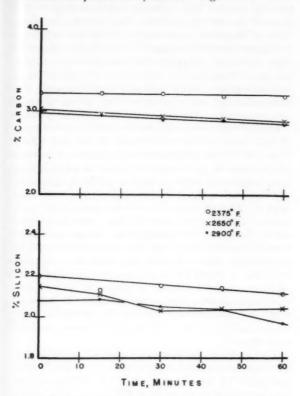


Fig. 5—Curves showing composition changes of the carbon and silicon percentages in molten cast iron held at the temperatures indicated under an oxiderich black cupola slag in a silica crucible.

The one with the higher iron oxide content (Fig. 5) caused oxidation of silicon even at temperatures where silica reduction from the crucible walls would be expected to cause an increase in the silicon percentage in the melt. In other words, the silicon picked up by silica reduction is offset by the loss of silicon in the metal due to its reaction with the iron oxide in the slag. Table 3 shows that the oxide rich slag decreases from 12.62 per cent to 6.05 per cent FeO over an hour's period at 2900 F. This decrease in FeO in the slag is partly brought about by reaction with the silicon in the iron and also with carbon in the iron. The essential point involved appears to be that an oxidizing slag may cause oxidation of silicon at temperatures where the silicon pickup would normally be anticipated. It was noted that the loss in silicon over a 1-hr period was approximately 0.10 per cent at all three temperatures tested. The effectiveness of FeO in causing silicon oxidation at all temperatures will be referred to again.

The oxidizing slag should be considered as an extraneous slag, that is, one not normal to the molten iron in the sense that it would not be generated by the iron itself. This statement is made with the thought that at high temperatures the molten iron

normally co-exists with an atmosphere phase rather than with a slag phase. The slag itself is extraneously introduced from some source such as coke ash, low temperature oxidation products of the metal, or from the refractories.

The influence of a low iron oxide content slag on metal composition is quite different, as revealed in Fig. 6. Strangely, the iron oxide content of this slag increases at low temperature but decreases at high temperature. It can be seen from Fig. 6 that the low iron oxide content slag does not result in oxidation of silicon at the higher temperatures. In fact, a slight amount of silicon pickup is noted over the 1 hr period. It appears that a slag which is low in iron oxide content may inhibit the reduction of silica by carbon in the molten iron but does not completely prevent it. Thus, the fundamental direction of chemical composition changes at elevated temperature when molten iron is in contact with lower iron oxide slags is still in the direction of the silica reduction reaction.

The exact iron oxide content of the slag which will cause oxidation of the silicon at elevated temperatures was not determined in this investigation. It may be a function of several of the other ingredients of the slag and that in itself would require extensive investigation. However, for the purpose of this work it is clear that Reaction 1 is a fundamental factor in causing composition changes in molten cast iron. The tendency of this reaction to proceed to the right is always operative at temperatures above 2671 F. But the mitigating effects of the slag may over-shadow

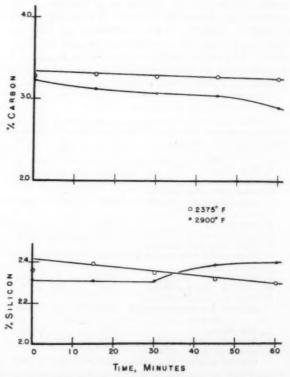


Fig. 6-Same as Fig. 5 but utilizing a normal green cupola slag rather than the oxide-rich slag.

the fundamental equilibria involved by causing some limited oxidation of silicon at the higher temperatures. With respect to carbon, by a comparison of Fig. 1, 5, and 6 it can be seen that the slags delayed the loss of carbon, probably through a simple barrier effect in preventing the reaction between the molten iron and the atmosphere above it.

A question is posed by the results of composition changes caused by slags containing iron oxides. The question may be simply stated as follows: Why does an iron oxide-containing slag cause oxidation of silicon at temperatures where the reduction of silica and consequent silicon pickup would normally be expected? The answer is obtained by considering that a new substance has been introduced into the system of reacting chemicals. As has been pointed out previously, a slag phase is not normally generated by atmospheric oxidation of the iron at high temperature due to the protective or reducing power of the carbon in the iron. Under these conditions, the reaction of silica reduction, 1. SiO_2 (s) + 2 C = Si + 2 CO (g), regulates the progress of composition changes.1 Only the excess carbon above the equilibrium concentrations for Reaction 1 is oxidized, Fig. However, in the presence of iron oxide a new reaction may occur as follows:

3a. Si+4 FeO (1)=2 FeO'SiO₂ (liquid slag)+2Fe (1) 3b.
$$\overline{\Delta}$$
F° (1600°C.) = -66,680 + 18.15T

The free energy changes which accompany Reaction 3a are given by Equation 3b. Equation 3b was calculated from data presented by Richardson.3 Free energy values may be viewed as a measure of the tendency for a reaction to occur. Comparing values at 1600 C (2912 F), \triangle F = -7860 cal gm atm¹ for Reaction 1, and \triangle F = -32680 cal gm atm for Reaction 3. The higher negative \triangle F value for Equation 3 is interpreted as meaning that this reaction is likely to occur to a greater extent than 1, providing a source

of FeO is available. The slag provides the FeO and . thus causes loss of silicon from the metal even at temperatures where such loss would not occur were the slag absent. The experimental data (Fig. 5 and 6) show that the effect of Reaction 3 overbalances the effect of Reaction 1 and results in a net loss of silicon providing the FeO content of the slag is sufficiently high. However, when the FeO content of the slag is too low, Reaction 3 is not able to progress to a marked extent. Silica reduction again becomes the predominating influence in causing silicon percentage composition changes and results in silicon pickup.

Further tests were made to determine the behavior of iron oxide as the active factor in the slag which causes silicon loss at elevated temperatures. Fourteenpound heats of gray iron were made, to which a constant magnitude addition of iron oxide as mill scale was made at temperatures of 2500, 2700, 2900 and 3000 F. The procedure by which these heats were made is described in a later section. The object here was to determine the composition changes produced by oxidation resulting from iron oxide additions. The composition before and after the addition is reported in Table 4. From Table 4 it is seen that iron oxide additions caused oxidation of silicon at all temperatures investigated. This effect is distinctly different from that produced by atmospheric oxidation even when accelerated rates of air oxidation are utilized as in Ref. 1.

The constancy of silicon loss reported in Table 4 at all temperatures indicates that Reaction 3 is relatively insensitive to temperature. The former observation was also noted as an effect of FeO in the slags tested and is a conclusion which agrees with the theoretical information available from Richardson.3 It is evident from Table 4 that the elements phosphorus and sulphur are little affected by the oxide additions; the latter a circumstance to be anticipated. From Table 4 it is also seen that the percentage of carbon

TABLE 4-EFFECT OF IRON OXIDE ADDITIONS ON COMPOSITION**

Temperature and Material	% C	$\triangle C_{\bullet}$	% Si	△Si*	% Mn	△ Mn*	% P	△ P*	% S	△ S*
2500 F										
1. Base Metal	3.20		2.52		0.48	**	0.79		0.072	* *
2. After iron oxide addition	3.19	01	2.44	08	0.47	01	0.79	0	0.072	0
3. After iron oxide addition										
followed by 0.20% Al addition 2700 F	3.19	01	2.38	14	0.44	04	0.80	+.01	0.075	+.003
1. Base Metal	3.23		2.58		0.47	*. *	0.81		0.072	
2. After iron oxide addition	3.21	02	2.45	13	0.45	02	0.80	01	0.077	+.005
3. After iron oxide addition										
followed by 0.20% A1 addition 2900 F	3.22	01	2.44	14	0.42	05	0.82	+.01	0.078	+.000
1. Base Metal	3.31		2.60	* *	0.53		0.79		0.072	
2. After iron oxide addition	3.29	02	2.46	14	0.51	02	0.81	+.02	0.075	+.003
3. After iron oxide addition										
followed by 0.20% A1 addition 3000 F	3.28	03	2.46	14	0.47	06	0.82	+.03	0.076	+.004
1. Base Metal	3.18		2.51	**	0.52		0.80		0.078	* *
2. After iron oxide addition	3.14	04	2.37	14	0.50	02	0.79	01	0.078	0
3. After iron oxide addition										
followed by 0.20% A1 addition	3.12	06	2.37	14	0.51	01	0.79	01	0.079	+.001

Values refer to total change in analysis from base metal to end of each step.

clea pro due it s iron of c ence net dire at a side ing poir post dow at 2 to 2

Si

E. /

lost

eral T

T Mate Melti

The

stoc

tion

* Silio seen ing h loss o 6.90 from

did r F in Fre ing s mate the e effect gener An causin

with by L ore a creasi Mons ore to tions

*Ins Wiscon

^{••} Iron oxide additions consisted of mill scale. The amount added to the metal in each case was equivalent to 0.20% oxygen for the oxide figured as Fe₃04.

NS

nd

at

re

nd

1e

n

ly

is

1e

IE-

p.

or

h

17-

n-

le

d

re

e

d

e

d

e

4.

ıt

n

d

ıt

r

d

ıl

d

lost by oxidation although small increases with temperature, a point made earlier in this paper.

The data discussed in the preceding paragraphs clearly emphasize the differences in oxidation losses produced by atmospheric oxidation and by oxidation due to iron oxide either alone or in a slag. However, it should be recognized that silicon losses caused by iron oxide-bearing slags do not mean that the ability of carbon to reduce silica is nullified by their presence. Such silicon losses are merely a reflection of net composition losses resulting from two opposing directions of change.

Since iron oxide promotes the oxidation of silicon at all temperatures, rust on the metal might be considered as a cause of unusual analyses variations during melting. A few tests were made to study this point. Some melting stock was rusted by outdoor exposure for four months. The rusty scrap was melted down, the slag skimmed, and then samples were taken at 2375 F, 2680 F, and 2900 F as the metal was raised to 2900 F, and finally after holding 60 min at 2900 F. The same procedure was followed but with melting stock that had been cleaned by blasting. Compositions obtained are reported in Table 5. It can be

TABLE 5-EFFECT OF RUST ON ANALYSIS OF MELT*

Material	Clean	Scrap	Rusty Scrap		
	%C	%Si	%C	%Si	
Melting Stock	3.40	2.60	3.35	2.56	
2375 F	3.28	2.42	3.20	2.38	
2680 - 0 min	3.27	2.42	3.16	2.36	
2900 - 0 min	3.21	2.47	2.94	2.30	
2900 - 60 min	A.	in 60 min 3% Si	Pickup i		
• Silica Crucible					

seen that the rusty material caused greater loss during heating to 2900 F than the clean material; a total loss of 10.1 per cent for the former as compared with 6.90 per cent for the latter. Further, losses continued from 2375 F up to 2900 F with the rusty scrap but did not with the clean material. Total pickup at 2900 F in each case was 0.13 per cent over a 60-min period.

From these data it appears that rust on the melting stock causes oxidation not characteristic of clean material and produces effects which carry over through the entire melting temperature range. In addition the effects of rust follow the principles established as general for iron oxide.

Another example of the influence of iron oxide in causing composition changes was noted in connection with a cupola research project. Some work was done by Leon Monson,* investigating the effects of iron ore additions to the cupola charge as a means of increasing the yield of iron obtained from the cupola. Monson⁸ studied additions up to 15 per cent of iron ore to a normal cupola charge. Increasing ore additions resulted in higher iron oxide content in the

From these data it is concluded that high iron oxide contents in cupola slags produce oxidation loss of silicon in the cupola as well as when those same slags are studied experimentally in the induction furnace.

Factors Influencing Composition Changes

Net composition changes occurring when cast irons are melted are seen to be the result of a number of interrelated factors. Silicon is most readily oxidized at low temperatures and functions as a protective agent for carbon. A dry slag is a normal low temperature phase existing in contact with molten iron and resulting from oxidation of silicon and iron. At temperatures above 2550 F, the reducing power of carbon inhibits the oxidation of silicon and, in fact, causes silica reduction at temperatures above 2670 F when the molten iron is in contact with the normal gas phase generated by air oxidation of the iron or in a CO atmosphere. The direction of composition changes at elevated temperature is governed by two fundamental reactions as follows:

2.
$$\underline{C} + \underline{O} = \underline{CO}$$
 (g)
1. $\underline{SiO}_2(\underline{s}) + 2\underline{C} = \underline{Si} + 2\underline{CO}$ (g)

Reaction 2 results in rapid atmospheric oxidation of carbon from the metal concurrently with an increase in silicon percentage which occurs as a consequence of Reaction 1. These reactions occur whenever cast irons are heated through normal melting temperature ranges in silicious refractories as a result of the fact that such irons are non-equilibrium compositions; the direction of change being that of the fundamental equilibria involved (Fig. 2).

The experimental work reports only composition changes for the oxidizable elements silicon and carbon. Manganese may be included as an element which behaves as does silicon; that is it is readily oxidized at low temperatures and protected from oxidation by carbon at high temperatures. This has been adequately proven in Reference 1 so that manganese has not been considered in this paper.

Slags may change the course of the reactions mentioned above. The iron oxide content of the slag is a major factor in determining whether silicon will be oxidized or silica reduced at temperatures where reduction would normally proceed. As a consequence of the two conflicting reactions, over all changes in composition may show either a silicon loss or silicon pickup and thus reveal the predominating reactions which have occured.

cupota slag. Slag from heats with no ore additions contained about 6.7 per cent FeO while the addition of 15 per cent of the charge as iron ore produced slags averaging 12 to 16 per cent FeO. Oxidation losses of silicon were tremendously increased by the higher FeO content in the slag. The initial composition averaged about 2 per cent silicon in the iron tapped from the cupola when no ore was utilized. The ore additions producing the higher iron oxide slag lowered the percentage of silicon to the range of 0.90 to 1.3 per cent for the same charge materials depending on the particular heats investigated.

^{*}Instructor, Dept. of Mining and Metallurgy, University of Wisconsin, Madison, Wisconsin.

Structural Effects Related to Melting

In the course of obtaining the data on chemical composition changes a number of observations were made relative to the structure of the cast iron produced. As a part of the procedure in studying composition changes, test bars were poured at points of interest. A standard wedge test for chilling tendency, a round test bar $3\frac{1}{2}$ in. long and varying from 1-in. diameter at one end to $5\frac{1}{8}$ in. at the other and occasional keel blocks of the type used for cast steel were poured in core sand molds. Considering all the melting experiment discussed in the section on chemical composition, certain general observations may be made.

Two factors were found to be of prime importance in altering the structure in the gray cast iron. These are, first, temperature; second, the effects of oxidation. The two factors were found to be interrelated.

To illustrate the importance of temperature and oxidizing melting conditions, the results of a number of tests will be cited. The first case considered is that of a series of heats in which iron oxide as mill scale. predominantly Fe₃O₄, was added to the melt. A quantity of mill scale equivalent to 0.2 per cent oxygen was added to a 14-lb heat at 2500, 2700, 2900 and 3000 F. From each heat a wedge test and a plate casting was poured prior to the addition of the mill scale and immediately after the addition of the mill scale. After pouring the second set of castings, an aluminum addition of 0.2 per cent aluminum was made in an attempt to remove the effects of oxidation produced by the mill scale. Subsequently a third set of castings was poured. Composition changes produced by the mill scale additions are revealed in Table 4. From Table 4 it is seen that composition changes of the same order were produced by the mill scale addition at each of the four temperatures. However, the effect of making the additions at various temperatures was quite pronounced in terms of structural changes.

Considering the chill tests at 2500 F no change was observed in the fractured wedge tests. There was a complete absence of chill in each of the three samples taken at that temperature. At 2700 F a white chill was observed in the apex of the wedge bar from the metal which had received the mill scale addition. At

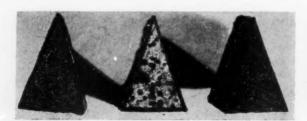


Fig. 7—Chill test of base metal poured at 2900 F (Left); (Middle), from 2900 F, but with addition of iron oxide; (Right), from 2900 F, with addition of iron oxide, and deoxidized with 0.20% Al. See Table 4 for composition.

2900 F the most marked difference became evident as revealed in Fig. 7. The fracture of the left test bar in Fig. 7 was that obtained from the base metal poured immediately after the temperature of 2900 F was reached. The middle test bar reveals the fracture obtained after the mill scale addition. The right hand test bar reveals the influence of the aluminum in removing the chilling tendency produced by the mill scale addition. These changes in chilling tendency could be produced at a temperature of 2900 F, by any of several oxidizing factors in melting. In other words, any oxidizing factor, in this case mill scale, produced a marked increase in the chilling tendency of the iron. This chilling tendency could be removed by the addition of a suitable deoxidizing agent.

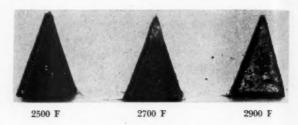


Fig. 8—Chill test after iron addition at 2500 F (Left); 2700 F (Middle); 2900 F (Right). Composition differences of these three test bars are seen to be insignificant as reported in Table 4.

The importance of temperature in obtaining this effect cannot be over-emphasized. Fig. 8 illustrates this point. Considering only the metal poured after the oxidizing addition of mill scale had been made, the chilling tendency was found to increase markedly with temperature. The first indications of increase in chill depth occur at about 2650 to 2700 F. Mottled structures such as those shown in the right hand fractured test bar may be produced at 2800 F and above by oxidizing melting conditions. In any case, the chill so produced may be removed by the addition of a deoxidizing element.

The chilling tendency evident from the wedge tests is quite sensitive to cooling rate. From the metal produced at 2900 F three test plate castings were



Fig. 9—Fractures of plate-type castings poured from 2900 F. Upper fracture obtained in base metal poured immediately after reaching 2900 F. Middle fracture resulted when iron oxide was added to base metal at 2900 F. Bottom fracture was obtained when 0.20% Al was added to metal of middle sample.

Fig.

me

sh

lo

In

ab

mi

die

tru

WO

of

flu

sin

wh

tio

sla

sla

are

10

me

DNS

ent

est

tal

F

ac-

tht

ım

he

id-

F.

In

ill

ng

ild

ng

t);

ni-

is

es

er

le,

ly

in

ed

C-

ve

ne

of

ts

al

re

poured. These castings were 2½ in. wide x 9 in. long x ¼ in. thick. Fractures of these test plates poured from 2900 F are shown in Fig. 9. The upper bar is metal prior to the oxidizing addition, the middle bar shows the effect of the oxidizing addition, and the lower bar reveals the effect of the aluminum addition. In other words, the chilling tendency revealed in Fig. 7 carried into small plate type castings, most noticeably at the edges where the cooling rate is higher.

Some of the structural changes referred to above might be ascribed to changes in composition caused by the mill scale additions. A study of Table 4 indicates that this is not the case. Furthermore, it is true that none of the compositions listed in Table 4 would be expected to solidify with a chill structure of the type shown in Fig. 8, right, unless the influence of oxidizing melting conditions is considered.

Effects of Melting under Slags

Since a slag usually contains iron oxide, effects similar to those described above may be anticipated when molten iron is held under a slag. The composition of metal held for 1 hr under two different cupola slags at different temperatures is given in Table 6; slag analysis being given in Table 3. Fractures of wedge tests and round bars poured from these irons are shown in Fig. 10 and 11, items 1 and 2. In Fig. 10 test bars 1 and 2 show the effect of holding the metal in contact with the two different slags at temp-

TABLE 6

	Carbon and Si 1 Hr under Ir		
	2375	2650	2900
C	3.20	2.90	2.86
C Si	2.11	2.04	1.97
	Carbon and Si 1 Hr under N		
	2375		2900
С	3.26		2.89
Si	2.30		2.40

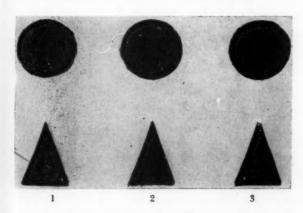


Fig. 10—Samples poured from 2375 F. Sample 12-1 poured from under green, low iron oxide content slag; 12-2 poured from under black, high iron oxide slag, and 12-3 poured from under CO atmosphere.

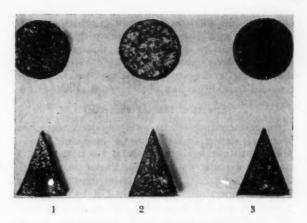


Fig. 11—Sample 13-1 poured from under green slag at 2900 F; Sample 13-2 poured from under black slag at 2900 F; Sample 13-3 poured from under CO atmosphere at 2800F. Chemical analyses are reported in Tables 1 and 6.

erature of 2375 F. Figure 10, test bar 1, applies to metal poured from under the normal cupola slag of low iron oxide content. Figure 10, test bar 2, applies to metal poured from under the cupola slag high in iron oxide. In both cases the metal is gray and shows no evidence of chilling tendency in the wedge bar. Figure 11, test bar 1, shows the chilling tendency of metal held under the normal cupola slag at 2900 F, while test bar 2 shows the chilling tendency of metal poured from under the iron oxide-rich slag at the same temperature. In both cases it can be seen that marked chilling tendency prevails at the higher temperature. In other words, the effect of the slag in promoting chilling is temperature dependent.

It is interesting to note that the metal of Fig. 11, test bar 1, is about 0.4 per cent higher in silicon content than that of test bar 2, while the carbon contents are about the same. In Fig. 11 it appears that higher silicon content diminishes the chilling tendency mildly in the slower cooled, round test bar but not in the wedge bar. The dependence of these chilling effects on temperature is indicated in Fig. 12. All three sets of test bars were poured from metal melted under the iron oxide-rich cupola slag. Test bar 2 poured from a temperature of 2650 F, shows only a slight chilling tendency in the test bar, whereas test bar 3, poured from 2900 F shows marked chilling tendency even though the compositions are essentially the same.

The work with the oxidizing slag again points out the fact that the influence of temperature and oxidizing melting environments are interrelated. At higher temperatures the oxidizing conditions promote chilling tendencies. It is interesting to note that even what is considered a normal green cupola slag is capable of producing chilling tendencies over extended periods of time at elevated temperatures. However, green cupola slags may vary considerably in FeO content so it should not be interpreted that all green slags will produce the aforementioned effect. It should again be pointed out that some investigators

E.

ba

sp

wh

by

ro

WE

sh

me

th

SO

tir

th

in

en

wl

lev

co

H

fre

so

wl

th

might consider the structural changes related to composition changes. However, the work with slags shows three compositions which are very similar in nature but which differ greatly in structure. Furthermore, the chilling tendencies produced may be completely removed by addition of a small amount of deoxidizer.

Effects of Melting under CO

Some interesting results were obtained when the iron was melted under a carbon monoxide atmosphere. As stated earlier, the metal was held under a carbon monoxide atmosphere for 1 hr at two temperatures, 2375 and 2800 F. Fractures of test bars poured at each temperature are shown in Fig. 10-3 and 11-3. Figure 10, test bar 3, revealed that no chilling tendency is developed at the lower temperature. Figure 11-3 shows that pronounced mottling is developed in the wedge test at the higher temperature, while a slight tendency for inverse chill occurs in the round bar. The composition of these two test bars is very similar, the lower temperature one containing 5.21 per cent carbon, 2.25 per cent silicon, while the higher temperature bars contain 3.10 per cent and 2.33 per cent silicon. The structural difference, then, is clearly one produced by some effect of temperature since this was the only significant variable between the two sets of test bars. Furthermore, it may be noted that the chilling tendency promoted under the CO atmosphere is not accompanied by any oxidation of silicon.

At first it might be considered strange that melting under a carbon monoxide atmosphere produced the same effects as obtained by treating the molten metal with an oxidizing agent. Carbon monoxide is considered as a reducing atmosphere in the sense that it is supposed to lower the oxygen content of the metal. However, this appears to be a fallacious point of view. The mechanism whereby oxygen escapes the metal at temperatures above 2671 F is through the evolution of carbon monoxide.1 The metal becomes saturated with CO at the elevated temperatures. Saturation is promoted by CO bubbles passing through the metal by Reaction 1. The latter effect may be visually noted. The evolution of such carbon monoxide depends on the difference in atmospheric pressure of the ca bon monoxide in the atmosphere over the melt and the vapor pressure of the carbon monoxide in the molten metal.

When the pressure of carbon monoxide over the melt is one atmosphere, as in this experiment, the difference in pressure is almost one atmosphere less than when the metal is exposed to air. Clearly, then, the ability of carbon monoxide, or oxygen, to leave the metal is lowered by melting the iron under a carbon monoxide atmosphere. In this sense, melting under a carbon monoxide atmosphere is actually oxidizing in that it may raise the oxygen content of the metal. This point has been amply demonstrated by Chipman and Marshall. They have proven that the oxygen content of iron-carbon alloys increases as the pressure of carbon monoxide over the melt increases. It thus appears that exposing the metal to a carbon monoxide atmosphere produces effects which are the

same as those produced by other oxidizing treatments at equivalent temperatures.

Effects of Steam Treatment of Melt

A series of tests was run in which live steam was bubbled through molten iron at 2600, 2700, 2800 and 2900 F. Steam was employed for the reason that it may act as a source of oxygen and may thus be oxidizing in nature by reacting with molten iron. As in the case of all the previous tests, temperature was found to be extremely important in determining the effects produced by this treatment. It was found that the temperature of 2600 F was inadequate for producing any change in the structure of cast wedge test bars even after bubbling the steam through the metal for 6 min at the rate of 2.6 gm of water per minute. However, at 2700 F the amount of chill equivalent to that shown in Fig. 10 was obtained after 5 min of treatment. Further treatment at 2700 F produced no additional effects.

However, at 2800 and 2900 F it was possible to produce mottled and completely white wedge tests in 2 to 4 min of steam treatment. Composition changes in this period are similar to those produced in the previous tests so that they are not here repeated. Test plates poured from the steam treated iron produce variations similar to those shown in Fig. 9. The question may be asked whether these effects are due to the oxygen obtained from the steam or whether they are due to hydrogen. It should be pointed out that changes in structure of the kind produced by steam are not produced by bubbling dry hydrogen through the melt. Furthermore, the chilling tendency produced with steam can be removed by the addition of a deoxidizing element such as aluminum. Chilling tendencies promoted by steam are thus concluded to be due to the oxidizing effects produced by dissociated water on the metal.

Effects of Cooling Rate

The discussion presented in the preceding sections dealt mainly with the results obtained in wedge test

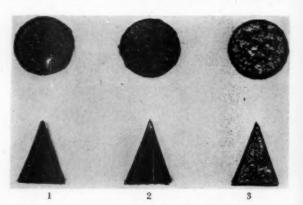


Fig. 12—Fractures illustrating influence of temperature on effects produced by oxide-rich slag. Sample 14-1 poured from 2375 F; 14-2 from 2650 F and 14-3 from 2900 F. No significant difference in composition exists between 14-2 and 14-3.

el

d

11

d

bars. An observation of major importance with respect to the increased chilling tendency obtained by oxidizing melting conditions consisted of noting that such metals are extremely sensitive to the rate at which the molten metal is cooled. This is illustrated by comparison of the wedge fractures and fractures of round test bars, as shown in Fig. 10, 11 and 12. A wedge fracture test showing considerable mottling is shown in these figures to present considerably less mottling in the round test bar, especially whenever the carbon content is above about 2.8 per cent. In some cases only a few specks of white iron are noted in the round bar, even though the wedge bar is entirely mottled. Especially in the case of bars melted under a carbon monoxide atmosphere it was noted that the wedge bar was mottled, the round bar presented only a few specks of white, and a keel block poured from this metal showed no mottling whatso-

From these observations it is concluded that the type of chilling tendency promoted by oxidizing melting conditions is transient in nature. The effect is entirely lost during cooling in heavy sections and also when the carbon content of the iron is at a higher level. Both factors, heavy sections and higher carbon content, produce the effect of slower rates of cooling. Higher carbon content means that more of the metal freezes as a eutectic and hence a greater portion of solidification occurs at lower temperature and is thus delayed.

The transient nature of the chilling tendencies produced by oxidizing melting media at elevated temperatures is related to the general chemical effects which are produced by heating and cooling cast irons through normal melting temperature ranges. Reversi-

bility of the chemical effects, reported early in the paper, above and below the temperature range of 2600 to 2700 F is the phenomenon likely responsible. Slow cooling would favor the return to conditions normal at the lower temperatures, i.e., a gray fracture in the wedge test. It was found experimentally that the chilling tendency introduced into these irons at high temperatures could be removed by cooling to a lower temperature. Pronounced chilling tendencies produced by iron oxide additions at 2900 F, Fig. 7, were removed by cooling the metal to 2375 F in the furnace holding at that temperature and then pouring wedge tests. Such tests approximate the effects which are obtained by continuous cooling in massive castings. Massari and Lindsay7 have reported the aforementioned decrease in chilling tendency produced by cooling irons to low temperatures below the maximum melting temperature prior to pouring the chill

However, their work was done with very low silicon, 0.55 per cent, irons and much less severe changes were obtained. Higher sensitivity of the iron to rapid cooling rate is to be expected in higher silicon gray irons if reversibility of the chemical reactions is the factor responsible for the variable response to chilling. The former statement is based on a consideration of Fig. 2. From Fig. 2 it appears that low silicon irons must be cooled to much lower temperatures before the reaction $SiO_2(s) + C = Si + CO(g)$ ceases to go to the right or even can reverse and go to the left. On this basis the low silicon iron tested in Ref. 7 would be expected to show less loss of chilling tendency on cooling from a higher melting temperature to a low temperature prior to pouring. In other words, the results reported in (7) and herein may be con-

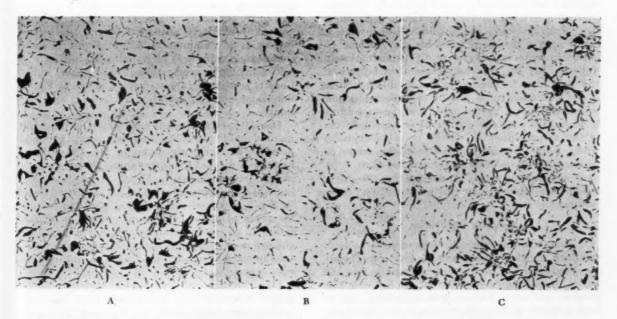


Fig. 13—Graphite distribution in plate casting poured (A) from 2500 F, no treatment; (B) 2500 F, after add-

g poured ing iron oxide; and (C) 2500 F after adding iron oxide feer add-followed by 0.20% Al. See Table 3 for chemical analyses. Mag. 100×.

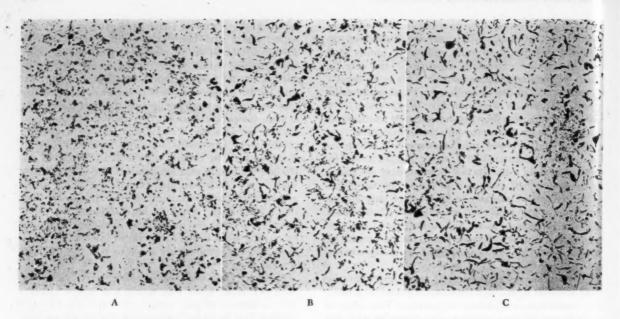


Fig. 14—Graphite distribution in plate casting poured from (A) 2700 F, no addition; (B) 2700 F, after adding

iron oxide; (C) 2700 F, after adding iron oxide followed by 0.20% Al. See Table 4 for chemical analyses.

sidered as complementing each other. This depends on considering the alterations in chilling tendency produced by temperature to be due to the reversible nature of chemical reactions which occur when molten cast irons are heated or cooled through normal melting temperature ranges.

Reversibility of chemical reactions on slow cooling molten iron introduces a new facet of the chemistry of gray irons. During cooling reversal of Reaction 1 suggests a self-deoxidizing power present in the iron. On cooling to temperatures where carbon is no longer capable of reducing oxides of silicon or manganese it would appear that these elements may become effective in removing oxygen or CO from the melt. Nonmetallics, usually silicates, may then be formed and these may be present in the solidified iron as shown in Reference 1. Cessation of carbon monoxide evolution from the molten metal, and the appearance of nonmetallics, may be considered as visual evidence of the aforementioned reversal, or self-deoxidizing power. Such reactions are, of course, dependent on time so that cooling rate or factors that affect it become of vital importance in determining the extent to which equilibrium is approached at the lower temp-

Graphite Distribution

A considerable number of observations were made regarding the microstructure of the gray irons investigated in this work. Factors such as graphite type and size, nature of the nonmetallic inclusions, phosphide pattern, and other features were examined. An extended report of these observations is considered outside the scope of this paper. However, brief comment will be offered regarding graphite distribution. Graphite pattern was found to be related to chilling

tendency. In general, cases of increased chilling tendency in the wedge test were accompanied by modified types of graphite in the round test bar and plate castings.

One illustration of the aforementioned effect will be given. For this purpose, the iron oxide treated heats of Table 4 will be utilized since they offer a narrow range of variation in chemical composition. Graphite distribution in the plate-type castings produced from heats at 2500, 2700, and 2900 F is shown in Fig. 13, 14, and 15. Figure 13 reveals that graphite pattern is relatively unaffected by the addition of iron oxide or aluminum at 2500 F. Figure 14 shows considerable refinement of graphite obtained as a result of the higher temperature, 2700 F. Some indications of modified graphite may be noted in Fig. 14 but no pronounced influence results from the additions.

by

ter

tic

ate

SO

hi

an

siz

WE

ma

CO

pr

ch

the

tal

bu

Figure 15 reveals marked modification of the graphite produced by the addition of iron oxide. The modified graphite structure shown in Fig. 15b occurred in the gray areas of the plate casting illustrated in the middle fracture shown in Fig. 7. A graphite distribution in the location of white areas such as shown in Fig. 7 appear as shown in Fig. 15c. Deoxidation by means of adding aluminum resulted in marked return to the more normal graphite, Fig. 15d. The changes in graphite distribution discussed above were usually noted when marked changes in chilling tendency occurred.

Like chilling tendency, graphite distributions obtained in castings were found to be very sensitive to cooling rate. For example, metal poured from under a CO atmosphere, Fig. 11-3, was noted to present a modified graphite distribution in the round test bar but a perfectly random distribution in a keel block.

ded te

n

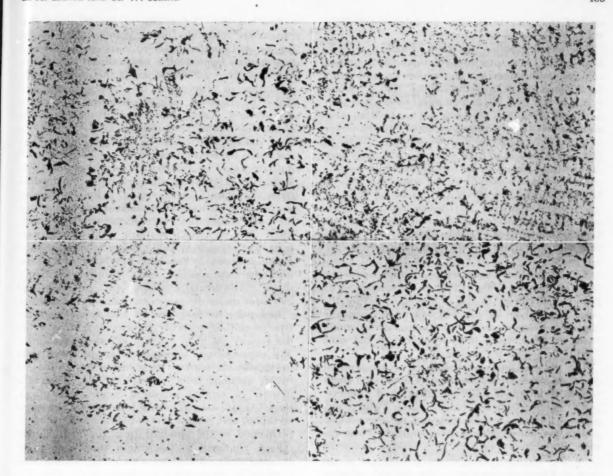


Fig. 15—Graphite distribution in plate casting poured from (A) 2900 F, no additions; (B) 2900 F, after adding iron oxide, gray area; (C) 2900 F, after adding iron

oxide, mottled area; (D) 2900 F, after adding iron oxide followed by 0.20% Al. See Table 4 for chemical analyses. Mag. 100×.

Such sensitiveness to cooling rate is further evidence of the fleeting nature of some of the effects produced by melting under oxidizing conditions and at elevated temperatures. Thus, it is concluded that modified graphite so produced is the end-product of a combination of the melting and cooling rate factors delineated. The combination is such as to be most trouble-some in thin castings which must be poured from higher temperatures and which naturally cool rapidly.

Observations of nonmetallic inclusions, sulphides and silicates, were made. Marked changes in type and size of inclusions as influenced by melting conditions were noted. Elevated temperatures favored the formation of large sulphides and some silicates under all conditions of melting studied. Deoxidation with aluminum was found to cause marked refinement of sulphides and ultimately duplex sulphides in a manner very similar to that reported for white cast iron providing initial temperature exceeded 2800 F. The changes in inclusions produced by aluminum followed the general trend of Type I, II, and III inclusions established as characteristic of deoxidation treatments but modified by the mechanism of solidification of

gray cast iron. It is not desired in this paper to report a study of nonmetallic inclusions in gray cast iron but merely to point to the similarity of principles operating. Therefore no further reference will be made to inclusions.

Summary

Interrelationships are a prominent feature of the data presented in this paper. Both structural characteristics as revealed by the wedge test and the direction of chemical composition changes were shown to vary greatly above and below the critical temperature range of 2600 to 2700 F.

Lower temperatures favor the reaction between silicon in the iron and oxygen while carbon is relatively unoxidized. The oxidizability of silicon at low temperatures results in the formation and persistence of a dry slag and favors the protection of carbon when oxidation is caused by exposure to gases. No bubbling of CO from the melt is visually evident. Under these conditions no unusual chilling tendency is produced in wedge tests or larger castings. Thus, at low temperatures composition changes are those normally anti-

E

ZO

ac

116

cu

lir

cipated from oxidation while variations in chilling tendency appear to be absent.

The reducing ability of carbon in molten iron becomes active at temperatures coincident with marked changes in the direction of composition and structural trends. At temperatures above 2700 F, carbon is readily lost from the molten metal by reaction with oxidizing gases. Yet, silicon is not oxidized at these temperatures by oxidizing gases because of the power of carbon reducing the reaction product, silica. In fact, silica reduction from silicious refractories produces silicon pickup in the metal analysis unless this effect or carbon is masked by slag reactions. Thus at higher temperatures effects the reverse of those at low temperatures exist and carbon loss and silicon pickup are normal trends of composition changes. Definite bubbling of CO from the melt due to reaction between carbon and the silica refractory is visually evident.

The reversal of effects discussed above over the temperature range 2600 to 2700 F is accompanied by pronounced change in the chilling tendency of the iron. Above about 2670 F, chilling tendency is greatly increased by exposure to oxidizing melting conditions; an effect which is not produced by the same

conditions at low temperatures.

The physical and chemical effects of temperature described above are concomitant. The physical observations of increased chilling tendency and evolution of CO bubbles from the molten iron at higher temperatures follow as a consequence of the chemical reactions prevailing at those temperatures. When surveyed for a major variable all reactions considered favor the retention of higher oxygen content in the iron at higher temperature. This variable whether considered as oxygen in solution, CO in solution or any other form is postulated as a dominating influence in producing the chilling tendencies described. The carbide stabilizing tendencies produced by oxygen in white cast irons used for malleable iron have been studied, albeit indirectly, and reported in Reference 5. The ability of deoxidizers to remove chill and the influence of oxidizing melting media to cause it support this viewpoint.

Furthermore, it has been pointed out in detail that the only effective deoxidizer in normal molten cast iron above 2670 F is carbon.1 The latter element functions less effectively as a deoxidizer of the metal as temperature increases due to increased solubility of CO. The silica reduction reaction above 2670 F favors increased solution since CO bubbles are generated and percolate through melt. In this respect, Chipman and Schneble⁶ have concluded that increased CO content in cast irons are responsible for the effects caused by superheating of gray cast irons. As pointed out previously superheating favors increased percentages of CO, and therefore oxygen, in solution. If oxygen in either form is accepted as a carbide stabilizing element, it is not surprising then that temperatures promoting chemical reactions which prevent the efficient deoxidizer silicon from functioning coincide with temperatures producing marked increases in chilling tendency.

The fact that the chilling tendencies under discussion are reversible further points to the role of the chemical reactions involved. Slow cooling through temperatures below 2550 F where silicon and manganese may act as self-deoxidizers of the iron results in mitigation or removal of the chilling effects promoted by oxidizing melting conditions at elevated temperatures. Rapid cooling prevents the attainment of the basic equilibria involved at the lower temperatures and favors retention of the higher gas content reached at the elevated temperatures. Therefore cooling rate would be expected to exert a profound influence in determining the end result obtained at room temperature. The essential point with respect to cooling as well as heating molten cast iron is that the usual temperatures encountered cover the range where an element such as silicon may either actively react with oxygen or not. This viewpoint is thoroughly presented in Reference 1. When such temperature ranges are spanned, interrelationship of composition and structural trends are thus congruent and inevitable under the proper circumstances of melting.

In arguing for oxygen in the iron as a cause of the type of chilling tendency reported herein, it is not the authors' intent to discredit all other possibilities. The theory offered is postulated as one which seeks to establish the common denominator for the parallel results obtained from investigating the effects of a number of melting variables; namely, refractories, slags, CO atmosphere, normal atmosphere, iron oxide additions, and steam treatment of the melt. The experimental results obtained from the melting procedures employed in this investigation seem to be fundamentally satisfied by considering the sum total of effects of temperature in relation to the principles of oxidation-reduction reactions prevailing in molten cast irons.1 Thus, structural changes in terms of chilling tendencies do not actually occur independently of fundamental chemical changes in the molten

metal.

Acknowledgments

The authors wish to express their appreciation to Prof. K. F. Wendt, Associate Director of the Engineering Experiment Station, The University of Wisconsin for his support and encouragement of this phase of investigation of cast iron metallurgy. Professor P. C. Rosenthal, Research Metallurgist, Lew F. Porter and Prof. G. J. Barker, Chairman Department of Mining and Metallurgy, have furthered this work through their helpful suggestions and criticisms and the authors extend their thanks.

Bibliography

 R. W. Heine—"Oxidation-Reduction Principles Controlling the Composition of Molten Cast Irons—Part I," A.F.S. Preprint No. 51-18.

2. R. W. Heine—"Oxidation-Reduction Principles Controlling the Composition of Molten Cast Irons—Part II, Air Furnace Melting," A.F.S. Preprint No. 51-18.

3. F. D. Richardson—"The Constitution and Thermodynamics of Liquid Slags," Disc. Faraday Society, No. 4, p. 244, 1948.
4. J. Chipman and S. Marshall—"The Carbon-Oxygen Equilibrium in Liquid Iron," Trans., A.S.M., vol. 30, p. 695 (1942).

NS

15-

of

gh

a.

in

O-

ed

nt

a-

nt

1.

n-

at

ct

at

ee.

ly

r-

id

g.

ie

ot

ks

el

a

S,

le

x.

0

æ

al

es

n

of

t-n

orn of C.d gh

11

5. R. W. Heine—"Some Effects of Deoxidation Treatments on Graphitization of White Cast Iron," Transactions, A.F.S., vol. 57, pp. 315-331, (1949).

vol. 57, pp. 315-331 (1949).
6. J. Chipman and A. W. Schneble—"Factors Involved in Superheating Gray Cast Iron and their Effects on Its Structure and Properties," Transactions, A.F.S., vol. 52, pp. 113-158 (1944).

7. S. C. Massari and R. W. Lindsay-"Some Factors Influencing the Graphitizing Behavior of Cast Iron," Transactions A.F.S. vol. 49, pp. 953-991 (1941).

ACTIONS, A.F.S., vol. 49, pp. 953-991 (1941).

8. Leon A. 'Monson-"Effects of Adding Iron Ore to the Cupola Charge," Unpublished master's degree thesis on file in the University of Wisconsin Library, Madison, Wis.

DISCUSSION

J. E. Rehder (Written Discussion): 1 This paper is a useful addition to the literature on the variables encountered when melting cast iron, and much work remains to be done in this field. In general the experimental data agree with the experience of the present writer, but there are points of interpretation of results on which there is not entire agreement.

This writer must take exception to the statement that "normal cast irons are of a chemical composition which is non-equilibrium in nature." This statement is unproven as it stands, and is meaningful only when the surroundings, such as container, are considered and specified. This may seem a minor point, but is important in its implications.

The data showing that the addition of excess free carbon as coke or graphite increases the rate of silica reduction from refractories, is of interest, and is of importance in cupola practice. In view of the high temperatures attained in the melting zone and well of a cupola, the siliceous slags obtaining with the acid refractories used, and the excess of hot coke, the writer has been and is of the opinion that a considerable amount of reduction of SiO2 takes place in the cupola well, and that normal melting losses of silicon from charge to tuyere level are much higher than the net loss figures obtained from analysis of iron at the spout would indicate. This is believed to be the reason why overall silicon losses are less than those of manganese in cupola melting, there being no source of replacement of manganese. In the basic cupola the more basic slag and basic well lining decrease opportunity for silicon pickup, and apparently high silicon melting losses result. Manganese melting loss is not changed as much; and the higher carbon contents in basic cupola melting are at least partly accounted for by less carbon in the iron being consumed by the SiO2 reduction reaction.

In fact, if the MnO content of a cupola slag is increased, re-

duction of MnO can take place. In the early 1940's the writer experimented with lump manganese ore as a cupola charge component for introducing manganese into cast iron. The manganese content of the iron increased, and a cost analysis showed that the overall cost of increasing the manganese content of the iron in this way was less than by using spiegel. However, this was during wartime and sufficient quantities of lump manganese ore were not available to use it as a normal practice.

were not available to use it as a normal practice.

The work on the effects of slags of different iron oxide contents is interesting, and should be extended in scope. It is believed that better correlations will appear if the whole slag composition is taken into consideration, and the "free" FeO calculated, since the presence of lime and alumina markedly affect compound formation in slags and at a given FeO content a cupola slag may or may not contain available FeO depending on the lime-silica ratio and the alumina content.

The work on additions of iron oxide and of iron oxide and aluminum to cast iron parallel unpublished work done by the writer, and in general is in agreement. The writer finds that the addition of as little as 0.02 per cent aluminum will produce much increased mottling and graphitization after an addition of iron oxide. In fact any reducing agent, such as ferrosilicon, zirconium, graphite, aluminum or titanium will produce a similar effect, with very small additions, and it would be expected that a reducing gas such as carbon monoxide would have a similar effect.

The effect of higher temperature as a "whitener" of cast iron, producing less primary graphitization, is evident, and it is well known that simply superheating an iron can change it from a gray to a white fracture. The iron can retain this carbide stabilized condition on decreasing the temperature to a certain extent, heating to 2900 F and pouring at 2800 F giving less mottle than heating to 2800 and pouring at 2800. This effect must be considered separately from others due to additions or to slag covers. The present writer considers that the effects observed in melting under a CO atmosphere are due to melt temperature effects, and not to the atmosphere alone, and that it has not been demonstrated that superheating favors increased percentages of CO in solution in the iron. Mention is made of the conclusions of Chipman and Schneble that increased CO contents in cast iron are responsible for the effects of superheating, but if their data is re-assessed in view of the major changes in chemistry involved, the conclusion is not justified, as shown by the present writer in discussion of their paper at the time of presentation.

¹ Foundry Engineer, Canadian Bureau of Mines & Technical Surveys, Ottawa, Ont., Canada.

TEMPERATURE, I DIV. = 100 F (55.6C)

FREEZING OF WHITE CAST IRON IN GREEN SAND MOLDS

By

H. A. Schwartz and W. K. Bock *

ABSTRACT

The time-temperature relations at the geometric center of white cast iron before, during and after freezing in green sand molds were observed in the course of experiments on another subject.

The castings were rather small and very simple.

The observations cannot be reconciled with Chvorinov's observation that the freezing time of steel castings is proportional to the square of their volume to area relation and approximately independent of other conditions.

Introduction

Chyorinov has proposed ("Control of the Solidification of Castings by Calculation," Foundry Trade Journal, vol. 61, no. 1199, pages 95 to 98, August 10, 1939) a relation between the volume and surface area of a steel casting and its freezing time in green sand molds. His diagram reproduced by Schwartz in Fig. 34 of "Solidification of Metals," A.F.S. Transactions, vol. 53, pp. 1–35 (1945), indicates that the freezing time was proportional to the square of the ratio of the volume of the casting to its area. Variations in pouring temperature and sand properties were not introduced as variables in this diagram, which included data from many sources and it is specifically implied that the shape of a casting of given V/A (volume to area ratio) is unimportant.

The V/A ratio, not its square, was used by Caine in a successful elucidation of the problem of feeder size. ("A Theoretical Approach to the Problem of Dimensioning Risers," Transactions, A.F.S., vol. 56, p. 492, 1948). Although this author was motivated in his approach by Chvorinov's paper, he used V/A not (V/A)² and in any event the correctness of Chvorinov's assumption is not logically a necessary consequence of Caine's success. This fact has been frequently overlooked.

Chvorinov's principle attracted much attention because of its simplicity although this very simplicity appeared to those familiar with heat transfer principles as worse than doubtful.

Paschkis' ("Comparative Solidification Studies," TRANSACTIONS, A.F.S., vol. 57, p. 100, 1949), working under the auspices of the A.F.S. Heat Transfer Com-

mittee, demonstrated that the $(V/A)^2$ relation was acceptable for geometrically similar figures at equal pouring temperature if the sand body be so thick as not to heat sensibly at the outer surface but that spheres, long cylinders, and slabs of great area having equal V/A ratios freeze in very different times. There is thus a "shape factor."

Procedure

Within the past few months there has been occasion, in the Research Department of National Malleable and Steel Castings Company, to take time-temperature curves during the freezing of white cast iron bodies cast in green sand from white cast iron having a carbon content in the general region 2.50 per cent and a silicon content near 1.10 per cent. The conditions of this experiment were imposed by the investigation of which it formed a part and were not chosen as ideal for the purpose of this paper. Nevertheless the observations can be used for an interesting comparison with Chvorinov's views and should, therefore, be of some general interest. Also, pyrometer and stop watch readings may be a more familiar approach to the problem of when a casting has finished freezing than the results of experiments on the Heat Flow Analyzer even though that device seems to be an entirely reliable and useful tool.

The bodies to be cast were rectangular blocks: 2 in. x 3 in. x 6 in., 1½ in. x 3 in. x 6 in., 1 in. x 3 in. x 6 in., 3¼ in. x 3 in. x 4 in. and ½ in. x 3 in. x 4 in. in size. As already stated the dimensions were chosen for reasons beyond the scope of this paper. They were poured from numerous heats varying within commercial limits of control and in the green molding sand habitually used. It is the custom to maintain the commercial sand here used near the following properties:—

Moisture,
$$\frac{6}{0}$$
 - 34 - 37
Permeability - 90 - 110

The purpose of the major investigation required a sampling during randomly chosen operating conditions as to metal composition and temperature and as to sand characteristics. This should not invalidate a comparison with Chvorinov for, since he gathered data from many sources, he implicitly accepted unknown variations of this character.

Director of Research and Research Engr., respectively, National Malleable & Steel Casting Co., Cleveland.

C-

lo

ıg

a

e

11

h

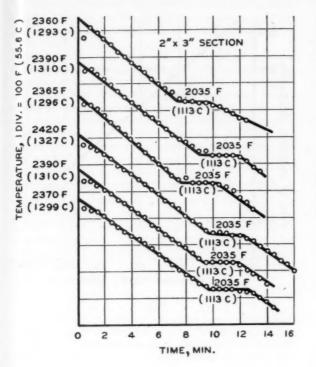
e

S

ıl

a

į.



(1277 C) I"x 3" SECTIONS (55.6 2320 F (1271 C) 2300 F (1260 C) 2330 F (1277 C) 00 2030 F (IIIOC) <u>ک</u> 2335 F TEMPERATURE, I (1280 C) (IIIO C) 2035 F (III3C) 4 8 10 12 14 0 2 TIME, MIN.

Fig. 1

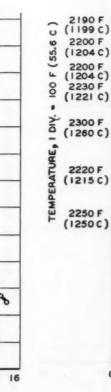


Fig. 3

1945 F

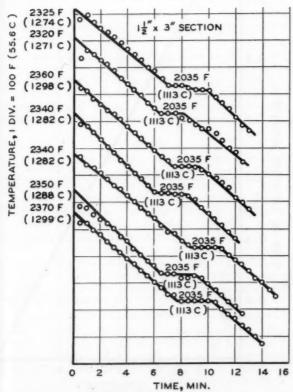
€(1063 C)

3" x 3" SECTION

1945 F

(1063C)

(1063 C)



0 2 4 6 8 10 12 14 16 TIME, MIN.

Fig. 2

Fig. 4

0

al

rl la ir tl

te

P co P fi co P n

ro tl ro st e: tu

n

d

th n

6

ti n th ei V st p

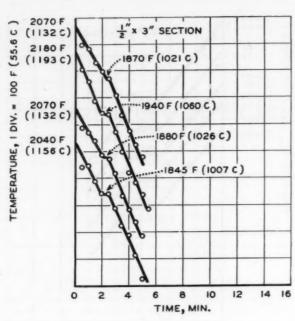


Fig. 5

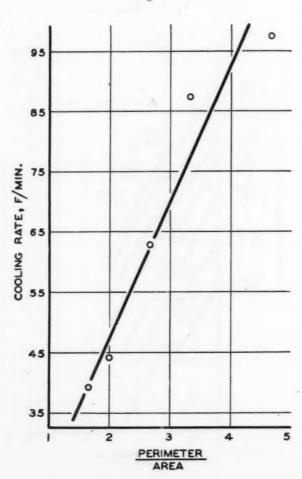


Fig. 7

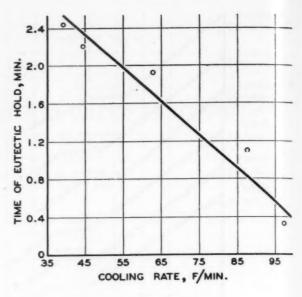


Fig. 6

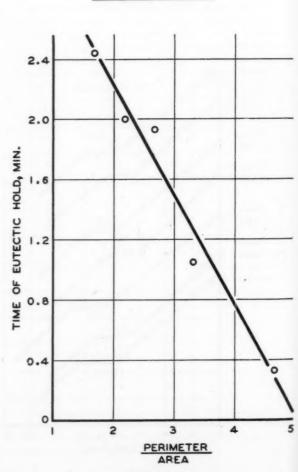


Fig. 8

There is, however, one distinct difference between our conditions and his in that we are concerned with an eutectiferrous alloy and he with a non-eutectiferrous one.

Our thermocouples were of the platinum—platinum rhodium type threaded through two-hole ceramic insulators and encased in silica tubes 5/16 in. O.D. x 3/16 in. I.D. The thermocouple was mounted, passing through the cope in such a position that the head was as nearly as we could determine at the geometric center of the body and was in contact with the closed end of the quartz sheath. The cold junction was maintained at the freezing point by mounting it in ice and water contained in a vacuum bottle.

Temperature readings were by a Leeds and Northrup portable potentiometer. Although we were not concerned with the precise measurement of the freezing point, we point to the close agreement of our eutectic freezing points in the larger bodies, free from supercooling, with the values of the iron-carbon-silicon phase diagram as evidence of the substantial correctness of our temperature measuring set-up.

Data

Figures 1 to 5 inclusive summarize our temperature readings plotted against time after the mold was filled. These readings were evaluated by computing, by least squares, the loci of the cooling curves above and below the eutectic point separately. The eutectic point is represented by a horizontal line through the nearly constant temperature readings. We neglected a few of the ealiest points while the couple was coming to temperature, and also an occasional reading near the beginning or end of eutectic freezing. This might represent some rounding off of a sharp inflection.

We can detect no thermal evidence of the liquidus and use the cooling curve extrapolated to zero time as our best knowledge of pouring temperature. The correctness of this for the smaller sections is doubtful. We note a marked depression of the eutectic temperature by supercooling in the two smaller sections.

Discussion

For our major problem, concerning graphitization during freezing, we were principally concerned with the time occupied in eutectic freezing. This point is not that of major interest in the present paper. Figure 6 shows the relation of cooling rate to the solidus on the time at the solidus. The relation is crudely linear although it may well be that as the cooling rate increases the time at constant temperature decreases more rapidly than in direct proportion. We shall later come to a time to complete freezing, which is more significant here. The points are the average values for each section size.

Figures 7 and 8 show the relation of cooling rate and time of eutectic freezing to the relation P/A (minimum girth of specimen to cross-sectional area). Were the specimens of infinite length, i.e., were there no end effects, this would be the reciprocal of the Chvorinov V/A. In either case, but especially in the latter, a straight line seems fairly satisfactory. At this point the purpose of these measurements, in the application to our major problem, was complete.

There are, however, some further evaluations to the general problem of heat transfer, and especially to Chvorinov's rule which form the subject of the present paper.

The V/A relationship with which we are concerned is tabulated below for the several bodies:—

TABLE 1-GEOMETRIC CONSTANTS OF FREEZING BODIES

Dimensions, in.	2x3x6	11/2x3x6	1x3x6	3/4×3×4	1/2x3x4
Volume, cu in.	36	27	18	9	6
Surface Area, sq in.	72	63	58	341/2	31
V/A in.	0.50	0.43	0.33	0.26	0.20
V/A-corrected					
for gate, in.	0.51	0.43	0.33	0.27	0.20

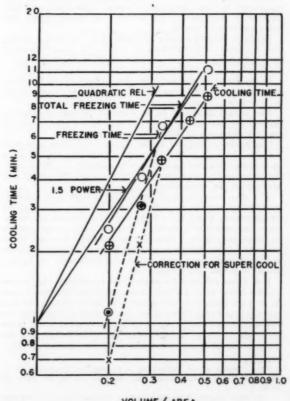
The correction for gate area through which no heat is lost is to be neglected as insignificant in relation to the precision of our knowledge as to the temperature data.

The corresponding cooling data averaged from Fig. 1 to 5 inclusive are as shown in Table 2.

TABLE 2

Dimensions, in.	2x3x6	11/2×3×6	1x3x6	3/4×3×4	1/2×3×4
Section thickness, in.	2	11/2	1	3/4	1/2
Cooling to Eutectic Temp	. 8.9	7.1	4.8	2.1	$\frac{1/2}{0.7}$
Cooling to Solidus	8.9	7.1	4.8	3.1	2.1
To complete solidification	11.4	9.5	6.7	4.1	2.5
Corrected to solidification				3.1	1.1

On the assumption that if there had been no supercooling the arrest at the solidus would not have been altered in length.



VOLUME / AREA

Conclusion

The data of Tables 1 and 2 are plotted, on log-log paper in Fig. 9. The relation between time to cool to the solidus (not the eutectic temperature but the temperature to which supercooling depressed the solidus) can be approximately represented by a line corresponding to the time being proportional to roughly the 1.5 power of V/A. A similar relation exists for the actual freezing time. Neither the relation between time for complete solidification, corrected for time spent in supercooling, nor the freezing time corrected for supercooling, can be represented satisfactorily by a straight line

Acknowledgments

We are indebted to National Malleable and Steel Castings Company, Cleveland Works personnel for making and pouring the molds; to Ira Young, Reidar Eriksen and James Anthony. Robert Everett of Sharon Works assisted in making observations and Mr. Anthony evaluated the observations as already described.

DISCUSSION

Chairman: H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: E. C. Troy, Foundry Engineer, Palmyra, N. J. V. PASCHKIS (Written Discussion): The authors add another valuable proof to the rising evidence, that Chvorinov's square law is incorrect. The authors also refer to the successful method of Caine with regard to risering: Caine works with the shape factor, S = V/A, whereas Chvorinov wants to apply S^a . However, the authors present paper shows, that at least for eutectiferrous alloys also the first power correlation, S, does not hold. There would then seem to be a contradiction between Caine's successful application of S and the present findings. This discrepancy can, perhaps, be explained by the difference in metals: Caine uses non-eutectiferrous metals.

However, there is another possibility for explanation. Reference is made to a discussion by the writer to a paper "Solidification of Steel Against Sand and Chill Walls" by H. F. Bishop, F. A. Brandt and W. S. Pellini, which appears on pages 435-450.

In the discussion the writer suggests to break down the shape factor S into a critical dimension D, and a proportionality factor p, $S \equiv D/p$.

The critical dimension is the diameter of long cylinders or

The critical dimension is the diameter of long cylinders or spheres, the thickness of large slabs with no end effects; and the smallest dimension of slabs with end effects.

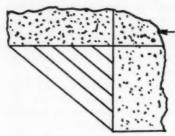
The bodies investigated by the authors have values of p from 4 to 2.5. This is a fairly wide range. Conceivably the difference in p in case of Caine's work was much smaller, i.e. the p value for casting and riser were more nearly the same. This n.ight well account for the discrepancy.

The great convenience of being able to work with a simple relationship such as the unfortunately incorrect one suggested by Chvorinov, would make it an attractive research job, to explore the possibility of finding a series of shape factors, possibly in relationship to the p value and to superheat. In other words, it would be desirable to explore the limits of validity of Chvorinov's obviously only very rough approximation to various shapes and sizes and temperatures.

J. B. CAINE (Written Discussion): 2 It would seem that we are rapidly losing sight of a basic axiom, that heat dissipation must be in some relation to the surface area of the casting, for how else can heat be extracted? Then too, we seem to forget that the effective surface area in regards to heat dissipation is not necessarily the geometric surface area. The steel foundryman has learned through experience that a small core surrounded by a

relatively large mass of metal does not extract any appreciable amount of heat, for he must riser the casting as if it were solid. He has learned from bitter experience that re-entrant angles cool slower than flat surfaces. In other words the effective surface area of these cores and re-entrant angles is less than unity.

Conversely, a corner is known qualitatively to freeze faster than a flat surface exposed to an infinite sand wall. Such a corner is in contact with more sand per square inch of surface area than a flat surface as shown in the accompanying sketch.



Where does the heat come from to satisfy this volume of sand?

po

T

If a flat surface in contact with an infinite sand wall has an effective surface area of 1, would not some of the surface area at a corner have an effective surface area of more than 1? It would seem to the discusser that this explanation if valid would be more logical than complicated shape factors.

Some quantitative information as to the relative heat extraction rate of such geometric departures from a flat surface would be of great help in quantitative risering, even though it be not theoretically correct.

W. S. PELLINI AND H. F. BISHOP (Written Discussion): ^a Chvorinov's proposals (reference cited) have aroused great interest and considerable contention. In view of the opinions and data presented by the authors it is well to reduce the original contentions of Chvorinov's to fundamentals for judgment as to their validity and practical value. Before proceeding with this I wish to go on record in agreeing without reservations with Dr. Schwartz that Caine's use of volume to surface area concepts in riser calculation does not stand or fall on Chvorinov's principles. Caine's relationships represent empirical relationships documented in themselves by the very experimental results used in the derivation of the curves.

It should be understood first of all that the mathematical relationships developed by Chvorinov apply as he specifies to "semi infinite mass of metal" or as he stated otherwise for plates to "infinitely large plates." Specifically, therefore Chvorinov sets his conditions as sections devoid of end and riser effects. More simply, this means that the length of section must be such as to eliminate end effects. If such conditions are met the following relationships may be derived starting with the basic expression of the rate of progression of freezing in a semi-infinite mass of metal:

$$X \equiv M/t$$
 $t \equiv time after casting$
 $X \equiv thickness of solidified metal$

 $M \equiv constant$ By assuming conditions such that $X \equiv S/2$ (or half thickness of a plate) , M may be considered a constant subject to calculation or empirical evaluation. The formula then becomes:

$$T = (S/2)^{s} / M$$

Since S/2 also expresses the ratio of V/SA of a semi-infinite plate the formula may be rewritten to

$$T = (V/SA)^2 / M^2$$

Thus, M (equal to 2.09) which is the solidification constant, is derived assuming conditions of semi-infinite plates. In this respect Dr. Schwartz is certainly correct in questioning the general applicability of Chvorinov's rules. In substantiation to his claims Chvorinov presents a logarithmic plot of solidification times vs V/SA ratios. He is careful, however, to state that these are mostly for simple geometrical forms. Variations of 10 to possibly 15 per cent from theory are indicated by his plotted points.

^{*}See "Control of the Solidification of Castings by Calculation" by N. Chvorinov, Foundry Trade Journal, Aug. 10, 1930, vol. 61, No. 1199, pp. 95-98.

¹ Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York.

³ Foundry Consultant, Wyoming, Ohio

⁸ Naval Research Laboratory, Washington, D. C.

LDS

able

olid.

cool

face

aster

h a

face

ome

vol-

an

area

It

ould

rac

bluc

not

vor-

rest

lata

con-

heir

vish

Dr.

s in

oles.

ocu-

in

ela-

emi

to

sets

lore

s to

ring

n of

of

s of ion

ate

int. reral ims US are bly

h.

The problem of intricate castings entailing pronounced end effects is treated separately; complex formulas, which are difficult to interpret, are proposed. While claims are made as to the applicability of these last concepts no data are presented as public proof thereof.

It is not clear how Chvorinov justifies use of an expression derived for plates of semi-infinite size to bars and other sections. The expression S/2 = V/SA for plates does not for example hold for bars for which S/4 = V/SA. Dr. Schwartz thus has valid grounds for questioning the lack of consideration for a "shape factor." As a by-product of other investigations 1,2 the writers have accumulated data as to the solidification times of bar and plate steel castings of semi-infinite lengths. Figure 10 shows a graphical summary of the final solidification times plotted according to Chvorinov's quadratic (V/SA)2 relationship. It is apparent from this plot that while a shape factor does exist it is only in the order of 10 to 15 per cent between plate and bar shapes. It is extremely interesting that the 2.09 factor for plates calculated by Chvorinov gives quite an exact fit to the plate data. For bars the factor must be taken as 1.92. These data represent extensions of Chvorinov's relationships by the writers.

Attempts by the writers to calculate solidification times of castings having severe end effects according to these relationships failed as expected. That such cannot be done is fully recognized and predicted by Chvorinov in his discussion of the complicated casting problem. It is to be expected therefore that Dr. Schwartz's experiments likewise would fail to check the quadratic (V/SA)2 relationships. The castings used do not in all cases conform to semi-infinite requirements as indicated by the following:

Casting Section in.	Actual Length, in.	Minimum Length for Semi-Inf Condition to be Developed,* in
2x3	6	12
11/2×3	6	9
1x3	6	6
3/4×3	4	4.5

3

4

1/2×3 While these data may not be considered to show lack of validity of the (V/SA) a relationship for conditions defined by Chvorinov they nevertheless emphasize a very important practica: point relative to calculations involving complex castings. Inasmuch as Chvorinov's contentions have been misunderstood by many Dr. Schwartz has performed a valuable service in focusing attention to the limitations of the Chvorinov concepts. While we may in the not too distant future be able to calculate the solidification times of casting having end effects such is not possible at our present state of knowledge.

REFERENCES

1. H. F. Bishop and W. S. Pellini, "The Contribution of Riser and Chill-Edge Effects to Soundness of Cast Steel Plates," A.F.S.

Transactions, vol. 58, pp. 185-197 (1950).

2. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars," A.F.S. Transactions, vol. 59, pp. 171-180 (1951).

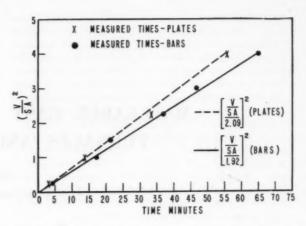


Fig. 10-Solidification of plates and bars as a function of (V/SA) factor.

Dr. Paschkis: Were the castings in your Fig. 10 all poured at the same temperature?

Mr. Pellini: All castings were poured at 2950 F ± 30 deg. DR. PASCHKIS: You state that Chvorinov limits his square law to castings with no end effects. That would mean such simple shapes as cylinders and spheres rather than very long slabs. In his Fig. 19 which he presents as proof for his contention, Chvorinov shows an odd variety of shapes which definitely have end effects-very thin slabs of short length, square bars, etc.-so that his equation is not limited to an infinite slab, as you explained. His figure is presented for bodies with end effects and therefore I think we should disregard what he said in his mathematical derivation because his figure is based on these others.

It is interesting to find that with comparatively limited amounts of temperature variations on which you based your Fig. 10, you still get only about 10 per cent variation between the sphere and slab. What happens when comparing much larger bodies?

You say one can compute solidification of simple shapes. Do you mean by simple mathematics or by approximations found from actual experiments?

Mr. Pellini: Our data were obtained by measurement of solidification times of actual castings. We found that the 2.09 factor used by Chvorinov applies over a limited temperature range of pouring for plates only. It does not apply for bar sections. We agree with Dr. Paschkis that Chvorinov derived the relationships based on no end effects and it was for this condition that validity was shown. He does not provide data to show that the formulae hold for practical castings. However, the empirical curves that Chvorinov gives have scatter to the extent of ± 10 per cent; thus, Chvorinov shows approximate validity of his findings at least to ± 10 per cent. It should also be ob served that his experimental data were obtained on fairly simple shapes. His treatment of the problem of complex castings is exceedingly sketchy and not convincing.

Experimental data by the writers.^{1,2}

MALLEABLE CAST IRON—ANNEALING FURNACES AND ATMOSPHERE

By

O. E. Cullen and R. J. Light*

History

It was the alertness of an American inventor that brought about the introduction of black-heart malleable as an improvement over white-heart malleable iron. To the foundryman goes the credit for the development of this black-heart malleable process to the advanced stage of progress which it now enjoys.

It is a little difficult to understand what factors influenced the hesitation of iron founders in other countries to adopt the new black-heart process after its techniques became generally understood. Perhaps this hesitancy was the result of necessity for change in foundry practice and improvement in techniques. Perhaps it was the fact that early processing of blackheart malleable required time cycles comparable to white-heart processing and therefore did not appear to offer any advantage in the use of the former for the type of castings produced. Whatever the reasons, the fact remains that black-heart malleable iron production soon revolved itself into an entirely new process and now bears no real relationship to the white-heart process either in product or, in many cases, in the time of treatment.

The history of black-heart malleable is one of continued progress. Much has been learned about the reactions involved in converting hard white iron to the soft malleable product. It has been demonstrated by many investigators that the production of black-heart malleable iron is not bound by one inflexible reaction such as the decarburizing reaction employed in the white-heart malleable iron process.

The black-heart malleable iron process has been constantly improved from the standpoints of quality of product. Production rates have been substantially increased by improved annealing practices which shortened the time required for the development of the desired temper carbon structure. Application of these shortened time cycles depends upon many factors including control of scrap, melting methods, control of chemistry, pouring technique, shakeout

practice, mechanical handling equipment and types of annealing furnaces. The degree of success in achieving the ultimate advantages to be derived from the development work in any particular foundry will depend upon that foundry's adherence to this whole scheme of modernization. In any case, precise process control is vital to control the quality of the product.

In the field of furnace design and construction, it has been necessary to provide means for controlling temperature cycles. In modern furnaces these cycles must fully meet the minimum heating and cooling time requirements for bringing about the intra-metal reactions necessary for malleablizing without adding excessive time for interim heating or cooling of heavy pots and packing material. Also, it has been necessary in most instances to provide suitable atmospheres to prevent scaling of open-loaded work which had formerly been more or less protected by being packed in sand or oxide and sealed in heavy containers.

Heat treating equipment has kept abreast or ahead of many other foundry practice improvements to the extent that scale-free malleablized castings are now being produced in commercial furnaces in 20 hr or less. This is a far cry from 160 to 200 hr which was and still is, required in many older types of batch or pack furnaces.

Continuous malleable furnaces are designed with three major objectives in mind. First, temperature control throughout the length of the furnace must be such that each and every tray load of work follows exactly a predetermined time-temperature pattern. Secondly, a suitable controlled atmosphere must be provided to eliminate scaling of the work and at the same time have no deleterious effect upon the malleabilizing reaction within the metal. Thirdly, the basic temperature control arrangement plus the physical design of the furnace must permit considerable flexibility in the matter of work production per hour, varying time cycles for white irons of different analyses and section and, of course, combinations of both.

Furnace Design

A typical modern controlled-atmosphere continuous malleabilizing furnace is shown in Fig. 1. The phofor

TI

th

A

era

SII

br

th

TI

tu

Chief Metallurgist, and Development Engineer, respectively, Surface Combustion Corp., Toledo, Ohio.

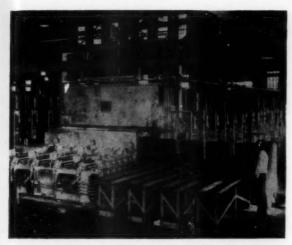


Fig. 1-Photograph of malleablizing furnace.

tograph taken at the charge end of the furnace, shows four loaded trays being moved into the vestibule. Thereafter the work progresses through the furnace malleablizing chamber and is finally discharged at the other end through a vestibule of the same design. A cross-section of the furnace and typical time-temperature curve is shown in Fig. 2.

The furnace is essentially a fabricated and welded gas-tight steel chamber lined with insulation and insulating refractory brick and tile. The insulating brick and tile are fashioned to provide essentially three separate sections within the furnace chamber. These sections represent heating and first-stage anneal, fast cool and slow cool zones.

Suitable heating elements, for example radiant tubes for fuel gas or oil firing, are provided above and below the work as shown in Fig. 2. In the fast cooling zone air-cooled tubes are similarly positioned. Both the heating tubes and the cooling tubes are so designed and installed that no combustion products or cooling air can enter the work holding chamber to contact the work. All tubes are sealed into the furnace casing to prevent leakage around the tubes and products of combustion and cooling air are exhausted outside the furnace.

A number of independent zones of temperature control are set up along the length of the furnace chamber to provide the most suitable heating, holding and cooling pattern for the malleablizing cycle. Seven or more such temperature control zones may be used with special attention being paid to accurate control of the critical slow-cooling section of the furnace. A control thermocouple is installed in each zone to regulate the heat input or the cooling air requirements of the particular zone. In addition provision is made for the insertion of test thermocouples at frequent stations along the length of each zone. Temperature readings taken at these stations supply the information necessary to insure correct adjustment of the fuel input at each heating element of the zone. To those experienced in the art, such flexibility of temperature control will be readily appreciated.

As shown in Fig. 2 each end of the malleablizing chamber is closed by a refractory-lined door. In keeping with the purpose to maintain as near air-tight construction as possible, these doors are provided with liquid-cooled machined faces which contact similarly cooled and machined faces of furnace frames when the doors are in closed position. When both doors are closed therefore, there is substantially no chance for ingress of air or other outside gases into the furnace chamber.

In order to charge and discharge the furnace without pollution of furnace atmosphere, vestibules are

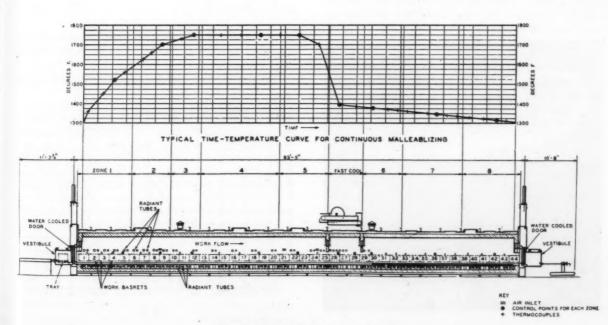


Fig. 2-Malleable furnace and cycle.

provided at each end. These vestibules are also of gas-tight construction.

The work moves into the charge vestibule and out of the discharge vestibule on a transfer car. One end of each of these cars is a solid sealing panel which closes the vestibule tightly when the car is inside. A bypass of controlled size is usually provided around each inner tight door to permit purging of each vestibule with the atmosphere gas inside the malleablizing chamber. Thus it is possible to open either inner door without materially affecting the composition of the furnace chamber atmosphere.

The interior construction of a typical malleablizing furnace chamber is shown in Fig. 3. The refractory arches which permit controlled gradients in heating and cooling are clearly shown. The rails along which the trays are moved through the furnace are well defined. In this particular example of a furnace, three rows of trays move side by side on the six rails. The rails are deeply grooved to permit the use of trays fitted with rollers for easy movement of heavy loads through the furnace. The roller type of tray shown in Fig. 4 is considered to be of definite advantage for many reasons particularly because it eliminates the need for positioning rollers or other moving parts within the furnace where frequent inspection or possible replacement become a serious problem. The possibility for inspection, sand removal and lubrication of the only wearing parts after each passage through the malleable furnace is conducive to economical and trouble-free operation.

Atmospheres

While the first objective of continuous malleablizing furnace equipment was to speed up the malleablizing process by taking full advantage of the improvements in chemical analysis, melt control, and foundry practice, it was a natural step forward to insist on scale-free castings. For this reason the problem of providing suitable furnace atmospheres has occupied the attention of both the furnace manufacturer and the furnace user.

The most logical approach to this problem of furnace atmospheres as related to scaling of the iron is through a study of the equilibrium data relating to the reduction and oxidation of ferrous materials. These are quite familiar to most foundry metallur-



Fig. 3-Photograph of furnace interior.

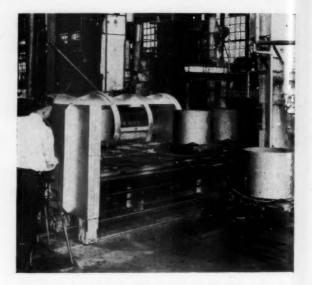


Fig. 4-Photographs of tray and baskets.

gists but will be set forth here as background material for the present discussion.

Thus—

Reactions (2) and (3) are reversible reactions and the equilibrium constants,

$$K_1 = \frac{P_{co}}{P_{co_2}} \tag{4}$$

$$K_2 = \frac{P_{\pi_2}}{P_{\pi_2 o}} . (5)$$

are well known, the values of these constants K at different temperatures can be shown in graphic form (Fig. 5).

There are definite indications in the literature that

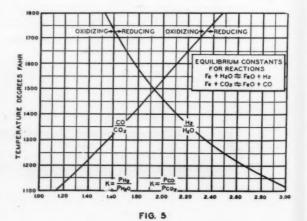


Fig. $5-CO-CO_2 + H_2 - H_2O$ ratios.

diffings the men the ally van furnatm

per

iroi

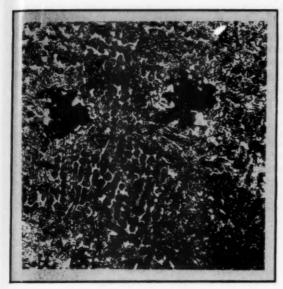
will 2:1 E that stro tard evid mon

a to atm hyd abli pro con pro

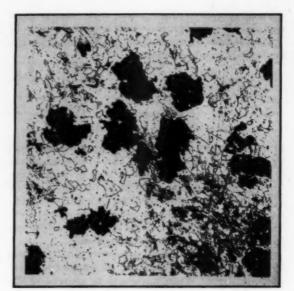
seen aces C not bur

In aces

ann



100 Hr, 40% Hydrogen



30 Hr, 6% Hydrogen

Fig. 6-Photomicrographs of malleable iron.

diffusion of atmosphere gases into the malleable castings during annealing has a considerable effect upon the rate of annealing and upon its quality. Statements have been made by well-known authorities that the diffusion of the oxidizing gas CO₂ should materially aid in the rate and quality of annealing. Advantage is taken of this fact in modern continuous furnaces where it is common practice to maintain an atmosphere containing the maximum CO₂ content permissible, consistent with minimum scaling of the iron. From the equilibrium curve shown in Fig. 5 it will be noted that a CO/CO₂ ratio of approximately 2:1 will safely meet this requirement.

Early in the malleablizing study it was determined that hydrogen is not at all desirable in spite of its strongly reducing properties. Hydrogen definitely retards malleablizing. Actually, there is considerable evidence that the other common reducing gas, carbon monoxide, tends to slow down malleablizing. However, carbon monoxide is not nearly so active in retarding malleablizing as is hydrogen. Figure 6 shows a test bar after a malleablizing cycle of 100 hr in an atmosphere containing approximately 40 per cent hydrogen. The duplicate test bar was fully malleablized in 30 hr in an atmosphere which contained approximately 6 per cent hydrogen. It appears that control of hydrogen is of paramount importance in providing most rapid annealing conditions. This seems to be particularly important in continuous furnaces.

Carbon monoxide, in concentrated amounts while not too detrimental in the first-stage anneal, can carburize the metal surfaces during the second-stage anneal.

In the original development of continuous furnaces, it was realized that success in producing scale-free malleable castings would depend upon maintaining a non-scaling atmosphere in contact with the work.

It would be necessary to prevent air from entering the furnace heating chamber during periodic openings of doors for charging and discharging. It would be necessary to shield the work from flue gases produced by the heating burners as these flue gases are highly oxidizing. It would be necessary to overcome the possible infiltration of air through furnace wall seams, leaky doors and other such possible inlets to the furnace heating chamber.

An original concept, and one still in considerable use, was to provide an atmosphere from an outside generator to insure a positive-controlled atmosphere pressure within the furnace. The most popular type of generator, Fig. 7, for this purpose was an exo-

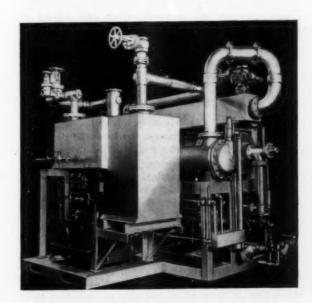


Fig. 7-Photograph of DX unit.

thermal generator which burns hydro-carbon gases such as fuel gas or propane with a limited supply of air to produce an atmosphere gas of approximately the following composition:

It will be noted that the CO/CO₂ ratio satisfactory for annealing without scale. The hydrogen content is controlled although it is present in sufficient quantities to raise some question about its effects on annealing time. Generally speaking such an atmosphere has been considered to be quite satisfactory.

More recently a new concept of furnace atmosphere appears to be of considerable importance.

Continuous furnaces, as previously described, can be so constructed that they are, for all practical purposes, air tight. When work is to be charged or discharged from the furnaces, it is done through tightfitting vestibules at each end. Therefore, the problem of preventing air entrainment becomes one of suitably purging the air from these vestibules after each opening to the outside air but before the inner doors of the vestibules are opened to admit or remove work from the furnace heating chamber. If generator gas with analysis just described were to be used for this vestibule purging, it would appear that there would be only slight advantage in building furnaces of such tightness. However it has been convincingly demonstrated that air-tight furnace construction presents far more opportunities for improved malleable practice than the elimination of an external controlled atmosphere generator. In fact, there are certain applications where it is considered good practice to use a gas generator atmosphere for purging the vestibules of air-tight furnaces. When the generator is so used, it is customary to produce an exothermic gas atmosphere extremely low in combustibles and devoid of free oxygen for example:

CO₂, % - 11
CO, % - 0.5
H₂, % - 0.6
N₂, % - Balance
Dew Point -
$$+$$
 80 F max

From the CO/CO₂ equilibrium values shown in Fig. 5, it will appear that the introduction of an atmosphere such as the above will be oxidizing to the malleable iron if the gas enters either or both ends of the furnace heating chamber. Actually, there is no reason for this oxidizing gas to enter the heating chamber in sufficient quantities to cause scaling of the work.

As fresh loads of work are moved into the heating chamber there are constantly replenished sources for internally generated atmosphere which is usually more than sufficient to satisfy the volume requirements of the heating chamber. This atmosphere may come from one or more of several sources including gases absorbed in the white iron, oxides on the iron sur-

faces and on trays and baskets, as well as purge gas from the vestibule. The gases thus encountered will contain carbon dioxide and some water vapor. Near the charge end of the heating chamber these oxidizing constituents are activated by temperature to combine with some carbon from the white iron according to the following reactions:

0.

dep

ma

gas

far,

to

sur

be

pro

of h

out

a g

ma

seei

qua

zon

titie

doe

car

mos

Th

con

out

rati

Th

cha

con

Th

hea

ing

ann

brie

maing

of

tim

it i

and

tion

the

atta

imp

C

Cings

McC

dev

$$CO_2 + C \approx 2CO$$
 (6)
 $H_2O + C \approx CO + H_2$ (7)

Since the decarburizing reactions preced the oxidizing reactions (2) and (3), the CO/CO₂ ratio quickly builds up to a sufficient value to prevent scaling. Under such conditions it is not unusual to find the gas analysis in the charge end and the high temperature section of the furnace approximately as follows:

CO₂,
$$\%$$
 - 4.0
CO, $\%$ - 20.0
H₂, $\%$ - 6.0
N₂, $\%$ - Balance
Dew Point - + 80 F max

Inasmuch as the reactions (6) and (7) are reversible with temperature changes according to the well known water gas reaction:

$$CO + H_2O \rightarrow H_2 + CO_2$$
 (8)

the CO will drop to about 15 per cent and the $\rm CO_2$ will rise to about 6 per cent in the slow cooling zone of the furnace. A typical curve of $\rm CO/CO_2$ ratios throughout the length of a continuous malleable furnace operating under described conditions is shown in the graph, Fig. 8.

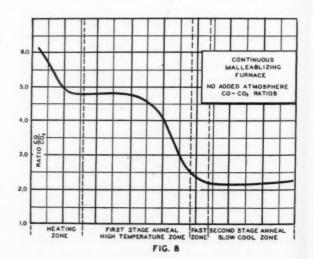


Fig. 8-CO-CO₂ ratios in furnace with no external atmosphere addition.

One advantage of this type of operation is complete elimination of any free oxygen from the system. The scaling propensities of free oxygen at relatively low temperatures are such that it would oxidize the iron at either end of the furnace before it could react with the carbon in the iron. As a result, either RE

18

11

ur

ig

1e

10

ly

g.

e

a-

le

e

11-

11

W

n

t

sponge iron or scale would be the end surface product depending upon whether the free oxygen attacked the iron at the charge end or at the discharge end of the malleablizing chamber. The use of oxygen-free purge gas in the vestibules prevents such reactions. Thus far, the CO₂ in this purge gas has not been present in sufficient quantities within the furnace chamber to oxidize the iron rather than to combine with the surface carbon of the metal.

Quite recently, a rather unusual method has been devised to provide a furnace atmosphere which would be non-scaling and which would more nearly approach what is believed to be a better furnace analysis of higher carbon dioxide and lower hydrogen throughout the length of the malleablizing chamber. Such a gas should meet the requirements for most rapid malleablizing. This has been accomplished by the seemingly unorthodox procedure of adding metered quantities of air directly to the high temperature zones of the furnace. With proper control of quantities and directional flow of the air, free oxygen does not contact the iron but rather is converted to carbon dioxide by interreaction with the furnace atmosphere according to the equation:

$$O_2 + 2 CO \rightarrow 2 CO_2$$
 (9)

The CO/CO₂ ratios can thus be maintained fairly constant in analysis and non-scaling in action throughout the length of the chamber. A typical CO/CO₂ ratio curve for an actual furnace is shown in Fig. 9. The average analysis of atmosphere from this furnace chamber is:

Conclusion

It has been the object of this paper to present to the toundrymen a review of progress in the field of continuous atmosphere annealing of malleable iron. The problems attending the use of one continuous heat treating chamber for all phases of the malleablizing reaction including heating, high temperature anneal, fast cool and slow cool have been mentioned briefly. Various types of atmospheres and their control to prevent scaling and to aid in speeding up malleablizing reactions have been described according to best present knowledge of the subject. In view of the rapid strides which have been made in continuous annealing processes in a relatively few years, it is to be expected that more concrete information and data will prove of practical use in the construction of heat-treating equipment, which will permit the malleable foundrymen to continue their efforts in attaining shorter cycles, lower operating costs and improved malleable iron products.

DISCUSSION

Chairman: W. D. McMillan, International Harvester Co., McCormick Works, Chicago.

Co-Chairman: WM: ZEUNIK, National Malleable & Steel Castings Co., Indianapolis, Ind.

F. Coghlin, Jr. (Written Discussion): 1 Messrs. Cullen and

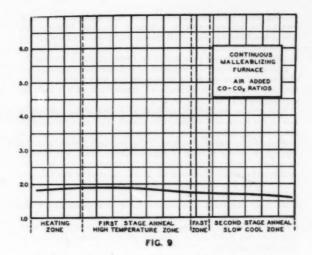


Fig. 9-CO-CO, ratios in furnace with air additions.

Light are to be congratulated on a very excellent presentation of some of the more important facts on continuous annealing equipment and protective atmospheres for the process.

Through personal contact with the authors, I am somewhat familiar with the work they have been doing to develop and perfect equipment for rapid and continuous annealing of white cast iron. Work of this nature has been too scarce in the malleable industry and generally the work that has been done has not been thoroughly appreciated by the majority when presented. It is hoped that this paper will serve as a nucleus for further work and publications on the subject.

Continuous, or so-called rapid, annealing of white cast iron has taken a long time to gain even partial acceptance by the malleable industry. For years, following its inception, many regarded this form of rapid annealing as nothing but a cheap method of producing an inferior grade of malleable iron. Even after the problems encountered with the first units were mastered and temperature control and atmosphere control fell within the scope of close regulation, the attitude of most producers and purchasers of malleable still displayed skepticism. It is my belief that the reason for this, in most instances, was that the degree of control being attained in the oven was not being met by proper metallurgical control of the metal prior to annealing. However, the oven was usually blamed for any trouble that developed.

A continuous oven, when properly set-up, is a highly precise piece of equipment capable of producing malleable iron of the highest quality and uniformity. However, if the iron going into the oven is of inferior quality, or of greatly divergent analysis, these capabilities will be lost.

There are several continuous installations in the country today producing good malleable iron by the use of cycles ranging from 14 to 50 hr. With each favorably operating unit it will be found that annealing success starts with control in the melting department.

The writer is familiar with one continuous oven producing 56,000 tensile, 39,500 yield and 15 to 18 per cent elongation malleable iron in 21 hr.

The newly developed method of creating a protective atmosphere by the introduction of air into the high temperature zone of a continuous oven is interesting and should be of value to users of this type of annealing equipment. It must be cautioned, however, that practical use of the method should be preceded by a thorough analysis of oven conditions and requirements.

E. A. Welander (Written Discussion): The authors are to be commended for their interesting and informative paper. The production experience we have had closely substantiates the statements presented in this paper.

Our first experience with a controlled atmosphere in a con-

¹ Foundry Consultant, Albion, Mich.

² Metallurgist, Union Malleable Iron Works, E. Moline, Ill.

tinuous short cycle annealing furnace was with an externally prepared atmosphere rich in hydrogen. We soon learned that this atmosphere was not at all conducive to rapid annealing. It was necessary to increase our time cycle to over 100 hr in order to produce a product that would be commercially acceptable.

This gas generator was replaced with a charcoal burning type of generator, that produced an atmosphere high carbon monoxide. We were able to maintain an atmosphere with a ratio of

approximately 5 to 1, CO to CO2.

This comparatively rich atmosphere retarded the annealing rate of the metal to a considerable extent. The next step was to discontinue the atmosphere addition, and allow the castings to produce their own atmosphere. This method resulted in an atmosphere in the furnace of approximately 3 to 1, CO to CO₂. In order to further reduce this ratio it was necessary to make an addition of a small amount of air in the hot zone of the furnace. This addition was controlled by the use of an orifice plate and a definite air pressure.

We have continued this practice for several years with satisfactory results, and are able to maintain a ratio of from 1.5 to 1 up to 2 to 1, CO to CO₂. This produces a decarburized skin approximately 0.015 in. in thickness on the castings that is not

considered harmful.

MILTON TILLEY: Why do we have to have CO₃ in the malleable annealing atmosphere? One theory states that the precipitation of the carbon in the iron takes place by means of CO gas moving back and forth in the metal. The carbon dioxide picks up the carbon in the metal, forming carbon monoxide. That carbon monoxide proceeds to deposit that one atom of carbon on the spot of precipitation, so you cannot malleablize unless you have carbon dioxide present. If you have a high carbon monoxide content you slow the operation up to the point where it is not commercially possible.

In a tight container or a tight furnace you can build up the carbon monoxide to as high as 70 per cent at which point the reaction does not go on very rapidly because of the lack of carbon dioxide. In the case where you start with an open pot which you fill with iron you have oxygen in that container starting the reaction on the carbon monoxide which starts the

graphitizing operation.

In the case of a continuous purging atmosphere using a purging atmosphere at the door, you would have no carbon monoxide so you would have to introduce carbon dioxide or oxygen in the antechamber.

CHAIRMAN McMILLAN: I would like to ask Mr. Cullen a question about the atmosphere that contained 40 per cent hydrogen. How was it prepared and what were the other constituents of the

atmosphere, CO, dew point, etc.?

MR. CULLEN: This 40 per cent hydrogen was made in an endothermic gas generator. It contained in addition, about 20 per cent carbon monoxide and 40 per cent nitrogen which was a very good reducing gas as far as we were concerned.

Normally the dew point for that type of gas is around plus 30 F. It was an endothermic generator in which the dew point is controlled by the air-gas mixture supplied to the generator. That is rather an unusual gas atmosphere as far as foundries are concerned and we suspect it was about the only time that it was ever tried out in a foundry.

The carbon monoxide may have had some action, but the main point was the predominance of hydrogen. We came back to our laboratory and checked hydrogen-bearing atmospheres and when we got down to 10 per cent hydrogen and below we started to get good malleablizing results. These laboratory

atmospheres were artificially made hydrogen and nitrogen mixtures so there was no carbon monoxide problem at all. They duplicated what we had observed in the field.

CHAIRMAN McMillan: This was an endothermic job containing some CO. Later you used mixtures of nitrogen and hydrogen. The reason I asked about the high hydrogen atmosphere was to learn if anyone has any idea as to why hydrogen retards anneal, i.e., the mechanism by which hydrogen seems to prevent

migration of carbon

MR. LIGHT: I would like to review some of the effects we observed in the laboratory. We cracked anhydrous ammonia which gave us 75 per cent hydrogen and 25 per cent nitrogen. Using this as an atmosphere we found that the annealability was poor. The next test was run with this 40 per cent hydrogen and 40 per cent carbon monoxide. A flue gas containing 9 per cent CO, 4 per cent CO, and 11 per cent Ho was used next. Then we went to a leaner gas which contained around 11 per cent carbon dioxide and only 0.5 per cent carbon monoxide. Another gas was a nitrogen-bearing gas which had only traces of carbon monoxide and carbon dioxide, and the last test which was run was very unorthodox, you might say. These samples were annealed in no atmosphere, or a vacuum, or as close as we could get to a complete vacuum. If you would plot the annealability against the hydrogen content in all of these gases you would find that the greater the hydrogen content you have, the less it will anneal, so from those tests we concluded that hydrogen is detrimental to annealing.

The water vapor content in the annealing atmosphere was

hri

cas

2 6

use

cor

and

less

pho

bee

low

du

the

effe

ert

are

wil

me

the

1100

am

Ga

phe

hig

ma

and

noo mu

for It v 0.20 Wo

Bran 51-43

approximately 0.5 to 1 per cent.

F. J. PRYATEL: 4 Can the dew point be used as a control for

this process?

MR. CULLEN: No, since we are dealing with an atmosphere consisting of a mixture of gases at various temperatures. It will always contain carbon dioxide, some hydrogen and carbon monoxide. Unfortunately, the water-gas reaction tends to set up a definite ratio of water vapor, hydrogen, carbon monoxide and carbon dioxide at any one temperature and this ratio varies with temperature. Since a large portion of the furnace chamber is at low temperature and considerable diffusion of gases can take place, it would be hard to evaluate atmosphere by dew point. It settles out at some rather constant value however.

CO-CHAIRMAN ZEUNIK: The authors showed a curve illustrating the relationship of carbon monoxide and carbon dioxide in the annealing atmosphere. Apparently that gas is under pressure. Would not diffusion cause those gases to remain the same

throughout the entire furnace?

Mr. Cullen: The only time we have ever found the conditions referred to was when there was no atmosphere put into the furnace proper. A flow of very lean flue gas, 11 per cent carbon dioxide and 0.5 per cent carbon monoxide and 0.5 per cent hydrogen, was used in the furnace vestibules. This furnace is 80 ft. long.

The carbon monoxide-carbon dioxide ratio would change a little as you opened the doors, but generally it did maintain the described characteristics. When we introduce air or put a rich flue gas into the furnace, we do not get this variation in gas

ratios.

Mr. Light: I would like to add one more comment on these laboratory tests which we ran. The samples that we annealed in the vacuum were annealed completely in one-half the normal cycle time. I just thought that would be of some interest and it may be more or less of a hint as to what mechanics take place in the anneal.

Metallurgist, National Malleable & Steel Castings Co., Cleveland.

⁴ Chief Metallurgist, Oliver Corp., South Bend, Ind.

EFFECT OF PHOSPHORUS CONTENT ON MECHANICAL PROPERTIES OF A NODULAR CAST IRON*

By

J. E. Rehder **

It is well known that phosphorus has an embrittling effect on most ferrous metals, such as steel, cast iron, and malleable iron. Since gray cast iron is not a ductile material, the phosporus contents normally used are from 0.20 to 0.60 per cent, since the embrittling effect is of less relative importance; phosphorus contributes to some extent to the fluidity of cast iron; and higher phosphorus content pig irons are usually less expensive.

With the advent of nodular cast irons which possess appreciable ductility as cast, the embrittling effect of phosphorus became of more importance, and it has been generally recognized that it is desirable to have low phosphorus content in nodular irons if the highest ductility and impact strengths are to be obtained. It is the object of this paper to explore quantitatively the effect of phosphorus content on the mechanical properties of nodular iron. Magnesium-treated irons only are discussed herein but it is believed that the results will be applicable to nodular irons made by other means.

Previous Work

There has been little quantitative data published on the effects of phosphorus content on the properties of nodular iron. Donoho 1 stated that phosphorus in amounts more than about 0.15 per cent is undesirable because of increased shrinkage and lower ductility. Gagnebin, Millis, and Pilling 2 showed that use of a phosphorus content higher than 0.05 per cent (the higher amount not stated) resulted in decreased ultimate strength, Izod impact strength, and elongation, and increased yield strength and hardness in as-cast nodular irons. It was shown that the presence of as much as 0.65 per cent phosphorus did not inhibit the formation of nodular or spherulitic graphite structure. It was further stated that phosphorus in amounts over 0.20 per cent promotes an embrittling phosphide network resulting in moderately lower strength.

The effects of phosphorus content on the graphitization rate and mechanical properties of black-heart malleable iron were investigated some time ago by the present writer as a University thesis (unpublished) and the conclusions are of interest and believed applicable here. Although there is difference of opinion in the literature on the effects of phosphorus content on graphitization rates, the thesis showed clearly that phosphorus has a definite, though not large, retarding action on graphitization rates in both first and second stage annealing. Increasing phosphorus content increased the ultimate strength to a minor extent, and seriously decreased the impact strength. It was further shown that a first-stage annealing temperature of 954 C (1750 F) resulted in complete break-up and spheroidization of the phosphide eutectic areas where there was sufficient phosphorus present to form phosphide eutectic as cast.

With respect to the effect of phosphorus on graphitization rate, it is a matter of common observation in cast irons that where phosphide eutectic is present, the last traces of pearlite invariably cling around the phosphide; a phenomenon easily observed in gray cast irons annealed for better machinability.

Experimental Method

Two heats of nodular cast iron were made in a 500-lb capacity induction furnace in a routine way, using identical charges of steel scrap, nodular iron returns, low-phosphorus pig iron, and wash metal. The metal was tapped at 1482 C (2700 F) into a 700-lb capacity bull ladle, a magnesium-ferrosilicon addition being made with half in the bottom of the ladle and half added to the stream. After filling, the ladle was skimmed and the iron tapped in 100-lb quantities into a preheated 150-lb capacity pouring ladle. The phosphorus, as crushed ferrophosphorus, and the post-inoculant of crushed 50 per cent ferrosilicon were mixed and added to the stream while filling the 150-lb ladle. To achieve increasing phosphorus contents, the amount of ferrophosphorus was increased for each 100-lb tap, the amount of ferrosilicon remaining constant. After stirring and skimming each tap, four keel blocks in dry sand cores were poured at each phosphorus level.

The described method of introducing phosphorus

ERF

hey aindrohere ards

onia ren.

gen per ext.

aces nich ples

eal-

the

was

for

ere

vill

on

nd

ith

16

ike

It

he

re.

me

di.

ote

me

ce

he

ch

as

ed

al

10

ce

^{*} Published by permission of the Director-General of Scientific Services, Department of Mines and Technical Surveys, Ottawa.

^{**} Foundry Engineer, Physical Metallurgy Division, Mines Branch, Ottawa, Ontario, Canada.

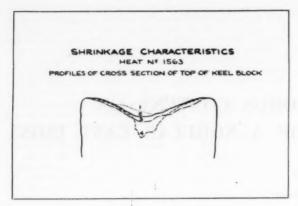


Fig. 1-Shrinkage characteristics.

was used in an attempt to avoid changes in magnesium content, and was successful as inspection of chemical analyses in Table 1 will show. The present writer has found that under certain poorly-defined conditions magnesium will act as a dephosphorizing agent (e.g. addition of a mixture of magnesium and ferrosiliconzirconium to melted pig iron of 0.16 per cent phosphorus content at 1427 C (2600 F) resulted in three cases in dephosphorization of 26 to 57 per cent) and it was desired to maintain magnesium content uniform in the present work. The mixing of the ferrophosphorus and ferrosilicon before addition to the iron in the present case prevented loss of magnesium, possibly because of the reducing nature of the ferrosilicon.

Chemical Analyses

Chemical analyses were performed on drillings from the keel blocks, with the exception of carbon content which was determined on small pencils cut from the coupons and combusted as a unit. The sulphur contents were determined by the combustion method, and the magnesium contents by the wet, mercury cathode method. Phosphorus was determined by a standard gravimetric method.

The keel blocks were allowed to cool in the sand, and after shaking out, the legs were cut from the blocks with a band saw. Since a marked increase in the surface shrinkage was noted with increasing phosphorus content, the tops of the keel blocks were cut transversely to cross-section the shrinkage, and tracings made of the depth of shrink.

From each phosphorus level, two keel block coupons were left as cast; two oil quenched from 1 hr at 843 C (1550 F) and drawn 3 hr at 621 C (1150 F); two fully annealed by heating to 843 C (1550 F) and slow cooling to 704 C (1300 F); and two fully annealed by heating to 954 C (1750 F), cooling in 20 min to 816 C (1500 F), and slow cooling to 704 C (1300 F). The Brinell hardness of each coupon was measured, and then a standard 0.505-in. diameter 2-in. gage length tensile test bar was machined from each coupon. Dumbbell-ended bars with self-adjusting grips were used. Mechanical properties were determined on a 60,000-lb Baldwin Southwark tensile machine with recording extensometer, and on two Olsen impact testing machines. The yield strengths reported were at 0.2 per cent offset, and the notched Charpy impact bars had a standard "V" notch.

Tensile tests were all made and reported in duplicate, and the impact test results reported are the average of four specimens each. Values of impact strength below 2.0 ft-lb are not considered reliable.

Metallographic examination was performed on samples cut from broken tensile test bars.

Results

1543

1543

156

156

156

Chemical analyses of the eight sets of bars, four sets from each heat, are given in Table 1.

TABLE 1

Bar No.	C	Si	Mn	S	P	Mg
1543-1	3.08	2.92	0.35	0.011	0.022	0.059
1543-2	enen.	2.85	_	-	0.062	0.056
1543-3	_	2.85	_	_	0.156	0.058
1543-4	3.11	2.87	0.35	-	0.275	0.053
1563-1	3.06	2.60	0.36	0.012	0.022	0.049
1563-2	-	_	_	_	0.067	_
1563-3	_	-	-	_	0.144	_
1563-4	3.10	2.70	0.38	0.014	0.222	0.049

In Fig. 1 is shown the relative amounts of shrinkage encountered in the keel blocks. No quantitative measurements were made, but the qualitative indication is clear that shrinkage increases with phosphorus content.

The results of mechanical tests of the bars are given in Tables 2 to 5 inclusive and in graphical form in Fig. 2 to 5 inclusive.

TABLE 2-MECHANICAL PROPERTIES As-CAST

.0	Ultimate Strength, psi	Vield Strength, psi	Elong. Per Cent	Modulus Elasticity psi × 10-6	Brinell Hardness Number	Charpy Impact, ft lb	
Bar No.						Un- notched	Notched
1543-1	85,300 87,500	56,100 56,600	15.0 14.0	197	255 255	45	3.5
1543-2	87,800 91,700	60,200 61,200	6.0 7.5	217	250 250	14	2.9
1543-3	83,600	54,800	3.0	225	250 235	6	1.1
1543-4	71,800 84,200	69,200 65,200	1.5	235	240 245	3	0.9

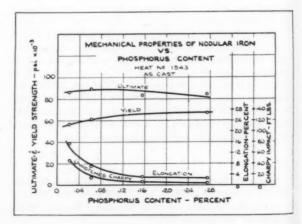


Fig. 2-Mechanical properties us phosphorus content as cast.

RON

ate,

e of

am-

sets

Mg

0.056

0.053

.049

.049

age

eas-

n is ent. ent. en

ct,

hed

ıŧ

Table 3-Mechanical Properties Annealed From 1550 F

Bar No. Ultimate Strength,	gth,	gth,	ent	city 10-6	ll ness	Charpy Impact, ft lb	
	Yield Streng psi	Elong. Per Cent	Modul Elastic psi X	Brinell Hardne Numbe	Un- notched	Notched	
1543-1A	68,600 68,700	51,000 50,750	24.0 23.5	24.0 25.0	167	115	13.2
1543-2A	70,000 70,400	53,600 54,600	24.0 22.0	25.0 23.0	170	105	9.2
1543-3A	71,350 71,700	56,000 57,200	18.0 17.5	25.0 25.0	174	36	2.7
1543-4A Note:	65,600 61,300 Both 1543-4A	57,700 58,700 bars were	4.5 3.0 slightly	23.5 22.8 flawed	181	11	2.0

Table 4-Mechanical Properties Annealed From 1750 F

Bar No.	Ultimate Strength, psi	Yield Strength, psi	Elong. Per Cent	Modulus Elasticity psi × 10-6	Brinell Hardness Number	Charpy Impact, ft lb	
						Un- notched	Notched
1563-1A	65,200 65,000	48,800 48,500	24.0 23.0	24.5 24.0	148	123	14.6
1563-2A	67,400 67,400	50,500 50,500	24.5 24.5	24.5	159	109	14.6
1563-3A	70,700 70,400	54,000 53,700	20.0 21.0	24.0 25.0	164	73	10.3
1563-4A	72,400 71,400	56,500 55,500	16.0 15.0	24.5 24.5	170	34	3.0

Table 5-Mechanical Properties Oil-Quenched and Drawn

Bar No.	Ultimate Strength, psi	Yield Strength, psi	Elong. Per Cent	Modulus Elasticity psi X 10-6	Brinell Hardness Number	Charpy Impact, ft lb	
						Un- notched	Notched
1563-1D	96,100 97,100	74,700 76,200	10.5 10.5	24.0 25.0	264	53	6.8
1563-2D	131,700 131,000	102,700 102,000	7.0 6.5	25.0 23.5	274	40	4.3
1563-3D	128,000 130,500	101,000 105,000	4.0	25.0 24.5	289	19	2.1
1563-4D	119,000 116,700	104,700 103,000	2.0	24.0 24.0	293	12	1.6

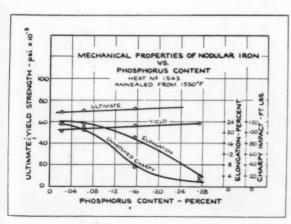


Fig. 3—Mechanical properties vs phosphorus content, annealed from 1550 F.

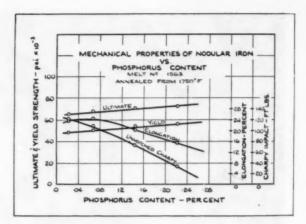


Fig. 4—Mechanical properties vs phosphorus content, annealed from 1750 F.

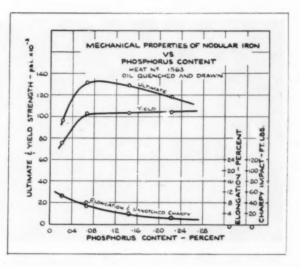


Fig. 5-Mechanical properties vs phosphorus content, oil quenched and drawn.

It was noted as a consistent feature of all of the tensile bars that as the phosphorus content of any series increased, the fracture became lighter in color. This was particularly noticeable in the two fully-annealed series, the lowest phosphorus content bars being a soft black in the fracture and the highest phosphorus content bars being completely a sparkling crystalline fracture.

The microstructures of the bars are of interest, and provide explanations for most of the changes in properties. The as cast bars are shown in Fig. 6 to 9 inclusive, and it is evident that the amount of pearlite increases with phosphorus content. The pearlite was of increasing fineness as phosphorus content increased, the pearlite in the 0.022 P iron being mostly resolved at 750X, while the pearlite in the 0.275 P iron was nearly all unresolved at 750X. The nodules become slightly more regular in outline and there is more uniformity of size of nodules as phosphorus content in-

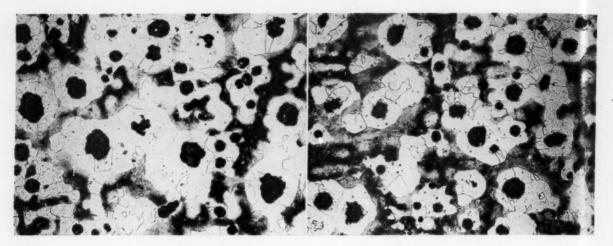


Fig. 6-Bar 1543-1, as cast, 0.022 P. X100, nital etch.

Fig. 7-Bar 1543-2, as cast, 0.062 P. X100, nital etch.

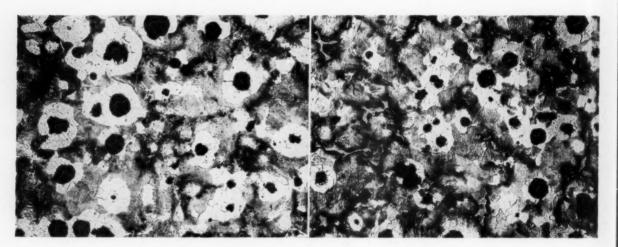


Fig. 8-Bar 1543-3, as cast, 0.156 P. X100, nital etch.

Fig. 9-Bar 1543-4, as cast, 0.275 P. X100, nital etch.

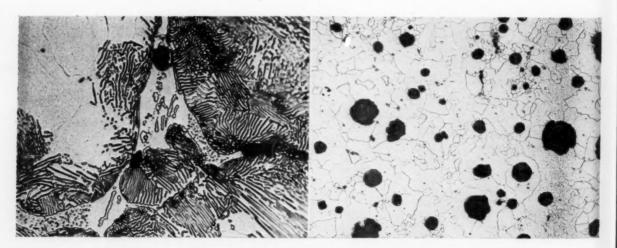


Fig. 10-Bar 1543-4, as cast, 0.275 P. X750, nital etch.

Fig. 11-Bar 1543-1A, annealed from 1550 F, 0.022 P. X100, nital etch.

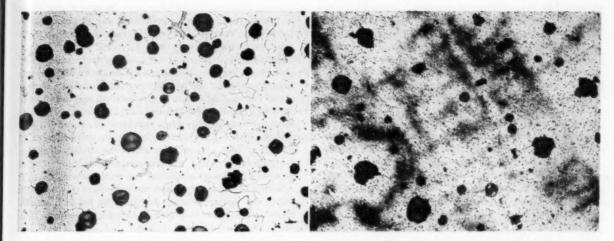


Fig. 12-Bar 1543-4A, annealed from 1550 F, 0.275 P. X100, nital etch.

Fig. 13-Bar 1563-1D, quenched and drawn at 1150 F. 0.022 P. X100, nital etch.

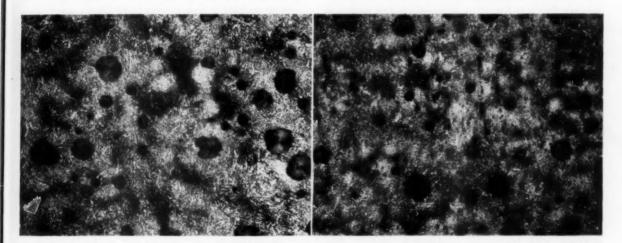


Fig. 14-Bar 1563-2D, quenched and drawn at 1150 F. 0.067 P. X100, nital etch.

Fig. 15-Bar 1563-3D, quenched and drawn at 1150 F. 0.144 P. X100, nital etch.

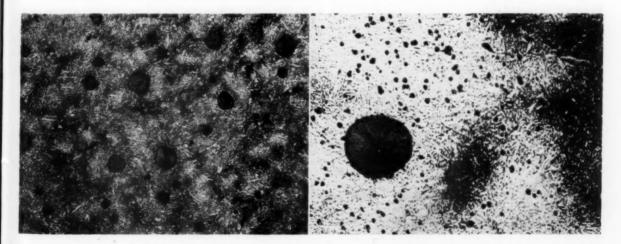


Fig. 16-Bar 1563-4D, quenched and drawn at 1150 F. 0.222 P. X100, nital etch.

Fig. 17-Bar 1563-1D, quenched and drawn at 1150 F. 0.022 P. X400, nital etch.

in

an

pa

SIL

on

on

(as

ph

CO

the

pre

an

tri

ult

no

sho

wil

po

firs

ten

op

wa

it 1

in

Fig

mo

(17

per

tou

ten

tha

in t ma of fro nea pho cate leve ma dec ver crea crea ten stre T crea the que agg

case

creases. The first evidence of phosphide eutectic was in the 0.144 per cent phosphorus bar, and it was estimated that the first appearance of phosphide eutectic would

occur at 0.08 to 0.09 per cent phosphorus.

A point worthy of note is that whenever phosphide eutectic was present in the structure, the brownish-black unidentified inclusions that the present writer and others have noted before in magnesium-treated nodular irons, were almost invariably associated with the phosphide eutectic areas, as shown in Fig. 10. Usually these inclusions were in the eutectic with clear evidence of having been present before the phosphide eutectic was formed. It is believed that this is coincidence, the phosphide eutectic areas and the inclusions being located in the last metal to solidify.

In Fig. 11 and 12 are shown the microstructures of the fully annealed bars 1543–1A and 1543–4A (annealed from 843 C (1550 F)). The structures of bars 1563–1A and 1563–4A (annealed from 954 C (1750 F)) were almost identical, except that the phosphide eutectic in 1563–4A had spheroidized completely.

The microstructures of the oil-quenched and drawn bars are shown in Fig. 13 to 16 inclusive. The matrix is a tempered martensite, partially graphitized, with the as-cast dendritic structure clearly visible. As the phosphorus content increases, the amount of graphitization of the tempered martensite decreases. The graphite resulting from this secondary graphitization precipitates as a dispersion of much smaller nodules, which is characteristic of these conditions as shown previously for quenched and pre-quenched malleable iron. The structure is shown more clearly in Fig. 17. In Fig. 18 and 19 are shown the microstructures of the lowest and highest phosphorus content quenched and drawn bars, wherein the difference in amount of graphitization, and possibly spheroidization, may be seen.

Discussion of Results

Liquid Shrinkage—It is evident from Fig. 1 that the liquid shrinkage of the nodular irons considered here increases with increasing phosphorus content, and heavier risering is therefore necessary as phosphorus

content increases in nodular iron. However, the effect is apparently not large until amounts of phosphorus are present that are intolerable from the viewpoint of good mechanical properties.

The carbon content of the nodular irons considered here may appear to be low, but it has been the writer's experience that when magnesium-ferrosilicon master alloys containing about 10 per cent of magnesium are used to introduce magnesium into the molten iron, the final carbon content, when determined on pencils and not on drillings, is usually in the range 3.00 to 3.15 per cent. This is true even when the molten iron before addition of the alloy contains over 4.0 per cent carbon. It is believed due to the well-known fact that addition of silicon decreases the solubility of carbon in iron, with apparently an overshooting on the equilibrium carbon value due to the sudden addition of a relatively large quantity of silicon.

There is apparently no disadvantage in this phenomenon, since such irons show excellent fluidity even at low phosphorus content, and the mechanical properties are, if anything, improved. In recent tests individual motor-car piston rings have been poured with no misruns and sharp edges in green-sand molds with the same iron as in 1543–1, at a temperature in the pouring ladle of 1320 C (2408 F) as measured with a platinum, platinum-rhodium thermocouple.

Mechanical Properties

It is considered that the uniformity of composition within each series of phosphorus levels is within the limits of error of chemical determinations, so that the test results may be considered as truly representative of the effects of phosphorus content alone. In fact the analyses of the two heats may be considered as identical without serious error. There is excellent internal uniformity in all of the mechanical test results which were done in duplicate or quadruplicate, and it would appear justifiable to consider the averaged results as representative of real changes in the properties measured.

Inspection of Fig. 2 shows that in the as-cast state, increasing phosphorus content results in little change

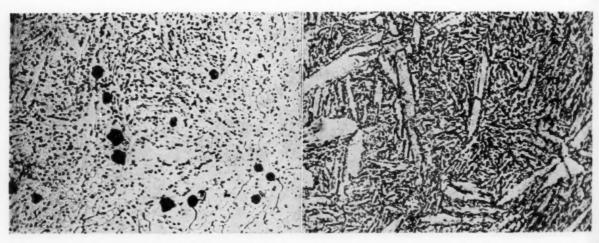


Fig. 18-Bar 1563-1D, quenched and drawn at 1150 F. 0.022 P. X1500, nital etch.

Fig. 19-Bar 1563-4D, quenched and drawn at 1150 F. 0.222 P. X1500, nital etch.

RON

fect

rus

of

red

er's

ster

are

the

and

.15

be-

ent

hat

lib-

ela-

he-

ven

op-

di-

ith

ith

the

h a

ion

the

the

ive

the

cal

ni-

ere

ap-

ep-

ed.

ite,

nge

in ultimate strength, a small increase in yield strength, and relatively sharp decreases in elongation and impact strength. The falls in elongation and impact strength are serious, and for best toughness and ductility as-cast, the phosphorus content must be as low as possible. The effect of increasing phosphorus content on as-cast nodular iron is the combination of two effects, one due to strengthening and embrittling of the ferrite (as will be evident below in the annealed bars) and the other due to an increase in the amount of pearlite present due to the carbide stabilizing influence of phosphorus. For this reason increasing the phosphorus content increases the Brinell hardness more rapidly in the as-cast state than in the fully-annealed state. Approximately, each doubling of the phosphorus content above 0.02 per cent results in halving the elongation and impact strength as-cast in the irons tested.

Fully-Annealed State

In the fully-annealed state (completely ferritic matrix) increasing the phosphorus content increases the ultimate strength and yield strength uniformly, though not to a large extent. Inspection of Fig. 3 and 4 also shows that the elongation and impact strength decrease with increasing phosphorus content, but with two important differences from the case with as-cast bars. The first difference is that the shape of the phosphorus content—elongation (and impact strength) curves are the opposite of those for the as-cast state, being convex upward rather than concave. This is very important since it means that a higher phosphorus content is tolerable in fully-annealed nodular iron than in the same iron as-cast.

The second difference apparent from comparison of Fig. 3 and 4 is that if the annealing operation for removal of pearlite is preceded by a temperature of 954 C (1750 F) to spheroidize the phosphide eutectic, the permissible phosphorus level above which decrease in toughness is serious, is increased. The use of the higher temperature is of important additional advantage in that any primary carbide or chill that may be present in the as-cast iron is also rapidly decomposed. It is estimated that for the subject irons, a phosphorus content of 0.12 per cent is permissible if the iron is annealed from 843 C (1550 F) and 0.16 per cent if the iron is annealed from 954 C (1750 F).

In the oil-quenched and drawn state the effect of phosphorus on mechanical properties is more complicated. As phosphorus content is increased from a low level the ultimate strength and yield strength rise to a maximum level, from which the ultimate strength then decreases while the yield strength is constant or rises very slightly. The elongation and impact strength decrease fairly uniformly as phosphorus content increases. Approximately, increasing the phosphorus content by three times decreases the elongation and impact strength by one-third of their former values.

The reason for the rise in strength with the first increase in phosphorus content is considered evident in the accompanying changes in microstructure. When a quenched, martensitic bar of cast iron (or nodular iron or malleable iron) is reheated, the martensite tempers or agglomerates to spheroidized pearlite exactly as in the case of a extectoid steel similarly treated, the amount of

agglomeration or spheroidization increasing as the draw or tempering temperature is ancreased. However, as soon as the draw temperature reaches levels of 538 C to 593 C (1000 F to 1100 F), graphitization or decomposition of the fine carbides becomes appreciable at a rate exponentially related to the temperature and to the silicon content (i.e. the tempered martensite will start to graphitize at a lower temperature, or will graphitize more completely at any given temperature, as the silicon content of the iron increases). In the irons considered here the silicon content is relatively high, and when phosphorus content is low, appreciable graphitization of the tempered martensite will take place in an hour or more at 621 C (1150 F). As phosphorus content increases, of course less graphitization will take place in a given time, other things remaining the same.

It will be noted that the quenched and drawn bars show the primary dendritic structure, especially clearly in the low-phosphorus 1563–1D bar. Examination at higher magnification shows further that graphitization has proceeded farther in the areas corresponding to the locations of the as-cast dendrites. This is another instance of the phenomenon noted several times previously by the present writer in malleable irons, that marked segregation apparently takes place in cast irons on freezing. The segregation is believed primarly that of silicon, and the higher silicon areas have not only higher hardenability (or more sluggish austenite-martensite transformation) but show more rapid graphitization.

A further point to be noted is that whenever martensites in cast irons are graphitized, the free carbon produced does not precipitate on existing graphite nodules or flakes, but precipitates in a large number of individual locations. Since the total free carbon produced is small (about 0.55 per cent in a 2.50 per cent silicon iron) none of the new graphite precipitation centers grow to appreciable size. This phenomenon has been well explored and reported for malleable irons, but the reasons or underlying theory are completely unknown at present.

The photomicrographs in Fig. 13 to 17 inclusive of the quenched and drawn bars are understandable in view of the above considerations. At the lowest phosphorus level (bar 1563-1D) appreciable graphitization of the tempered martensite has taken place, and the finely divided graphite is clearly shown in Fig. 17. It would be expected that such a structure, containing lower combined carbon content, would have lower strength. As the phosphorus content increases, the amount of graphitization in these bars decreases, with resulting rise in combined carbon and strength. Apparently there is an optimum phosphorus content at which the best combination of yield strength and elongation is obtained, and this phosphorus content is estimated for the subject irons as 0.07 to 0.10 per cent. It will be noted that this is the range which was estimated above as the limit of solid solubility of phosphorus in these irons. It is not known at present whether this coincidence extends to other draw temperatures and

Considerable work has been done in these laboratories on heat-treated nodular irons (normalized, quenched and drawn, partially or fully annealed) and it has been consistently evident that the presence of pearlite gives the poorest combination of strength and toughness or ductility. The optimum properties are obtained in nodular iron when the iron is either fully annealed, for maximum ductility, or quenched and drawn for higher strengths. The results described above are in accordance with the previously obtained data. Nodular iron as-cast has excellent properties, especially when compared with gray cast iron, but a further marked improvement in properties is obtained by a suitable heat-treatment.

Another consistent feature noted in other work and evident in the results given above, is that the modulus of elasticity is essentially uniform and independent of the matrix structure of the iron.

Conclusions

From the above data and discussion, the following conclusions may be drawn:

1. Increasing the phosphorus content of nodular iron from a low level results in slightly increased strength, and more or less seriously decreased elongation and impact strength.

2. In the as-cast state, the elongation and impact strength drop rapidly as phosphorus content is increased, decreasing by about one-half for every doubling of the phosphorus content in the irons tested. For maximum ductility as cast, the phosphorus content of nodular iron must be at a low level.

3. In the fully-annealed (completely ferritic) state, a higher phosphorus content is tolerable than in the as-cast state, and this tolerable level is increased by annealing from a temperature sufficiently high to spheroidize phosphide eutectic. For the irons tested, the permissible phosphorus content is estimated at 0.12 per cent if annealed from 843 C (1550 F) and 0.16 per cent if annealed from 954 C (1750 F).

4. In the oil-quenched and drawn state, for the irons tested an optimum phosphorus content apparently exists, below which ultimate and yield strength decrease rapidly and above which elongation and impact strength decrease less rapidly. For the irons tested, increasing the phosphorus content by three times decreases the elongation and impact strength by about one-third of their former values.

5. For the irons tested, the maximum solid solubility of phosphorus, above which free phosphide eutectic appears, is estimated as 0.08 to 0.10 per cent phosphorus. This coincides with the optimum phosphorus content for the same irons in the oil-quenched and drawn (1150 F) state.

6. Phosphorus apparently is a retarder of graphitization, at least in and below the critical temperature range.

7. Pearlite lamellae become finer as phosphorus content increases.

8. The brownish-black unidentified inclusion com-

mon in magnesium-treated irons is nearly always associated with phosphide eutectic where the latter exists, and evidently pre-exists the phosphide eutectic.

The liquid shrinkage of nodular iron increases with phosphorus content.

Bibliography

Donoho, AMERICAN FOUNDRYMAN, vol. 15, no. 2, (Feb. 1949).
 Gagnebin, Millis, and Pilling, Iron Age, vol. 163, no. 7, (Feb. 17, 1949).

DISCUSSION

Chairman: A. P. GAGNEBIN, International Nickel Co., New York.

Co-Chairmen: T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa.; W. B. McFerrin, Haynes Stellite Div., Kokomo, Ind.

j. T. Mackenzie: ¹ I cannot agree with Mr. Rehder that the carbon always comes out from 3.00 to 3.15 per cent. I was looking at some results of recent tests we made with about a 4.10 total carbon in magnesium-treated iron. We cast a series of bars from 0.6 in. to 3 in. The carbon on the 0.6-in. bars was 4.14 and it gradually fell off to 3.80 on the 3-in. bars. But I have never yet seen one drop from 4 per cent to 3 per cent in normal handling.

gati

allo

0.35

diag

and

air-

thei

500

gini

-20

1300

and

upo

acer

bair

feat

pro

650

hr,

At

tem

abon

soak

alte

is s

the

cool

rece

trai

stu

tion

Proc gan, Rese

MR. REHDER: I appreciate Dr. MacKenzie's remarks. I said I thought probably there would be a difference of opinion. I know other people are using magnesium ferro-silicon and are obtaining drops in carbon, and I have encountered it frequently. I believe the carbon analyses are correct. Actually I have not paid much attention to the phenomenon, since there has been little trouble with fluidity of the iron. I am pleased that Dr. MacKenzie mentioned the point, but I am not sure of its practical importance.

E. SPIRE: ² Is elongation really the most important property of nodular iron? Is not yield, the modulus of elasticity and the bending characteristics also important?

MR. REHDER: When one gives a paper like this, it is rather difficult to avoid giving the impression that one is concentrating on one particular property and saying that is the important one. I talked about elongation because I was using it as a measure of toughness, and the values for yield strength and modulus of elasticity are given in the paper. If you are going to use the iron for a given application and if you know the properties you want in the application, draw your own conclusions from the data presented. There is no inference that any property is the most important. It depends entirely upon the application. That is just as true of nodular iron as it is of any other material.

C. F. Walton: ^a I noticed in your microstructures evidence of a cored or dendritic structure. What element do you attribute that to?

Mr. Rehder: The dendritic structure visible in some of the photomicrographs is the sort of thing commonly visible in gray irons when conditions of heat treatment are suitable. Evidence now available shows that silicon is micro-segregated in cast iron-carbon-silicon alloys, and it concentrated largely in the primary dendrities.

In the low-phosphorus sample in the present paper the dendritic structure appeared because the amount of graphitization happened to be about right for good contrast. A less, or more, completely graphitized structure would be less revealing in this respect.

¹ Technical Director, American Cast Iron Pipe Co., Birmingham, Ala.

² Chief Metallurgist, Canadian Liquid Air Co., Ltd., Montreal, Canada.

⁸ Case Institute of Technology, Cleveland.

ISOTHERMAL TRANSFORMATION CHARACTERISTICS ON DIRECT COOLING OF ALLOYED WHITE IRON

By

F. B. Rote, * G. A. Conger, ** K. A. DeLonge ***

ABSTRACT

This paper represents the results obtained in an investigation of the isothermal transformation characteristics of an alloyed white iron containing nominally 3.40 T.C., 1.15 Si, 0.35 Mn, 4.50 Ni, 1.85 Cr, and 0.12 Mo. A complete T-T-T diagram was determined from data obtained by hardness tests and microscopic examination of wedge-shaped castings, directly air-cooled from the mold to sub-eutectoid temperatures for isothermal transformation.

The T-T-T diagram for this iron has two "knees;" one at 500 F, and the other at 1000 F. The time required for beginning of transformation is 1.3 hr at 1000 F and 1.4 hr at 500 F. Transformation is complete in 6 hr at 1000 F and in 32 hr at 500 F. The M, temperature is 200 F, and M_f is about -200 F. The upper limit of the diagram is near, but above 1300 F. In the range of temperature between 650 F and 850 F, essentially no transformation occurs in 120 hr.

Hardness data provided a rough approximation of the shape and position of the T-T-T diagram, but could not be relied upon for accuracy, due to the complexity of the structures in the alloyed iron. Microscopic studies were used to secure accurate, reliable data.

In the temperature range 300 F to 500 F, a fine needle-like bainite is produced, and between 500 F and 650 F open, feathery bainite. The latter is softer than the fine bainite produced at the lower temperatures. In the range between 650 F and 850 F no isothermal product was produced in 120 hr, so the appearance of products in this range is not known. At higher temperatures, up to 1300 F, pearlitic products are obtained. These are unresolvable at 500X, when formed at temperatures below 1100 F and are spheroidized at temperatures above 1100 F.

Austenite "conditioning" by slow cooling from 1700 F or soaking at 1400 F, each of which provides homogenization, alters the position of the T-T-T diagram. The pearlite nose is shifted slightly to the right, the bainite nose slightly to the left. The M, temperature is raised to above 400 F by slow cooling from 1700 F and to 400 F by soaking at 1400 F.

Introduction

SEVERAL INVESTIGATORS HAVE REPORTED in the recent past the results of studies of the isothermal transformation characteristics of alloyed irons. Their studies have been confined, in general, to small sections removed from larger castings which had been

allowed to cool to room temperature. Upon reheating for austenitization, these sections reach at least virtual equilibrium, and results secured in isothermal transformations are for uniform, homogenized austenite.

Actual castings are not normally subjected to heat treatments, and the microstructures which occur in alloyed irons are those which are produced during continuous cooling. Due to the rapid rate of solidification and cooling of most such castings, the austenite never reaches equilibrium composition, and the transformation characteristics are quite different from those observed in re-austenitized specimens of the same nominal analysis and structure.

In order to investigate the actual transformation characteristics of directly-cooled and of homogenized castings, the present investigation was undertaken. The iron selected for study was an alloyed iron which is normally cast in chillers to produce a hard, abrasion resistant, martensitic white structure in the chill affected area, but with some graphite at depth. Its nominal composition was 3.40 T. C., 1.15 Si, 0.35 Mn, 0.06 S, 0.07 P, 4.50 Ni, 1.85 Cr, and 0.12 Mo.

The isothermal transformation characteristics of the iron as directly air-cooled after shakeout at 1900 F, and after homogenization, were studied. A complete T-T-T diagram, constructed from studies of 15 isotherms from 300 F to 1300 F, was obtained for direct-cooled castings, and studies were made at 500 F and 1000 F for homogenized castings.

Experimental Procedure

Because of the large number of castings required for the complete investigation of the transformation characteristics of the alloyed iron, it was necessary to produce several melts and to standardize on a fixed procedure. This procedure was based on normal industrial practice. For simplicity, each of the major steps employed will be discussed individually.

Melting and Handling Practice—Metal for this study was produced in a 200-lb indirect arc Detroit Electric rocking furnace lined with a fireclay-silica sand patching mix over a mullite crock to produce acid lining characteristics. This thin (1/8 to 1/4-in.) coating slagged out after three or four heats, which was

RON

assocists,

949). o. 7.

New

d. t the

4.10

bars

4.14

have

rmal

aid I

n. I

are

fre-

lly I

here

eased

of its

ty of

the

ting

tant

as a

g to

ions

rop-

lica-

ther

e of

bute

grav

ence

ron-

nary

den-

tion

ore.

this

la.

^{*} Associate Professor, Metallurgical Engineering and Metal Processing, ** Instructor, Metal Processing, University of Michigan, Ann Arbor, Mich., *** Metallurgist, Development and Research Div.. International Nickel Co., Inc., New York.

Unetched

F. F

Six

mol

The

sepa

grai

cent

wer

cool

mov

wate

furr

ther

plac

of e

to be the peraund appropriate temping in the front hold har a tion, this and

dat wei

Etched

0

stan

the largest number made in any one production campaign.

The charge used consisted of scrap, pig iron, and ferro-alloys, proportioned as follows: 81.5 per cent scrap, 17.0 per cent pig iron. Balance nickel squares, ferro-70-chrome, and ferro-60-molybdenum.

A total of seven melts was required for the investigation. Chemical compositions are presented in Table

TABLE 1-CHEMICAL COMPOSITIONS OF ALLOYED IRON
MELTS INVESTIGATED

Melt	T. C.	Si	Mn	S	P	Ni	Cr	Mo
В	3.35	1.12	0.40	0.07	0.07	4.39	1.79	0.15
C	3.34	1.24	0.32	0.06	0.07	4.89	2.08	0.13
D	3.48	1.15	0.34	0.06	0.07	4.40	1.91	0.11
E	3.39	1.15	0.39	0.06	0.07	4.50	1.73	0.15
F	3.48	1.13	0.35	0.06	0.07	4.29	1.85	0.13
G	3.36	1.16	0.34	0.06	0.07	4.34	1.81	0.11
H	3.38	1.12	0.33	0.05	0.07	4.42	1.96	0.15
Ar	nalyses	by Det	roit T	esting	Labora	tory ar	d Mer	nber
	Labora							

At the beginning of each melting campaign the furnace was preheated thoroughly, in order to establish standard melting conditions for each melt, whether it was the first or a later melt in a succession made during any one day. In charging, the pig iron was placed on the bottom of the furnace, followed by the scrap. The nickel and ferro-alloys were added immediately after the entire base charge was molten.

The metal was melted, alloyed, then superheated to a maximum temperature of 2700 to 2740 F as determined by optical pyrometer measurement on the clear surface of the metal in the bath. Upon reaching the maximum temperature, each melt was tapped into a preheated fireclay-silica-lined ladle and poured into the molds. Pouring temperatures were 2550 F to 2575 F, as measured by an immersed noblemetal thermocouple.

Test Casting—The casting employed in this investigation was wedge shaped, with dimensions as follows: Minimum thickness 1/4 in.; Maximum thickness 11/2 in.; Width 21/2 in.; Height 2 in.

Fig. 1-Typical graphite patterns and dendrite sizes in tip and interior of experimental wedge casting.

Mag. 100×.

Interior at 1-in. Thickness

CON

the

es-

elt,

on

on

ved

led

en.

ted

as

on

on

was

nd

ere

ole-

es-

fol-

ess

Six such castings were produced in each open-top mold, with a connecting slab $2\frac{1}{2} \times 14\frac{1}{4} \times 1$ -in. thick. The slab was notched between castings to permit separation of the wedges immediately after shakeout.

The molds were baked cores made of A.F.S. 50 grain size lake sand bonded with 1 per cent oil, 1 per cent cereal, and 1 per cent bentonite. The cores were baked 16 hr at 375 F.

After pouring, the core molds were allowed to stand 7 min, during which the metal solidified and cooled to about 1900 F. The castings were then removed from the mold and broken apart.

One casting from each heat was air-cooled, one was water quenched, and the others were moved to electric furnaces which had previously been brought to equilibrium at the temperatures selected for isothermal transformation. Twelve specimens were placed in each furnace at the time that chromelalumel thermocouples in one or two of the wedges of each group indicated that the temperature of the interior at 1-in. thickness was 50 F above the isotherm to be investigated. The specimens were placed in the furnaces at this time, rather than when the temperature was at the isotherm, to prevent excessive undercooling of the tips. Cooling to 1000 F required approximately 16 min from the time of pouring.

Experience showed that the castings reached furnace temperature in approximately 15 min after placement in the furnace. Transformation time was calculated from this point. Specimens were removed from the holding furnaces after periods of from 15 min to 120 hr and water quenched in preparation for examination. A total of 15 isotherms were investigated in this work. The minimum temperature was 300 F, and the maximum 1300 F. Maximum temperature

range in the holding furnaces was plus and minu-10 F.

Testing Procedure—A standard procedure for sectioning the wedge castings was adopted to secure uniform samples for examination. All sectioning was done with a high speed abrasive cut-off wheel, using a type of disc which produced a minimum of heating during cutting. The first cut was made on a plane parallel to the top slab of the wedge at a location selected to provide metal at 1-in. wedge thickness. The heavier end of the wedge was discarded. A notch was made at the center of the wedge tip, and a full cut was made about ¾ in. from the center. The casting was fractured at the notch and ground for hardness tests, then polished for metallographic examination.

Rockwell "C" hardness determinations were made on the fractured and ground surface along the center line of the flat surface at two locations—near the tip and at a point of 1-in. wedge thickness. The entire surface was polished for metallographic examination.

Typical microstructures developed at the tip and in the interior of the wedge at 1-in. thickness are shown in Fig. 1.

Part A—Transformations on Continuous Air-Cooling to Isotherm

Voluminous data were accumulated during the investigation, covering several phases of the research. These will be discussed in individual sections.

Variations of Hardness with Isothermal Transformation—One of the standard procedures used to follow isothermal transformations of steels is the change in hardness with transformation, which provides reliable data, if used on homogeneous materials. It was thought that such tests might be used for this

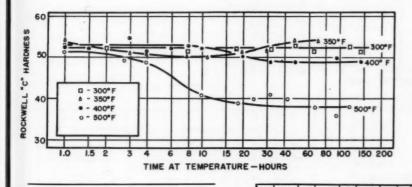


Fig. 2A-Time-Rockwell C hardness data for isothermally transformed wedge castings (interior at 1-in. thickness).

Fig. 2B-Time-Rockwell C hardness data for isothermally transformed wedge castings (interior at 1-in. thickness).

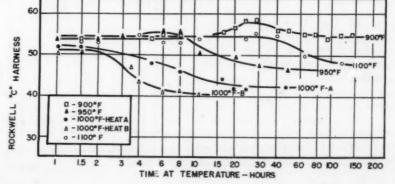


TABLE 2-HARDNESS OF WEDGE CASTINGS AFTER ISO-THERMAL TRANSFORMATIONS

(All Samples Air-Cooled to Indicated Temperatures)

	Isothermal Trans.		Percent		ip	1/2 in. thickness		in. cness
	Temp. F							
В	300	144	8	57	605	565	51.5	503
F	350	72	5	58	555	532	54	502
E	400	124	99	55	540	512	49	452
B	500	120	95	52	526	358	38	364
E	600	124	15	57	612	452	49.5	452
F	650	120	2	60.5	647	560	53	532
B	700	120	0	59	622	536	54	541
E	800	164	1	60	628	583	55	516
D	900	168	18	56.5	578	578	55	551
F	950	48	95	53.5	541	473	46.5	448
A	1000	441/	99	51	477	444	42.5	430
D	1100	120	45	54.5	532	466	48	490
E	1200	164	25	51.5	532	469	42.5	477
F	1250	120	5	58	630	555	53	550
B	1300	663/4	0	60.5	683	593	54	560
D	Air-Coole	ed		60.5	609	560	51	520
E	Air-Coole	ed		61	642	536	53	540
F	Air-Coole	ed		60	585	555	52	545

alloyed white iron, even with its heterogeneous structure. Accordingly, hardness data obtained on isothermally-transformed specimens from locations at the tip and at 1-in. thickness were plotted for each of the 15 isotherms investigated. Typical Time-Rockwell "C" hardness data for transformations at 300 to 500 F and 900 to 1100 F are plotted in Fig. 2a and 2b, respectively. Similar curves were obtained for all other isotherms.

ha

Th

pel

mı

an

no

at

for

hai

500

ten

tin

int

the han

Mi

for "C

in

the

for

Total

ter

as inc

Specimens continuously cooled in air had hardnesses in the interior at 1-in. thickness of 51 to 53 Rockwell "C," while specimens water quenched after shakeout had hardnesses of 50 to 51 Rockwell "C." The tip hardnesses were 60 to 61 Rockwell "C" on air-cooling and water quenching.

Little significant change in hardness occurred on holding at 300 F and 350 F. At 400 F the hardness decreased to 49 Rockwell "C" after 32 hr and remained constant up to 144 hr. At 500 F the hardness decreased continuously from 2.5 to 20 hr, then remained fairly constant at 38 Rockwell "C."

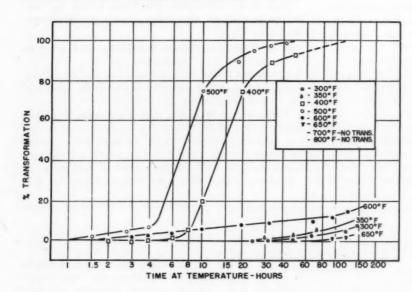
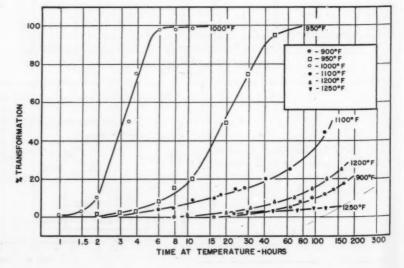


Fig. 3A-Time-Transformation data for isothermally transformed wedge castings (from metallographic structures of the interior at 1-in, thickness).

Fig. 3B-Time-Transformation data for isothermally transformed wedge castings (from metallographic structures of the interior at 1-in. thickness).



IRON

struc-

iso-

is at

each

Time-

ns at

g. 2a

iined

nard-

0 53

after

"C."

on on

l on

lness

re-

ness

re-

lata

dge

ruc-

ick-

At temperatures between 900 F and 1300 F the hardness changes with time were more consistent. The maximum hardness changes within this temperature range occurred at 1000 F. The hardness began to drop after 1.0 to 1.5 hr and reached a minimum of 40 Rockwell "C" in 10 to 20 hr.

Hardness changes began later at temperatures above and below 1000 F, and proceeded slower. Little or no change occurred at 1250 F and 1300 F.

The hardness data may be summarized as follows:

1. Minor changes occur, even with long soaking, at 300 F and 350 F, indicating either little transformation, or the formation of a constituent with a hardness approaching martensite.

2. Greater hardness changes occur at 400 F and 500 F, with indicated complete transformation at both temperatures. Minimum hardness at shortest soaking time was developed at 500 F, 38 Rockwell "C" in the interior and 51 Rockwell "C" at the tip in 20 hr.

3. The magnitude of the hardness changes decreased, as did the rate of change in the temperature range of 600 F to 900 F. At 700 F, 800 F, and 900 F there was an actual hardness increase.

4. In the temperature range 950 F to 1200 F the hardness decreased appreciably with time at temperature. The hardness change was most rapid at 1000 F and slower at 950 F, 1100 F and 1200 F. Minimum hardness was developed by complete transformation at 1000 F. This produced 40 Rockwell "C" in the interior and 48 at the tip.

5. At temperatures of 1250 F and 1300 F no change in hardness was evident in the 1-in. section, although there was a slight decrease in hardness in the tip after about 50 hr at 1250 F.

The hardness data from the isothermally transformed specimens established the rough shape of the T-T-T diagram for the alloyed iron, but the magnitude of the hardness changes with transformation was relatively small, due to the presence of large quantities of cemenite, which remained unchanged. This undoubtedly would have led to errors in the determination of beginning and end of transformation, as well as intermediate stages. Further, hardness increases at certain temperatures, 700, 800, and 900

TABLE 3-Austenite Transformation Data for Alloyed Iron

	AI	L	OYED IRON	
	(Fro	m	Plotted Data)	
(Castings	Air-Cooled t	0	Transformation	Temperatures)

	Time for	Transformat	ion, hr	
Temp. F	2%	10%	20%	98%
300	55	200	**	
350	40	135		**
400	7.0	9.0	10	100
500	1.5	4.6	5.2	33
600	2.7	40		
650	120			
700				
800				
900	25	100	185	
950	3	6.8	10.5	64
1000	1.3 -	1.95	2.2	6
1100	4	15	44	
1200	20 -	64	130	
1250	35			
1300				

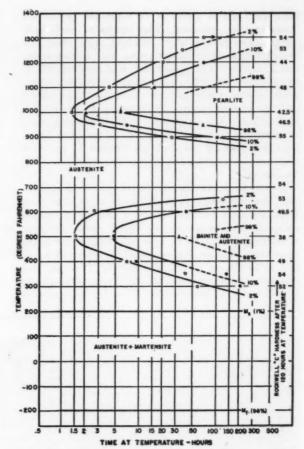


Fig. 4—T-T-T diagram for alloyed white iron with 3.40 T.C., 1.15 Si, 0.35 Mn, 4.40 Ni, 1.85 Cr and 0.12 Mo. Castings were air-cooled to transformation temperatures.

F, were unexplained. However, the hardness data did show the presence of a double "knee" in the T-T-T diagram, with minima at about 500 F and 1000 F. They also showed that essentially no transformation occurred in up to 120 hr at 300, 350, 700, 800, and 1300 F, which roughly outlines the shape of the diagram.

Brinell and Rockwell hardness of air-cooled and of isothermally transformed castings are summarized in Table 2.

Variation of Microstructure with Isothermal Transformation—Because the accuracy of the hardness data was not adequate to exactly position the T-T-T diagram, it was necessary to secure complete structural data by metallographic examination. The specimens used for hardness tests were reground and polished, etched with nital and examined microscopically. All specimens consisted of cementite, graphite and complex matrix constituents ranging from mixed martensite and austenite; martensite, bainite and austenite; pearlite and martensite; or pearlite.

The specimens which had been held at each isotherm were examined microscopically by three investigators, each of whom examined several areas on each specimen to arrive at an estimated percentage transformation product. The results of each individual examination were averaged with those obtained by the other investigators, and a final result was obtained. This procedure was necessitated by the complex carbide and graphite structures which destroyed the continuity of the matrix constituents and made measurement of areas or lineal analysis impractical. Approximately 1,000 specimens were examined to secure the data for the complete investigation. Part of the data are plotted as per cent transformation versus log time at temperature in Fig. 3a and 3b.

It will be noted that the structural data confirm the hardness data previously discussed, in that little transformation occurred at 300 F and 350 F, almost complete transformation at 400 F, and complete transformation at 500 F. Less than 20 per cent transformation product was obtained in over 120 hr at 600 F, and none at 700 and 800 F. The rate and amount of transformation increased with temperature in the range of 900 to 1000 F, with completion at

950 F and 1000 F. The rate then decreased with increase in temperature at 1100, 1200, and 1250 F. At 1300 F, only nucleation was observed in 6634 hr.

Data for the T-T-T diagram for the alloyed iron were taken from the time-transformation curves in Fig. 3a and 3b. The "beginning" of transformation was taken as 2 per cent product, and "end" as 98 per cent product. Data are summarized in Table 3, which shows the time (hours) required for 2, 10, 20, and 98 per cent transformation. These points are plotted in Fig. 4, a partial T-T-T diagram for this iron.

Fig. 4 shows the double "knee" type of T-T-T diagram, with minimum transformation times at 1000 F. There is no transformation in 120 hr at temperatures between 650 F and 850 F. The upper limit of the diagram appears to be about 1300 F, at which temperature there was only an indication of nucleation in 663/4 hr.

M. - Mr Determinations

 $M_s - M_t$ data were secured by a standard cooling-transformation-tempering procedure. A series of

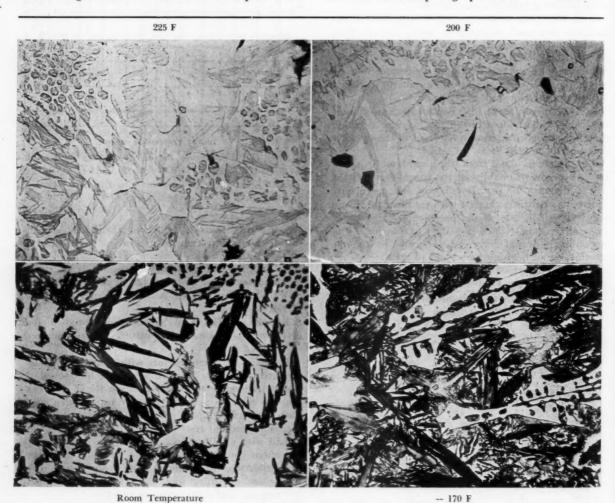


Fig. 5-Microstructures of castings (interior at 1-in. thickness) used for M_8 - M_f determinations. Tempera-

interior at 1-in. ture indicated is the one to which casting was cooled tions. Tempera-before tempering at 850 F. Etchant-1% Zephiran Chloride in 4% Nital. Mag. 500×.

th T te: at

ici

ur

co

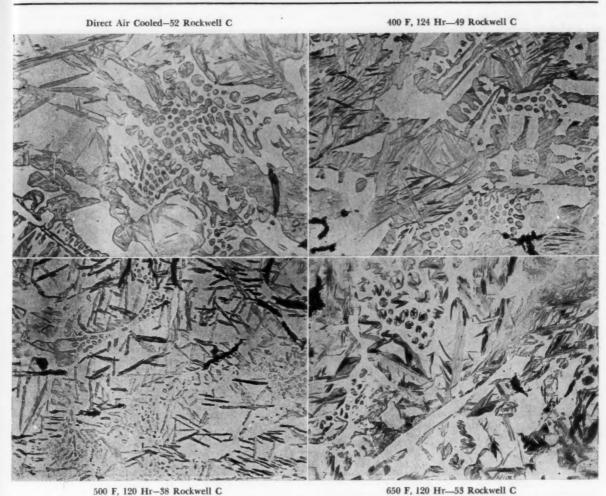


Fig. 6A – Matrix microstructures produced on long soaking for isothermal transformation; interior of casting at 1-in. thickness. Etched. Mag. 500×.

wedge molds was made up with chromel-alumel thermocouples in the wedge cavities at 1-in. thickness. These were poured, shaken out, and air-cooled to temperatures between 400 F and room temperature at 25 F intervals. Other samples were air cooled to 300 F, then quenched to 32 F, 0 F, -50, -100 F, -150 F, -200 F, or -225 F in ice water, alcohol-dry ice, or liquid air baths. Upon reaching the selected temperature between 400 F and -225 F the specimen was placed in a furnace at 850 F, held 15 min, and water quenched. All specimens were sectioned, using the procedure described previously, hardness was determined, and the specimens were examined microscopically.

The etchant used for preparation of the M_s-M_f specimens was 4 per cent nital with 1 per cent zephiran chloride, which darkens martensite formed before and tempered by reheating to 850 F and leaves untempered martensite formed on subsequent quenching relatively lightly etched. With this structural contrast it is possible to distinguish between the martensite formed on air-cooling and that formed on

quenching after reheating to 850 F, and thus to determine the highest temperature of formation of martensite on continuous cooling. Photographs of representative specimens so treated are presented in Fig. 5. No dark martensite appears in the sample reheated to 850 F after cooling from melt to 225 F. One dark needle appears in the sample cooled to 200 F before reheating. This is the M_s temperature. More darkened needles appeared in samples cooled to 175 F and 150 F. All martensite in samples cooled to room temperature or below was darkened by subsequent tempering and etching.

It is obvious that the amount of martensite increases with decreasing temperature to -170 F, which indicates that the $M_{\rm f}$ temperature is below room temperature, and probably near -200 F. This was deduced from the structures shown in Fig. 5 and from results of hardness tests which showed increasing hardness with decreasing temperature to -225 F. Therefore, the T-T-T diagram of Fig. 4 was completed with $M_{\rm s}$ at 200 F (fairly accurate) and $M_{\rm f}$ at -200 F (approximate).

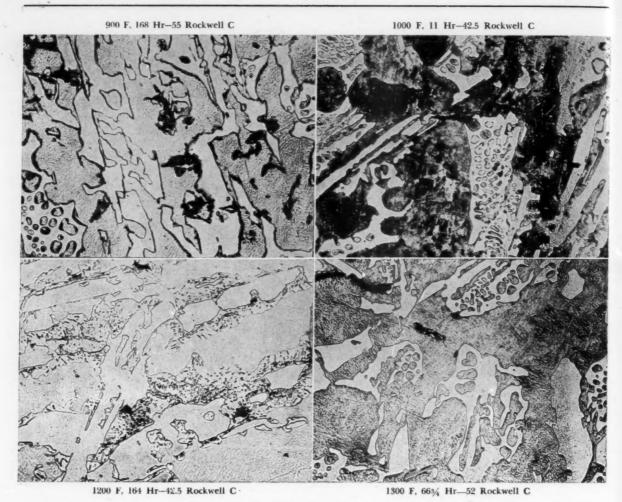


Fig. 6B-Matrix microstructures produced on long soaking for isothermal transformation; interior of casting at 1-in. thickness. Etched. Mag. 500×.

Microstructures Developed by Isothermal Transformation—The microstructure developed by continuous air-cooling of the wedge casting is demonstrated in the upper left photograph in Fig. 6. This consisted of coarse martensite, retained austenite, cementite, and graphite, and had a Rockwell "C" hardness of 51 to 53. Photographs of the microstructures developed by soaking for the longest period of time at some of the isotherms investigated are also presented in Fig. 6. Certain of the specimens were intentionally under-etched to accentuate the structure of the transformation product and to subdue the confusing background structure.

The constituents produced at 300, 350, and 400 F were characteristic thin needle-like particles. No positive identification could be made of the nucleating particles, although it might be assumed that the constituent is a true bainite, therefore nucleated by ferrite. The specimen transformed at 400 F shows the true structure of the low temperature product. At this temperature virtually complete transformation to the needle-like product was secured.

Transformations to the various types of bainite failed to reach completion, insofar as complete elimination of austenite was concerned. After nucleation had occurred, bainite formed at the expense of martensite, such that in all specimens the total of bainite and martensite was fairly constant, up to the time period required to provide complete replacement of martensite by bainite in the structure. Longer soaking provided further formation of bainite, but at a slow rate, until a stable structure of bainite and austenite was secured. At temperatures where the stable structure occurred in less than 120 hr, it was assumed that the transformation had reached completion, and lesser degrees of transformation were estimated, using the stable structure as a basis. At other temperatures, where transformation continued beyond 120 to 150 hr at temperature, estimation of the degree of transformation was based on stable structures secured at higher or lower temperatures.

Transformation at 500 F produced a more complex structure, consisting of coarser needle-like particles and clear, open particles. At 600 F partial transformation to the open type bainite structure was secured. None of the needlelike type was observed.

Scattered particles of bainite similar to that obtained at 600 F were found in a wedge specimen soaked 120 hr at 650 F. Shorter time periods produced no evidence of transformation. The martensite associated with this constituent was coarse, and there was considerable retained austenite. The appearance of the martensite produced from austenite held at 650 F, and at lower temperatures, indicated no change in the austenite which could be reflected in alteration of the martensite produced on cooling to room temperature.

Soaking at 700 F produced no transformation of austenite at temperature. However, the martensitic structure obtained by quenching to room temperature was markedly different than that produced in samples discussed above. This martensite was finer and denser than in the previous samples, and the hardness of the quenched specimen (120 hr at 700 F) was higher than that developed in air-cooled or quenched specimens. This indicates, as does the structure, much more complete transformation to martensite than in direct air-cooled samples.

The martensite developed in quenched castings which had been soaked at temperatures above 700 F and up to 1300 F was likewise very fine and dense, and harder than that produced by air-cooling. In 164 hr at 800 F about 2 per cent pearlite was developed at the austenite-cementite boundaries. However, the hardness of the specimen was 55 Rockwell "C," indicating practically complete transformation to martensite of the austenite which had not transformed to pearlite. A specimen cooled to -225 F developed the same hardness.

Pearlitic structures were developed on soaking at temperatures between 900 F and 1300 F. Complete transformation was obtained at 950 F and 1000 F, and only partial transformation at the other temperatures. Between 1100 F and 1300 F, a spheroidized structure was produced, while at temperatures between 800 F and 1000 F, fine pearlite, unresolvable at 500X was produced. In general, nucleation of the pearlite occurred at the cementite-austenite boundaries, but at 1250 F, where very little pearlite was formed, even in 120 hr, nucleation occurred at or near graphite flakes. Pearlite produced at this temperature was spheroidized.

The results of hardness tests and structural examination of isothermally transformed castings indicated that slow rates of cooling or arrested cooling might have rather marked effects on the transformation characteristics of alloyed irons. Castings soaked for up to 120 hr at temperatures between 700 F and 1300 F underwent more complete transformation to martensite, with attendant higher hardnesses. This is apparently due to "conditioning" of the austenite as a result of carbide rejection at the elevated temperatures, producing a condition more nearly equilibrium. This results in higher M₈ and M_t temperatures on subsequent cooling to room temperature and, therefore, more complete transformation to martensite.

Because of those findings, a second phase of the investigation was begun to determine the influence of soaking at elevated temperatures and of slow cooling on the isothermal transformation characteristics and $M_s - M_t$ temperatures of the alloyed iron. This phase had dual objects: to investigate the possibilities of improving the hardness of highly alloyed irons by elimination of retained austenite, and to investigate

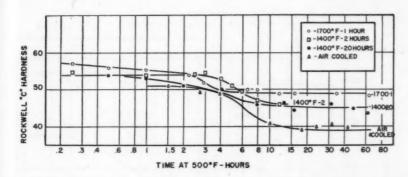
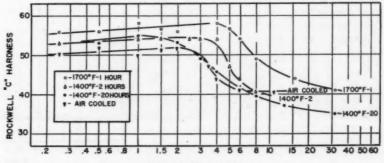


Fig. 7A-Time-Rockwell C hardness data for wedge castings cooled under varying cycles and transformed isothermally at 500 F. (Interior at 1-in. thickness).

Fig. 7B-Time-Rockwell C hardness data for wedge castings cooled under varying cycles and transformed isothermally at 1000 F. (Interior at 1-in. thickness).



TIME AT 1000° - HOURS

gate the transformation characteristics of castings cooled in a manner more nearly approximating equilibrium.

Part B — Influence of Austenite "Conditioning" on Transformations

As stated above, important possibilities of "conditioning" the austenite in alloyed castings prior to transformation were indicated by the results obtained in the first phases of this research and, accordingly, further work on this subject was initiated. Standard melting, casting and shake-out procedures were used in producing specimens for this work, but after shake-out, variations from the earlier air-cooling procedure were employed. Three conditioning procedures were employed. These were:

1. Hold 1 hr at 1700 F and cool in the furnace at a rate of about 60 to 70 F per hr to transformation temperatures. This procedure is designated 1700-1.

2. Hold 2 hr at 1400 F and air-cool to transforma-

tion temperatures. This is designated 1400-2.

3. Hold 20 hr at 1400 F and air-cool to transformation temperatures. This is designated 1400-20.

Castings handled under each of the three procedures were transformed isothermally at 500 F and 1000 F. Others, with integrally cast chromel-alumel thermocouples were used for M_n-M_f temperature determination.

Influence of Conditioning on Hardness-Time Relations—Standard sectioning and testing procedures were employed in examining wedge castings produced under the modified conditions detailed above. Hardnesses were taken at the tip and in the interior at 1-in. thickness. Time-Rockwell "C" hardness data for interiors at 1-in. thickness are plotted in Fig. 7a and 7b.

In general, the hardnesses of all specimens subjected to partial isothermal transformations after conditioning heat treatments were 3 to 8 points Rockwel!

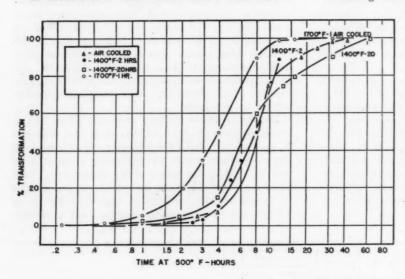
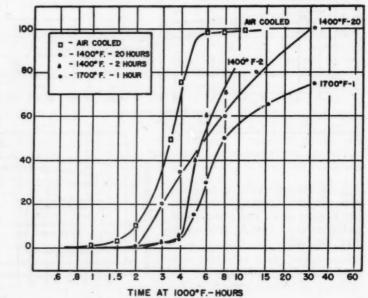


Fig. 8A — Time - Transformation data for wedge castings cooled under varying cycles and transformed isothermally at 500 F. (From metallographic structures at 1-in. thickness).

Fig. 8B — Time - Transformation data for wedge castings cooled under varying cycles and transformed isothermally at 1000 F. (From metallographic structures at 1-in, thickness).



ON

na-

ed.

nd

nel

de-

la-

ere

m-

rd-

1.

or

ad

b

1

"C" above those produced by directly air-cooling to the transformation temperatures. This is undoubtedly due to more complete transformation of austenite to martensite on subsequent quenching. In the interior sections the hardness tests gave indications of earlier nucleation and completion of transformation at 500 F after conditioning treatments, particularly by slow cooling from 1700 F, than after directly aircooling to 500 F for transformation. This is probably due to the presence of graphite in these sections which permated lower austenite carbon content on conditioning than on air-cooling. Little effect of conditioning was noted in the hardness changes in the tip sections, probably because of the practical absence of graphite, which retarded attainment of equilibrium conditions.

An important effect of conditions was noted in the hardness-time curves for specimens transformed at 1000 F. These showed considerably longer nucleation times and slower transformation rates in conditioned specimens.

Influence of Conditioning on Isothermal Transformations-Microscopic examination of the isothermallytransformed specimens produced in this phase of the investigation confirmed the results of the hardness tests. Time-Transformation data for 500 F and 1000 F isotherms are plotted in Fig. 8a and 8b.

At 500 F the specimens conditioned by soaking 1 hr at 1700 F and furnace cooling underwent transformation considerably more rapidly than air-cooled specimens or those conditioned at 1400 F. The three latter conditions produced essentially the same transformation characteristics at 500 F. There was an indication of somewhat more rapid attainment of completion in the specimens conditioned 2 hr at 1400 F than in those directly air-cooled to 500 F, while specimens conditioned 20 hr at 1400 F started transformation earlier, but proceeded slower than directly aircooled specimens.

At 1000 F the marked effect of condition of austenite was more apparent. Here all conditioned specimens proved less prone to transform to pearlite, with

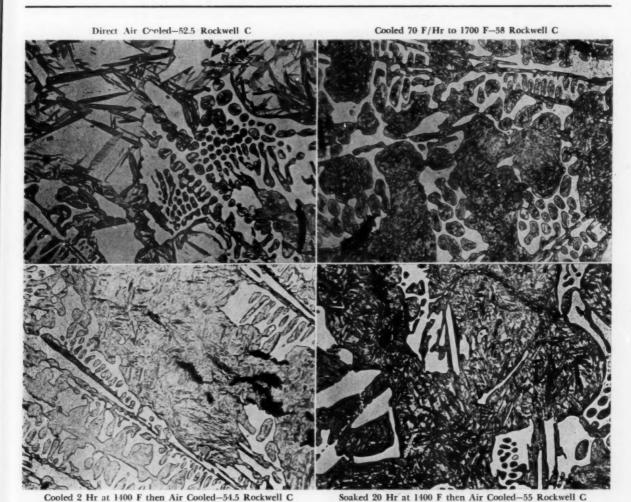


Fig. 9-Influence of conditioning on microstructure secured by continuous cooling; interior at 1-in. thickness. Etched. Mag. 500×.

those conditioned by slow cooling from 1700 F proving the most stable at 1000 F.

Influence of Conditioning on Martensite Transformation-The results of isothermal transformations at temperatures of 700 F and above indicated a conditioning effect on the austenite in alloyed iron which permitted complete (or nearly complete) transformation to martensite on cooling to room temperature. These results were checked by conducting Mo-Mf temperature determinations on conditioned specimens. The standard cool-temper-quench procedure discussed previously was employed for this purpose. Conditioning at 1400 F raised the Ms temperature from 200 F in air-cooled specimens to 400 F. Slow cooling from 1700 F raised the M_s to above 400 F, as evidenced by the presence of 10 per cent dark martensite in the specimen cooled to 400 F, reheated 15 min at 850 F, and water quenched. The fact that the M, temperatures were raised by conditioning is evidenced by the greater degree of martensite transformation at

any temperature in conditioned samples, as compared with air-cooled samples.

11112

die

du

the

in

wa

as

ne

ves

of

tio

all

dia

10

ob

tio

tio wh slo (1 eff du

cen

is an ing for ter

isti

shi wii pro bai

sol

rec

ten

Influence of Conditioning on Microstructures—Microscopic examination of specimens subjected to conditioning treatments prior to transformation disclosed many alterations of structure as a result of the conditioning. Photographs showing the effect on the martensite transformation are shown in Fig. 9. While the directly air-cooled specimen had coarse martensite and considerable retained austenite, the others had much finer and denser martensite and evidence of little or no retained austenite. The martensite was finest in the specimen from 1700-1, which had the highest M temperature and showed most rapid martensite transformation with decreasing temperature below 400 F.

The isothermal product produced at 500 F was altered appreciably by conditioning by slow cooling from 1700 F. This is shown in Fig. 10. The bainite produced at 500 F was coarser than in specimens directly air-cooled to 500 F, and the degree of transformation.

Heat E-Air Cooled to 500 F (5% Transformation, 49 Rockwell C)

Heat H—Cooled 75 F/Hr to 500 F (50% Transformation, 49 Rockwell C)



Heat H-Soaked 2 Hr at 1400 F, Air Cooled to 500 F (5% Transformation, 53 Rockwell C) Heat H-Soaked 20 Hr at 1400 F, Air Cooled to 500 F (8% Transformation, 50 Rockwell C)

Fig. 10—Influence of conditioning on isothermal transformation in 4 hr at 500 F; interior at 1-in. thickness. Etched. Mag. 500×.

ed

li-

ed

li-

11 -

10

le!

th

DI

in

g

mation in 4 hr was much greater. This was also indicated by hardness tests and micro examination of a long series of specimens. Conditioning at 1400 F produced much less alteration in structure, except for the generally finer martensite produced on quenching. The degree of transformation in 4 hr at 500 F was about the same in castings conditioned at 1400 F as in air-cooled specimens. This was shown by hardness test and in time-transformation curves.

Discussion and Conclusions

Although the most important findings of this investigation have been discussed in previous sections of this paper, some merit repetition and amplification.

The isothermal transformation characteristics of the alloyed iron are such that a double "knee" T-T-T diagram is produced. This diagram has one knee at 1000 F and the other at about 500 F.

The T-T-T diagram presented in this paper was obtained by examination of specimens rapidly aircooled to the temperatures selected for transformation. It is directly applicable to small castings, but requires alteration for large castings which cool slowly. The supplementary data secured in the investigation of conditioning treatments provide data upon which such alteration can be based. Conditioning by slow cooling or by soaking at an elevated temperature (1400 F in this investigation) has a rather marked effect on the transformation characteristics and produces the following effects:

1. It moves the pearlite knee of the T-T-T diagram to the right. This apparently is due to rejection of cementite during conditioning, which reduces the nucleating tendency for pearlite.

2. It moves the bainite knee of the T-T-T diagram to the left. This appears also to be due to the lowered carbon content of conditioned austenite which can form ferrite nuclei more readily than higher carbon, unconditioned austenite.

3. It raises the M_n and M_t temperatures. This also is due to the lowered carbon content of the austenite and results in considerably higher hardnesses on cooling to room temperature. Conditioned austenite transforms almost completely (if not completely) to martensite above room temperature.

4. It changes the appearance of bainite transformation product at 500 F. Rapidly cooled austenite transforms to a mixture of fine needle-like and open, feathery bainite, while conditioned austenite produces a coarser needle-like constituent at this temperature.

Because of the change in transformation characteristics as a consequence of conditioning of austenite, it might be concluded that conditioning also will shift the position of the knee of the T-T-T diagram with respect to temperature. The pearlite knee is probably shifted only slightly, if at all, while the bainite knee is probably raised slightly.

The process of conditioning austenite following solidification of alloyed castings offers a method of reducing the amount of austenite retained at room temperature. Some reduction of the quantity retained might be effected by the simple process of slow cooling after early shakeout, such as by burying castings in large groups to retard cooling or placing them in preheated furnaces.

Acknowledgment

The writers wish to thank the International Nickel Co., Inc., for financial support of the investigation upon which this paper is based and for permission to publish the data obtained under its sponsorship. Special thanks are extended to Mr. H. V. Beasley of the Pittsburgh Field Office for advice and support during the project.

The Technical and Research Committee of the Roll Manufacturers Institute advised the use of the procedures which were followed in this investigation, and their advisory contributions to the progress of the work are also recognized.

The assistance of W. H. Betts, John F. Carlson, George C. Evans, and Yeshwant P. Telang, students in the Department of Metallurgical Engineering, University of Michigan, was of material value in securing the large amount of data required in this investigation.

DISCUSSION

Chairman: R. Schneidewind, University of Michigan, Ann Arbor.

Co-Chairman: V. A. Crosby, Climax Molybdenum Co., Detroit. E. A. Loria (Written Discussion): Again-it should be emphasized that this is the first paper that presents a T-T-T diagram on alloyed white iron directly air-cooled to the transformation temperatures. Its importance as a guide in the processing of commercial castings is noteworthy and it will no doubt be used to advantage by the roll manufacturing industry.

From the academic viewpoint, several questions arise which are probably covered in the author's report to the Roll Manufacturer's Institute, but omitted in the abbreviated paper. First, in view of the known inadequacy of the present nomenclature for the decomposition products of austenite and the variations in microstructures observed (as in complex alloy steels), it is necessary to resort to micrographs in order to classify some of the microstructures.

Since this is the most comprehensive study of the isothermal transformation characteristics on direct cooling of alloyed white iron, it would be worthwhile to show the microstructures after 2 to 10 per cent transformation at 400 F, 350 F and 300 F. The thin plate-like structures occurring in this temperature range are very interesting for they may undergo changes in structure on isothermal holding. In high magnification micrographs of a complex alloy, hypereutectoid steel, the writer* showed that the carbon-rich plates (acicular carbide) decomposed or tempered slowly to a carbon-rich aggregate after long holding periods at temperature. Did the authors observe any evidence of this reaction in their micro-examination of the plate-like structures found in the lower part of the bainite knee (both before and after austenite conditioning treatments)? The comment is made that conditioning treatments shortened the start time and changed the appearance of the bainite transformation, particularly the plate-like type. More of the excellent micrographs (Fig. 10) especially at higher magnification would substantiate such con-

The statement is made that the needle-like bainite is harder than the more open, feathery bainite (which is the usual case), but the Rockwell C hardness values listed in Fig. 4 are practically the same at the top of the bainite knee as at the bottom. Would the authors comment on any evidence of a two-stage reaction occurring only in the upper part of the bainite knee

¹ Research and Development Dept., The Carborundum Co., Niagara Fails, N. V.

N. Y.

*E. A. Loria and H. D. Shephard, "Acicular Transformations in Alloy Steel," Transactions, ASM, vol. 40, 758-774 (1948).

which has been observed in the case of alloy steels with double

knee, T-T-diagrams by Troiano and others?

It is of practical value to know that conditioning treatments alter the position of the T-T-T-curves with respect to time. As in steel, probably the most important information relevant to the problem of slack quenching is the time for the initial decomposition of austenite at any temperature and the nature of the decomposition products. Just how much of a shift to longer times is there in the start curve for pearlite and to shorter times in the start curve for bainite formation? Again, as in the case of alloy steels exhibiting a double knee, T-T-T diagram, alloying elements (when dissolved in austenite) have their major effect in retarding the pearlite transformation but appear to have little effect on the bainite reaction. Actually, the fact that the bainite shelf in this iron is moved to the left by conditioning treatments would make the T-T-T-diagram similar to the T-T-T-diagrams shown for such steels.** Another question which arises is, would incubation treatments of the type proposed by Jaffet and the writertt have any effect on this iron when first given an elevated temperature conditioning treatment so that the two knees would then be out of balance with respect

There is a possibility that the austenite conditioning (soaking) treatment may have modified the transformation characteristics of the iron (as in steel), i.e., cause a shift of the T-T-T-curves towards shorter induction periods for bainitic transformation, as the holding temperature (in the austenitic range) is raised. If such a shift does, in fact, occur the knee of the transformation curve for continuous cooling of a large roll casting might well be intercepted giving a transformation to an intermediate product, whereas the knee of the curve for the unconditioned small, wedge casting would not be intercepted and martensite would be produced.

The small wedge castings used in the construction of the T-T-T-diagram were air-cooled to the transformation temperatures. The rate of decomposition of austenite is accelerated by burning in air.+++ The state of the impurities in the iron (particularly sulphides) is different during direct cooling transformation in comparison with transformation following reheating and such solubility effects may produce slight differences in

the location of the start curves.

Finally, a number of roll manufacturers do their melting for similar alloyed irons in an air furnace rather than an electric furnace, and there is an effect of the melting medium (principally superheating) on the type and amount of graphite that is produced in the iron. The authors have made an excellent start on plotting the course of the primary austenite transformation. Determining the course of the retained austenite transformation on reheating will be a more difficult problem which may or may not be amenable to metallographic interpretation. Retained austenite does exist in significant quantities in their iron and does affect the properties. In the case of steel, new unpublished work indicates that if austenite decomposes during cooling, tensile ductility and impact resistance will be decreased; if austenite is retained, it will decrease yield strength; if it transforms during plastic deformation, it will increase tensile strength, lower elongation and reduction of area and reduce notch toughness.

MR. ROTE (Reply to Written Discussion by E. A. Loria): The writers appreciate the comments of Mr. Loria who has contributed considerable to the literature on the subject covered in

The holding time at the lower temperatures, 300 F to 400 F, were restricted to a maximum of about 120 hr. In this time

period transformation to the acicular structures started and proceeded toward completion, depending upon the temperature of holding. The transformation products were thin and needlelike, but did not undergo any observable change in structure with increasing time. Reference to Fig. 4 will show that there was little transformation in 120 hr at 300 F and 350 F, and the constituent obtained at 400 F was mixed with a large amount of relatively stable austenite.

The microstituents secured at lower temperatures in conditioned specimens were coarser and more feathery than in direct cooled specimens. This is demonstrated by comparison of the lower left photograph in Fig. 6a with the upper right photograph in Fig. 10. The structure of a direct cooled specimen held 120 hr at 500 F is shown in the former figure, while a conditioned specimen held only 4 hr at 500 F is shown in the latter. The direct cooled casting transformed to a relatively open structure, while the conditioned specimen transformed more

rapidly to a coarser, closed structure.

The Rockwell "C" hardnesses shown in Fig. 4 (right) are for specimens held 120 hr at the indicated temperatures. As was stated in the paper, hardness measurements were not reliable indications of transformation, particularly at the lower "knee" temperatures. At these temperatures the transformation products were bainite, but considerable austenite was retained. In many specimens for which hardness are reported considerable martensite was also present. Thus, at the upper and lower limits of the bainite knee, similar hardnesses were obtained, because little or no transformation occurred and the structure consisted of martensite, austenite and cementite. Lower hardnesses were obtained at intermediate temperatures due to substitution of bainite for martensite in the ultimate structures.

No evidence of a two-state reaction was observed at the upper bainite temperatures, probably because of the sluggish nature of the transformation in the highly alloyed iron, and the relatively

short soaking times utilized.

The degree of shift due to conditioning in the pearlite and bainite start times can be read from Fig. 8A and 8B. These show the marked shift in the time-transformation curves as a result of conditioning, particularly by slow cooling from 1700 F. The data were not reported quantitatively, since they represent only spot" determination.

18

n

n

th

m

It would appear that our conditioning treatments approximate the "incubation" treatments proposed by Jaffee and Mr. Loria.

MR. LORIA: Would the authors comment on the difference in the start of transformation curves when working with larger test samples? What effect would mass have on direct cooling transformation? The statement is made that the results are directly applicable to small castings but require alteration for large castings which cool slowly. How much shift in the start curves would there be in a commercial Ni-hard casting? In the case of (reheat) austenite transformation in certain, complex alloy steels, it has been mentioned that some times the bainite transformation in samples taken from large sections occurs more rapidly than in samples taken from a small casting.

Mr. Rote: An attempt to evaluate the influence of mass on the transformation characteristics of the alloyed white iron was made when wedge castings were cooled slowly from 1700 F. It was felt that the conditions utilized, sand cast small section and control cooling rate at 60 F per hr from 1700 F, approximated the conditions encountered in large chilled castings. These solidify rapidly against the chillers, then cool slowly due to the retarding effect of the large mass of the interior. Under these conditions, the pearlite transformation is retarded in massive section and the bainite transformation is accelerated, as compared with smaller sections.

The comparative transformation characteristics of reheated versus direct cooled specimens were not determined. However, it is likely that the results secured in slowly cooled (conditioned) specimens would approximate reheated specimens, since the slow cooling rate from 1700 F allowed an approach to equilibrium, such as would be achieved on reheating.

^{**} E. A. Loria, "Kinetics of the Austenite Transformation in Certain Alloy Steels," Transactions, ASM, vol. 43, pp. 718-738 (1951).

[†] L. D. Jaffe, "Anisothermal Formation of Bainite and Proeutectoid Constituents in Steels," Transactions, AIMME, vol. 176, pp. 363-383 (1948).

^{††} E. A. Loria, Discussion of above reference, p. 377-379. ††† F. C. Thompson and L. R. Stanton, "The Effect of Oxygen on the

Isothermal Transformations of Steel," Journal, Iron and Steel Institute, vol. 153, pp. 259-263 (1946).

EFFECT OF SAND GRAIN DISTRIBUTION ON CASTING FINISH

By

H. H. Fairfield* and James MacConachie**

ABSTRACT

Six different types of sand grain distribution patterns were produced by blending screened silica sands. Bronze test castings were cast in these sands. A precision laboratory molding machine was used. Molds were made with different degrees of ramming.

The surface smoothness of the castings was measured using Hobman's method; that is, profile measurements were made with a surface gage using a needle-tipped dial indicator.

Test results indicated the effect of average grain size and the effect of permeability on casting smoothness. The true flowability of the different sands was also indicated.

Introduction

IN ORDER TO CONDUCT RESEARCH on casting finish, it is necessary to use special molding techniques, and to have a method of measuring surface smoothness. The writers¹ described a precision molding machine and its accessories. With this machine test molds could be produced free from the effects of variables encountered in ordinary molding operations. The test casting used in this work is shown in Fig. 1.

The smoothness values reported in this paper were obtained by using Hobman's method for measuring casting surface smoothness. A needle-tipped dial indicator is used to measure the hills and hollows of the surface. A set of 25 evenly spaced test readings is made. The standard deviation of these readings gives a good index of the casting finish.

A set of washed and dried sands was obtained from a sand producer. These sands ranged from coarse to fine. All sands were of the same composition and general shape. Mixtures were compounded from these samples, 5 per cent Western bentonite was added for bond, water was adjusted to temper. Thus all variables except grain distribution were controlled.

Materials

The special samples of graded sands used in this work were supplied by R. A. Woods.† The samples were compounded into six different mixtures as shown in Table 1.

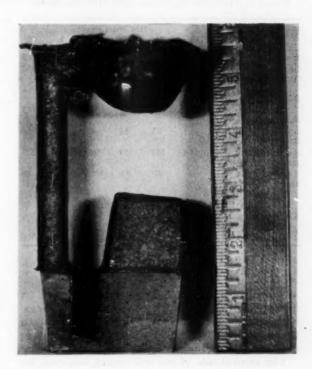


Fig. 1-Test casting as removed from the mold.

TABLE 1-SAND MIXTURES

Type of Sand			1	Mix Nun	bers			
			1	2	3	4	5	6
No.	25 W	& D				20	6	
No.	40 W	& D				20	6	78
No.	57 W	& D		50	331/2	20	12	
No.	70 W	& D	50	50	331/2	20	18	11
No.	100 W	& D	50		, ,			
No.	111 W	& D			331/3	20	58.0	11

Each sand mixture contained 5 per cent Western bentonite, and was tempered to optimum temper as judged by "feel." The properties of the sand mixtures are given in Tables 2 and 3.

Chief Metallurgist and ** Sand Supervisor, respectively,
 William Kennedy & Sons, Owen Sound, Ontario, Canada.

⁺ Sales Engineer, Geo. F. Pettinos Co. of Canada Ltd.

TABLE 2-SCREEN TESTS

U.S.			Mix N	Numbers		
Screen No.	1	2	3	4	5	6
20		**	0.1	0.6	0.5	0.1
30	0.1	0.4	1.5	5.6	2.6	1.3
40	0.1	4.2	1.9	13.3	5.2	26.9
50	0.6	11.0	10.0	14.1	6.9	27.0
70	14.6	18.4	23.8	16.2	12.9	14.4
100	61.5	18.8	28.2	18.2	19.4	10.0
140	15.5	21.2	16.7	13.8	22.6	8.0
200	1.6	15.2	9.6	9.6	17.1	5.1
270	0.7	3.8	1.2	1.0	4.0	1.2
Pan	0.6	1.4	0.9	0.9	2.0	0.6
A.F.S. Clay	4.8	5.6	6.0	6.6	6.8	5.0

TABLE 3-MOLDING PROPERTIES AND HOT STRENGTH

Property	Mix Numbers					
• •	1	2	3	4	5	6
Moisture, %	2.8	2.6	2.5	2.7	3.4	2.3
Green permeability	126	80	90	102	75	163
Green compressive						
strength, psi	5.3	8.6	8.1	8.4	11.0	8.1
Green deformation						
in./in.	.034	.046	.027	.022	.016	.0145
Flowability,						
Dietert No.	83	83	82	80	82	80
Weight of A.F.S.						
specimen, gm	150	155	154	162	158	163
Dry compressive						
strength, psi	60	75	83	93.5	117	86
Hot compressive strength, 500 F,						
psi (12 min)	55	75	73	110	122	82
Hot compressive strength, 1000 F,						
psi (12 min)	64	90	83	122	137	120

Equipment

A hand-operated molding machine was used. A metal pattern was used for the drag mold for the casting shown on Fig. 1 Figure 2 shows the flask filling device mounted over the drag mold. This equipment has been described in detail in A.F.S. Transactions.¹

The vertical side of the test casting was used for surface measurement. A dial indicator with a needle

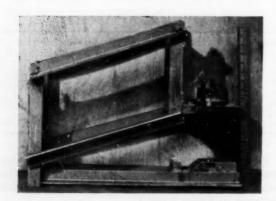


Fig. 2—Hand-operated jolt molding machine. Drag pattern is mounted on plate at right.

point was used to record the profile of the casting surface.

Method

Experimental procedure was as follows:-

- 1. Clean off pattern.
- 2. Position molding flask.
- 3. Position flask extension.
- 4. Fill flask by rubbing sand through sieve.

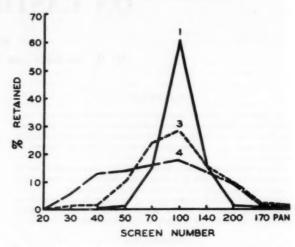


Fig. 3-Sand grain size distribution patterns for Mix No. 1, 3 and 4.

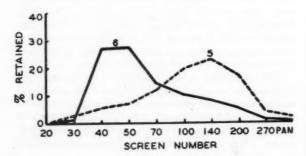


Fig. 4-Sand grain size distribution patterns for Mix No. 5 and 6.

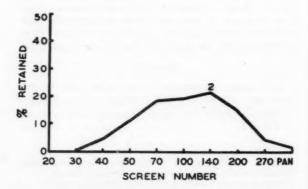


Fig. 5-Sand grain size distribution pattern for Mix No. 2.

- 5. Jolt table required number of times.
- 6. Remove flask extension.
- 7. Strike off, place bottom board.
- 8. Lift off mold, roll over.
- 9. Set gate core.
- 10. Put cope on.
- 11. Pour casting with 85-5-5 bronze at 2100 F.
- 12. Shake out, clean with a wire brush, cut off gate and head.
- 13. Measure profile of surface with needle tipped dial indicator.
 - 14. Determine standard deviation of 25 readings.
- 15. Express finish as standard deviation in units of $\frac{1}{1000}$ in.

Test Results

The six sand mixtures used in this report were tested by A. E. Murton at the Canadian Bureau of Mines. Test results are given in Tables 2 and 3. Results of surface finish readings are shown in Table 4.

Table 4—Surface Smoothness, Standard Deviation from True Surface in Units of 0.001 in.

		Mi	x Nun	bers		
Number of jolts	1	2	3	4	5	6
8	5.6	4.45	4.1	9.7	2.05	12.6
16	3.05	1.75	2.5	3.6	2.3	14.06
24	2.3	2.04	1.9	1.8	2.8	13.7
32	2.5	2.4	2.5	2.2	1.6	7.0

Discussion

Figures 3, 4 and 5 show the distribution patterns of the six sands. Sands 1, 3 and 4 show peaked, normal, and broad distribution. Sands 6 and 5 show right and left-handed distributions (nonsymmetrical).

Figure 6 shows the relationship between permeability and the smoothness of castings jolted 8 times. The effect of grain size on finish is shown in Fig. 7. It is apparent that fine sand grains and low permeability coincide with smooth finish.

It is assumed that true flowability of a sand is a measure of the relative amount of work necessary to compact the sand to a point where optimum finish is obtained. Table 5 and Fig. 8 show the true flowability of the six sands used. Sand No. 5 had the best flowability because it gave a smooth finish with only

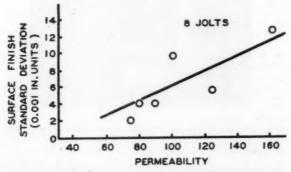


Fig. 6-Curve showing relationship between permeability and casting finish.

TABLE 5

Sand No.	Number of jolts required to get optimum smoothness	A.F.S. Flowability value
1	24	83
2	16	83
3	24	82
4	24	. 80
5	8	82
6	32	80

Table 6-Method for Calculating Surface Smoothness

Values in left hand Measurement	column are dial indicator reading Measurement
9.4	88.36
8.1	65.61
7.3	53.29
9.6	92.16
8.3	68.89
9.0	81.00
9.2	84.64
9.0	81.00
9.1	82.81
7.5	56.25
13.7	187.69
10.9	118.81
7.8	60.84
8.5	72.25
9.0	81.00
10.5	110.25
14.0	196.00
17.7	313.29
7.5	56.25
10.6	112.36
10.0	100.00
7.5	56.25
13.5	182.25
8.5	72.25
10.0	100.00
246.2	2573.50

Average surface is
$$\frac{246.2}{25} = 9.848$$

Unit = 0.001 in. Standard Deviation = R.M.S. = α

$$a = \sqrt{\frac{2573.50}{25} - \left(\frac{246.2}{25}\right)^2} = 2.43$$

That is, the standard deviation of the hills and valleys from level is 0.00243 in.

8 jolts. Sands 1 and 3 had the poorest flowability because with each it took 32 jolts to obtain a mold which resulted in the smoothest finish. The method of preparing test specimens and measuring finish actually is a precise true flowability test. The Dietert flowability test does not indicate any appreciable difference between the six sands tested. It is recommended that further work be done to develop an improved flowability test procedure.

The figures and table showing the effect of the

number of jolts on casting finish indicate that if the optimum number of jolts is exceeded, the finish may become rougher. Sand No. 3 had a 1.75 finish after 16 jolts; this changed to 2.4 after 32 jolts.

Table 3 shows how green, dry, and hot properties

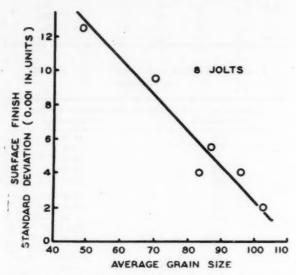


Fig. 7-Curve showing relationship between sand grain size and casting finish.

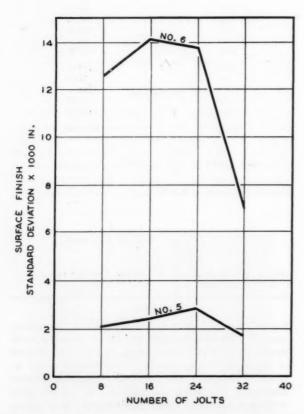


Fig. 8—Curves showing relationship between number of jolts and resulting casting finish using sand Mix No. 5 and 6.

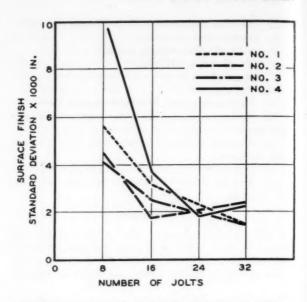


Fig. 9-Curves showing relationship between number of jolts and resulting casting finish using sand Mix No. 1, 2, 3 and 4.

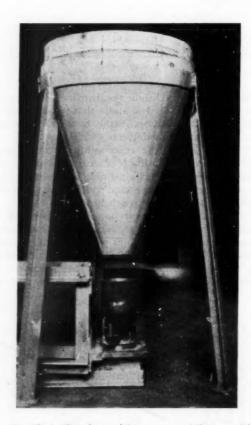


Fig. 10-Flask filler in position over molding machine.

Ca gi fi

st chim W

of sand are altered by changes in the distribution pattern. Sand No. 1 has the lowest green, dry, and hot strength. It is evident that 200, 270 mesh and pan material increase green, dry, and hot strength, also

they improve finish and flowability.

There is a definite limit to the amount of fine silica material that can be added to a sand to improve finish. If the hot strength is increased beyond a certain point, rat-tails and scabs may occur on the castings. This finding was reported by the A.F.S. Committee on the Properties of Gray Iron Sands at Elevated Temperatures.

Further work on casting finish should be done to evaluate materials that can be added to molding sand

to improve finish.

Conclusions

1. In these tests the sand with the lowest permeability and highest grain fineness produced the best finish.

2. Peaked or normal grain distributions did not have the best flowability. A sand with a nonsymmetrical, broad distribution had the best flowability.

3. The Dietert flowability test did not indicate the true flowability of the sands tested. It is recommended that further work be done to develop a more suitable flowability test.

4. There is evidence to show that if a sand is jolted too much, the casting finish is made rougher.

Bibliography

1. H. H. Fairfield and James MacConachie, "Casting Surface Finish," A.F.S. Transactions, vol. 57, pp. 521-525 (1949).

2. A. Hobman, "How to Measure Surface Roughness of Cast-

2. A. Hobman, "How to Measure Surface Roughness of Castings," American Foundryman, Oct. 1949, p. 46.

APPENDIX

Some further details of the molding machine are shown in Fig. 10. The mold is filled by riddling sand through the sieve into the hopper. The pattern plate is mounted on a hinged frame-work. To jolt the mold, the pattern is raised $1\frac{1}{2}$ in. and dropped on the bumping block.

Figure 11 shows other equipment used with the

molding machine.

Table 6 shows how the surface finish value is calculated. Twenty-five measurements are taken with a needle-tipped dial indicator.

DISCUSSION

Chairman: J. B. Caine, Foundry Consultant, Wyoming, Ohio. Co-Ghairman: R. H. Jacoby, The Key Co, E. St. Louis, Ill.

H. W. DIETERT: 1 We agree with the authors that more work must be done on casting finish and flowability and ramability of casting sands. I wish the authors had shown a measure of the green hardness of the molds. It would be helpful if we had those figures.

I do not agree with the figures on flowability in Table 3. My experience leads me to believe that these figures on flowability are in error. Here we have changed the green compressive strength from 5.3 psi to 11 psi in all sands. And we have changed grain deformation from about 0.014 in. per in. to 0.046 in. per in. The range of sand toughness is from 117 to 395. When you change that much in green compressive strength, green deformation, and in the toughness of the sand, the flowability indicator would range from possibly 75 to 85. The authors show a range from 80 to 83. These values cannot be correct.

Mr. FAIRFIELD: Our previous paper, "Casting Surface Finish"

Fig. 11-Left to right; cope flask, drag flask extension, drag flask, core box for gate core, and pattern for cope.

which is published in A.F.S. Transactions, vol. 57, p. 521 (1949) shows the relationship between mold hardness and casting finish. The flowability tests were conducted by A. E. Murton of the Bureau of Mines and Technical Surveys, Ottawa, Canada.

A. E. Murton: ² Unfortunately there was not sufficient sample left to check the test results. We agree with Mr. Dietert that the flowability values seem rather high for the recorded toughness values. However, we believe that if this is an indication that the results are in error, the fault is more likely to lie with the toughness values than with the measurement of flowability.

We have had considerable difficulty at our laboratory in geting what we consider to be significant deformation tests with certain types of sand. We attribute this to the fact that with these sands slight changes in tempering moisture have a great influence on the type of failure which is obtained. Because of the rapidity with which the sand is deforming at the instant of failure, slight differences in the type of failure obtained will result in great differences in the deformation reading. We believe that in such cases the so-called toughness value is extremely difficult to interpret.

The following measurements are typical of those obtained at these laboratories. These tests were made on a fine-grained synthetic sand, containing 7 per cent Western bentonite, tempered to different moisture contents. Results are shown in quadrupli-

cate to indicate the reproducibility.

Moisture, per cent	Green Compressive Strength, psi	Green Deformation, in. per in.	Toughness	Flowability Dietert No.
3.4	9.5	0.0115	109	79
	9.7	0.0115	111	78
	9.4	0.0345	324	78
	9.5	0.0125	119	78.5
5.8	7.7	0.026	200	75.5
	7.6	0.025	190	74.5
	7.4	0.0265	196	74.5
	7.6	0.026	197	75
7.0	7.2	0.0245	177	77
	7.3	0.026	190	76.5
	7.2	0.025	180	78
	7.2	0.026	187	76

At the instant before the failure of the above specimens the deformation readings were between 0.005 and 0.012. It is evident, therefore, that most of the deformation recorded by the deformation indicator had taken place at the instant before failure.

We believe that Mr. Dietert has introduced a worthwhile modification of his deformation test in his proposed "yield load limit." For this test the strength-deformation recorder attachment for the Universal sand strength machine is required. The point where the stress ceased to be proportional to the strain be recorded as the yield load limit. For those sands which do not have a definite yield load limit a factor analogous to the 0.2 per cent proof stress used in testing metals might prove to be of practical value. Certainly there would appear to be more hope of obtaining significant measurements of deformation before the

and-E

¹ President, Harry W. Dietert Co., Detroit.

^a Metallurgist, Department of Mines & Technical Surveys, Physical Metallurgy Division, Ottawa, Canada.

go

at

be

ar

me mi gr th th ca ist fo an an op

ist in ex lo

tu

wi ca du co

an tec fro is wi

specimen breaks than there is in the measurement of the deformation at the instant of failure of the sand specimen.

E. E. Woodliff: Did the authors record density? There certainly must be a wide differential, and if there was, did it show anything in relationship to finish? I have always felt that finish had a great deal to do with ram density or bulk density.

MR. FAIRFIELD: If we assume that mold hardness is related to density then it is apparent that with a mold hardness of 40, a smooth casting results when sand is riddled. Lumpy sand rammed to a mold hardness of 40 resulted in a rough casting. The riddled sand has a better flowability than does the lumpy sand.

I would say from our findings that surface finish is affected by the way you put the sand into the mold, how much compacting effort you put on the sand, and moisture content. Then properties of screen distribution come into the picture.

Mr. Woodliff: It would have been interesting to compare the sands by a study of density.

MR. FAIRFIELD: If you ram a sand hard enough you will probably reach a point where the finish is dependent upon density. The finish obtained from a soft rammed mold is more dependent upon flowability.

Mr. WOODLIFF: You have listed hardnesses up to 60. Would you allow a 60 mold hardness to prevail in your shop?

Mr. FAIRFIELD: On light bronze work yes, but not on grayiron and steel.

MR. DIETERT: I am interested in the statement that you can over-ram sand with 16 jolts, for example. In Fig. 9, Sand No. 2 imparted a poor finish when the sand received more than 16 jolts. Was the flask bolted down to the jolt molding machine to avoid whipping of the sand? That may account for the loss of surface finish. Does the author have an explanation for this? How hard was the mold?

MR. FAIRFIELD: When we first designed this molding machine we did get some whipping action of the flask. That is why we raised it only 1½ in, to jolt. You will note that flask has double lugs and the pin fits very close. Hence, the overramming is due to the increase in mold gas pressure as your sand packed more tightly. You get the same effect as you do with an increased moisture content.

MR. DIETERT: Mr. Fairfield stated that the increase in roughness was due to gas pressure. The permeabilities of the sand used ranged from 75 to 126. This is an exceedingly high permeability for the small test casting. So, one would not expect gas pressure difficulties from hard ramming of the sand. What appears to be the case is insufficient sand between the pattern and flask wall. The flask being of solid wall should aggravate this apparent poor rigging. Thus, the increase of roughness obtained on hard ramming may not be due to the sand primarily but due to nature of the rigging.

³ Consultant, Foundry Sand Service Engineering Co., Detroit.

WHAT THE COLLEGE GRADUATE CAN DO FOR THE AMERICAN FOUNDRYMAN

By

Clement J. Freund*

CAN THE COLLEGES OF ENGINEERING do any real good for the American foundry industry? Let us look at the state of technology in the castings industry; because technology and the colleges of engineering are very closely interrelated.

We certainly should commend American foundrymen for the notable technical advancement they have made, especially in the past 25 years. We must congratulate our Society because it is the strong men in the American Foundrymen's Society who have been the leaders in this technical advancement.

Some foundrymen can remember when making castings was largely a craft or manual skill. The chemist was the first technologist who came into the foundries. Usually he had no authority. He analyzed and reported on the ingredients of melts and scrap, and if he ever ventured an opinion on any general operating question, he was quickly and sharply told that he was not in charge of the business.

The metallurgist came a few years after the chemist, and then followed chemical, electrical, mechanical, industrial, ceramic and even ventilating engineers, experts in materials handling, sand control, metallography and x-ray technicians, and others.

Today the smart foundryman depends largely upon exact and experimental data, which he has substituted for opinion and guesswork.

But in spite of the progress which has been made, our industry still has a very long way to go, and it will be many long years before foundry technology can compare with the technology of the chemical industries, for instance, or the aircraft industry, or the communications and electronics industries.

American foundrymen have had the help of engineering college professors and graduates in the past, and if they expect to accelerate the progress of their technology, they will have to have still more help from the professors and graduates from now on. This is basic; there can be no first-class foundry technology without the colleges of engineering.

Confusion About College Aims

So much for foundries and foundrymen, at least for the present. Now let us look at the colleges of engineering.

There is confusion and misunderstanding about the job the colleges should do, about their objectives or function. You can hardly expect the rank and file of the citizens, the public, fully to understand the colleges and their work. But college presidents and deans sometimes have been equally befuddled. Indeed, the assignment of this talk has done me a good turn. It has forced me to go carefully into the philosophy of engineering college education, with the result that I am somewhat less perplexed than I had been.

I assume that we are not talking about the professional and graduate schools in the universities, and it is clear that we are not talking about trade or vocational schools. We are talking about colleges, and when we Americans talk about colleges we mean a four-year program of study following immediately upon graduation from high school or preparatory school, and leading to the first or bachelor's degree.

Possibly we can most easily clarify the job which the colleges can do for the foundry industry if we compare the colleges with the other types of schools.

Trade Schools

The freckled lad, with a shining new tool box under his arm, who comes into your shop after graduating from a trade school is a molder or coremaker, and he should be ready to do useful work right away. Of course, you have to give him simple tasks, ordinary cope and drag jobs on the floor, to be made from plain patterns, and without complicated parting lines or difficult core structures.

He has much to learn, he needs judgment and experience; but he is already competent, in the sense that he possesses the fundamental knowledge and skill of his craft.

Vocational School

The young man who comes into your plant from the vocational school is not yet competent or skilled, if I properly understand the purpose of the vocational

[•] Dean, College of Engineering, University of Detroit, Detroit.

in

fo

th

aı

de

ha

al

hi

bi

is

ex

pi

"1

di

01

th

to

ic

schools. He has had exercises in the school shop, but not enough to make him a qualified melter or molder. The purpose of the vocational school is to discover if the young man has any interest in or aptitude for a foundry career. He is ready to become an apprentice, or to begin whatever training program you have established in your work.

Or it may be that he does not care for any kind of formal training; many boys do not. In that case he should be an especially apt candidate for learning the operation of your molding machine, slingers, coremaking machines, ovens and other mechanical equipment.

University

Now let us jump over the college, for the time being, and consider the graduate schools in the universities.

The young men who come into our industry from the graduate schools have completed programs of advanced study after graduating from college. They are scholarly men, with master's or doctor's degrees. You many laugh at them, and call them long-haired theorists and dreamers, but do not forget that these men, and only these men, have created and have full charge of our immense system of engineering science and research; and without engineering science and research our industries and our economy would soon collapse.

These men are experts in chemistry, in metallurgy, in combustion, in ceramics. They can do your creative and projective work for you, your research, but do not expect them to take much interest in your routine testing and operating problems.

College

This brings us, finally, to the technical college graduate. He may come to you from a college of engineering or technology within a university, or from an independent institution. He has a bachelor's degree in engineering or in chemistry or in metallurgy, but he does not pretend to be a good molder, melter or salesman.

He may know how to use instruments, to make drawings and to operate machines and apparatus, and he may even be familiar with foundry terminology, but learning these things has not been his principal business in college.

This brings us right up to the most fundamental idea in this whole discussion. What, then, is the principal business of the college student? His principal business in college is to understand the elements of chemistry, physics and mathematics, and the engineering techniques which are the basis of your melting and molding operations. It is his business to study our society, or at least the organization of industry and the part industry plays in our society; it is his business to learn how to think.

The college graduate is ready to learn any phase of your business in which men have to think: chemistry, engineering, metallurgy, supervision, sales, management. He still needs apprenticeship or training in the details of foundry practice, but he will learn re-

markably fast if you have carefully selected him and if your training is first rate.

If your college graduate is above average in ability, and is interested in the foundry, but makes no progress under training, don't blame the boy. Sit down and go all over your training plan to find out what is wrong with it. Some years ago I was engaged in apprenticeship work under that master of training, Harold Falk. We had hired from one of the colleges a young man of outstanding talent, but after half a year of supervised work the lad had got nowhere. I reported the circumstances to Mr. Falk, and he commented, "If that young man is making no progress, we had better look into our training operations; it is my guess they can stand some revamping."

West Point; Annapolis

By way of analogy, I should like to refer to the United States Military and Naval Academies. During the past two years I have done a little consulting work in the two institutions, and happen to be reasonably familiar with their curricula and instruction. They are distinctly collegiate in their organization and function.

I imagine that most people would be surprised to know that the primary purpose of the service academies is not to train army and navy officers. Although the cadets and midshipmen receive commissions upon graduation, these commissions are of secondary importance. The primary purpose of the academies is to educate young men to be useful and loyal citizens, and to give them such schooling that they shall afterwards be able, in the regiments and on shipboard, quickly to become fully competent officers.

Similarly, the primary purpose of the colleges of engineering and technology is to prepare for industry young men who have what we ordinarily call brains, and who will learn your business faster than you ever expect, if you give them half a chance.

A. S. E. E.

The American Society for Engineering Education, formerly the Society for the Promotion of Engineering Education, has been the most diligent and influential custodian of the objectives, the procedures and the over-all philosophy of engineering education. What there has been of consistency and coherence in engineering education has derived largely, although not exclusively, from the Society.

Most of the members of the Society, as you may imagine, are presidents, deans and professors in the colleges of engineering. But hundreds of members are engineering employers. Nationally-known corporations are especially well represented, such as General Electric, Westinghouse, Allis-Chalmers, DuPont, Bell Telephone, General Motors, Chrysler and the like

Some of you may wish to investigate more thoroughly the objectives, methods and organization of engineering college education. If so, I strongly recommend that you read the reports of this Society on Engineering Education After the War, issued in 1944;

TE

nd

ty,

vn

is

D-

ar-

I

m.

We

ny

he

m.

ng

n-

n.

on

to

le-

gh

on

m-

is

ıs,

r-

d,

of

ry

ıs,

er

n,

er-

n-

es

n.

in

gh

ay

ne orne orof mon 4; on Aims and Scope of Engineering Curricula, of 1940; or even the Report on the Investigation of Engineering Education, published in 1934.

Foundry Industry in A. S. E. E.

Civil engineers, mechanical engineers, electrical engineers, chemical engineers and every other kind of engineers have had much to say in the affairs of the American Society for the Promotion of Engineering Education, but our American foundrymen remained aloof until seven or eight years ago.

The Foundry Educational Foundation will doubtless still further improve the association between the foundry industry and engineering education. During the past two or three years Geo. K. Dreher has taken an exceedingly active and intelligent part in the undertakings and projects of the engineering education Society.

Orientation and Training

Now you will ask, what are you going to do with the lad who comes from college into your plant? He has been conditioned; he knows something about science and about engineering; he knows something about our mode of life; he should be able to express himself, and he ought to know how to think. But he is not a foundryman, not yet.

The thing to do is to train him, to give him some experience, to break him in. He has to learn the practical side of the foundry business. Employers like to say that the college boy has to have experience, but sometimes those same employers forget that there is no place on earth where the boy can get foundry experience except in the foundrymen's own foundries.

I can tell you one good way not to train the young man, and that is to give him a shovel, show him a pile of sand and forget about him. A few old foundrymen seem to be still surviving who get vindictive and ugly satisfaction out of inflicting unnecessary common labor upon the present generation of young men. "It's the way I learned," growls the old foundryman. "And it's the way he'll have to learn."

The hard labor method is the best possible way to drive the talented young men out of the foundries; only the stupid boys will put up with it. The hard labor method has long since been discarded in every other industry. Does the veteran garage owner take the self starters out of all the cars and make his mechanics crank the engines by hand just because he had to when he was young? What is the good of mechanical progress if we do not take advantage of it?

The best way to break in the college graduate is to give him tasks and assignments which will challenge his mind and his resources. If you do not keep his mind busy he will not stay with you. Make him reason out and report to you why you do things the way you do them; why you anneal this casting and not that one; why you tumble these jobs and blast those; why you stir molybdenum into this melt and not into that one.

Unless you have already done so, why not set up for your college graduates a formal, supervised course, like the two-year college apprenticeships which have been so notably successful in such organizations as the Westinghouse Electric Corporation, the Falk Corporation, the General Electric Company and others?

I suggest that you lay out on paper the schedule of operations and departments for the apprenticeship and give the boys a copy. Personally, I have always been much in favor of the regular apprenticeship contracts. They inspire confidence in the young men, and they impose upon them, and upon you, a sense of responsibility.

If yours is a large organization, with a smoothly running personnel department to handle the training, well and good. But do not let anybody tell you that you have to have a fancy organization to make training work. You can just as easily design a work plan for two or three young men and supervise their training yourself. It will take very little time. All you need do is to get to know the boys well, take a personal interest in them, and see to it that their experience proceeds according to schedule.

Conclusion

And what will you gain if you hire a few graduates from the colleges, take them into your plant and teach them your business?

For one thing, before you know it, one of these boys will have grown big enough to take over a part of the constantly growing responsibility of running your affairs; you will have a chance to relax, for the first time in your life.

And you will have a hand in promoting the technological advancement of our industry. The chemical, electrical, utility, automobile, radio, aircraft and other industries have accomplished wonders by smart recruiting and orientation of engineering college graduates.

There is not a single reason that I can think of why the foundrymen of the United States should not do at least as well.

A CASTABILITY AND PROPERTY COMPARISON OF SEVERAL MAGNESIUM - RARE EARTH SAND CASTING ALLOYS*

By K. E. Nelson** and F. P. Strieter**

Introduction

Development of the Newer, jet-type aircraft engines has required the use of improved magnesium alloys for many of their parts. The higher operating temperatures of these engines has necessitated development of alloys possessing better tensile and creep properties at temperatures in the neighborhood of 400 F. Magnesium alloys containing rare earths possess good properties at this temperature. They have been described previously in the literature as pointed out by Nelson and Strieter.¹ It has been only in the last few years that active design and production interest has developed for this group of alloys.

A paper published by Leontis² shows the manner in which the properties of this type of alloy are affected by the composition of the rare earth mixture. Meier and Martinson³ report their results on the properties of magnesium-rare earth alloys, and Nelson and Strieter¹ describe their experience with a magnesium-rare earth-zirconium alloy composition.

Grube and Eastwood⁴ have described an alloy containing magnesium, rare earths, manganese, nickel, and tungsten in which the creep strength is reported to be superior to the other alloys of this group. Ball⁵ indicates that his choice of alloy is one containing magnesium, rare earths, zinc, and zirconium.

It was the purpose of this work to investigate the castability and the tensile and creep properties of actual aircraft engine castings made from a group of alloys representing, essentially, the choices of all of the above studies. It was believed that only after a study of this type could an intelligent decision be made as to the composition exhibiting the optimum combination of casting qualities and mechanical properties.

Accordingly it was decided that this work should

compare the following alloy compositions listed by their proposed ASTM alloy designations:

EK30A-T6 Magnesium-3% Rare Earth-0.25% Zirconium

EK30B-T6 Magnesium-3% Low Cerium Rare Earth-0.25% Zirconium

EK31A-T6 Magnesium-3% Rare Earth-0.6% Zirconium

EZ33A-T5 Magnesium-3% Rare Earth-3% Zinc-0.7 % Zirconium

EM61B-T6 Magnesium-6% Rare Earth-0.8% Manganese-0.2% Nickel-0.02% Tungsten

The critical properties of present jet engine castings for applications involving magnesium seem to be static tensile properties at temperatures ranging from 70 F to approximately 400 F and creep properties at the operating temperatures of near 400 F. It was believed that a determination of these properties would best compare the serviceability of the various alloys. It was further believed that in case two or more of the alloys possess equivalent serviceability properties, then the choice of alloys might rest upon the castability characteristics exhibited by them. The term castability as used in this paper refers to the ability of an alloy to satisfactorily fill a sand mold and to form a casting relatively free of deleterious defects.

Procedure

In order to obtain results which would be as nearly representative of production conditions as possible two castings of each of four patterns were cast making a total of eight castings in each alloy. In the case of one of these patterns three castings were poured in EM61B alloy for a total of nine castings in this composition. In addition to these castings, separately cast 1/2-in. diam standard tensile bars and a 4-in. cube to obtain grain size determinations were cast from each melt. The grain sizes were determined by the comparison method described by George.6 Several melts were required to make these castings in each alloy. The castings were then inspected for the usual foundry defects. They were heat treated along with the separately cast test bars according to the schedule shown in Table 1. After heat treatment, each castTA

Allo

EK3 EK3 EZ3:

ing ture room

stre

thir

ings

crue

[•]This work was made possible in part through the support of the Materials Branch, Airborne Equipment Division, Bureau of Aeronautics. U. S. Navy Department, Contract No. NO_A (S) 10576.

^{**}Magnesium Laboratories, The Dow Chemical Co., Midland, Mich.

by

Zir-

0.7

stbe om at oeld ys. of es, ilst-

le kse in st o

ts

e

TABLE 1-HEAT TREATMENT OF VARIOUS MAGNESIUM-RARE EARTH ALLOYS

Alloy Designation	Heat Treatment
EK30A-T6	18 hr at 1050 F, air cool to room tempera- ture, 16 hr at 400 F
EK30B-T6	18 hr at 1050 F, air cool to room tempera ture, 16 hr at 400 F
EK31A-T6	18 hr at 1050 F, air cool to room tempera- ture, 16 hr at 400 F
FZ33A-T5	12 hr at 350 F
EM61B-T6	24 hr at 1015 F, air cool to room tempera- ture, 16 hr at 400 F

ing was sectioned in order to obtain a complete picture within each casting of the tensile properties at room temperature, 400 F, and 600 F, and the creep strength at both 400 F and 600 F. Bars were taken for each test as uniformly as possible from both thick and thin sections of the casting. Figure 1 shows the castings used in this program.

The alloys were made, in general, by the customary crucible practice as described by Nelson.⁷ Some spe-

cial techniques not described by him were required in order to obtain improved alloying efficiencies for rare-earth metal additions and to obtain the desired zirconium level in the melts.

In producing all of the alloys the flux used for protection from oxidation during melting and handling was Dow No. 220 flux. A layer of fluorspar approximately 1/8 in. thick was applied to the surface of the melt after it had been refined with the above flux. The pouring temperature of a given casting was the same for each composition and was the one used by the foundry in pouring EK30A alloy production castings. Likewise the gating and risering used was that previously developed for EK30A alloy.

One part of E10 alloy ingot (magnesium-10 per cent rare earth) and two parts of KO ingot (magnesium-0.5 per cent zirconium minimum) were used in making EK30A alloy. The E10 alloy was produced with a low iron content. The EK30B alloy was produced in a manner similar to EK30A, except that in this case a special composition of Mischmetal low in cerium was used as the rare earth addition. Table 2 presents typi-

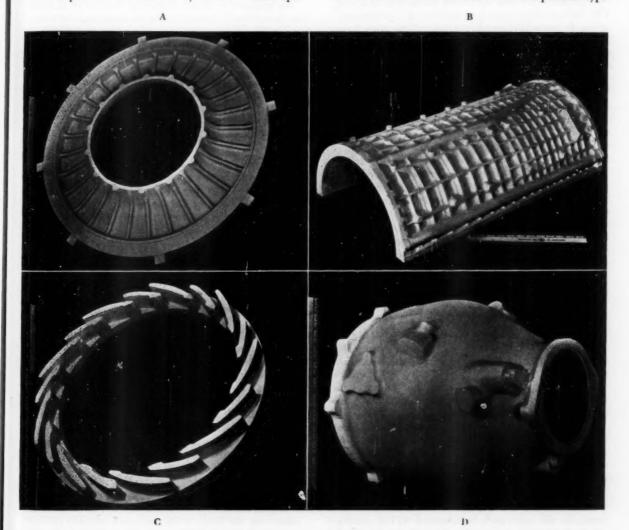


Fig. 1-Castings sectioned for property evaluation.

Table 2—Typical Analysis of Rare Earths Used in This Investigation

Rare Earth	Ce	La	Nd	Pr	Other Rare Earths and Impurities
Mischmetal	50.5	-22.5	18	6.5	2.5
Low Cerium			***		
Mischmetal	3.5	43	38	14	1.5

cal analyses of both the commercial Mischmetal and low cerium Mischmetal used for the rare earth additions in this study. The level of zirconium in EK31A is near the solubility limit and requires a different alloying procedure from that used on EK30A alloy. A zirconium hardener was added to EK30A alloy, satisfactorily producing the higher zirconium content required in EK31A alloy.

Another form of zirconium available at the time only in experimental quantities was a zirconium sponge made by the U. S. Bureau of Mines at Albany, Oregon. This material is generally regarded as a pure form of zirconium. Satisfactory zirconium alloying efficiencies were obtained using this material. The sponge zirconium was used as a "sweetening" agent. It was added to EK30A alloy along with 3 per cent zinc to produce the EZ33A alloy. Some of the zirconium-containing alloy melts were prepared partly of scrap from previous melts of same composition.

EM61B alloy was produced using cell magnesium, M1 alloy ingot (1.2 per cent minimum manganese) for the introduction of manganese, commercial Mischmetal, and a 90 per cent nickel-10 per cent tungsten hardener. The nickel-tungsten was obtained from Battelle Memorial Institute. A similar hardener was used by them in their investigation of this alloy.4 The addition of nickel-tungsten resulted in a 100 per cent recovery of nickel but little, if any, recovery of tungsten. The cell magnesium and M1 alloy, in the correct proportions, were melted and heated to 1550 to 1675 F. The nickel-tungsten hardener was then washed into the melt. The melt was cooled to 1450 to 1500 F and the rare earth alloyed. After refining and settling the melt was brought to temperature and poured. As much as 30 per cent process scrap was used in some of these melts.

Analysis of Melt

Each melt was analyzed for the major constituents and also for minor impurities. The total rare earth content, listed as T.R.E. in Table 4, the zirconium, zinc, and tungsten were all analyzed by wet chemical methods. The manganese, nickel, and the usual impurities including iron were analyzed for spectrographically.

The heat treatment was carried out in production furnaces. The solution treatment used in the alloys was done under a protective atmosphere of greater than 1 per cent SO₂ gas. The precipitation treatment on all alloys was done in ordinary atmospheres.

Following the heat treatment the castings were sectioned and the various pieces were machined to conform, for the most part, to standard ASTM specified shapes for tensile testing. In the case of a few of the specimens it was not possible to machine standard

ASTM bars. In these cases the bars were machined to dimensions listed as acceptable to the ASTM. Specimens machined for creep testing were generally 0.505-in, diameter with a 2-in, gage length. When this could not be done, the bars were machined to a ½-in, x ¼-in, cross-section with a 2-in, gage length.

The tension tests at room temperature were likewise conducted according to ASTM standard procedures. At the elevated temperatures the tensile tests were run on bars held only 10 min at temperature rather than the 60-min holding period specified in the ASTM standards. The yield strengths were obtained by measuring the stress at which the autographically obtained stress-strain curve deviates 0.2 per cent from the modulus line.

Creep tests were run'on the equipment and in the manner described by Moore and McDonald.8

Hardness, when obtained, was measured on a Brinell hardness tester using a 10-mm ball with a 500-kg load applied for 30 sec.

Castability Comparison

An evaluation of the degree of castability of these various alloys is limited to the observations made during the pouring of the castings, and to the quality of these castings as shown by various inspection techniques.

Table 3 lists these observations which were made on each casting. The pouring times required to fill the molds were measured in the expectation that this might show up gross differences in the ability of the alloy to fill the mold. The data in Table 3 indicate that no significant differences exist between these alloys.

The pits observed in these magnesium rare earth castings are found on the cope surfaces of the castings. It is felt that these pits are caused by turbulence of the metal stream. This cope pitting is less evident in EM61B alloy than in the other compositions. On the other hand, radiography of EM61B alloy castings shows similarly shaped voids to be present within the casting.

None of these alloys appears to be as prone to micro-shrinkage as the conventional Mg-Al-Zn alloy AZ63A. Radiographs of the castings indicate a greater amount of porosity in the EM61B alloy than in the other four alloys. This alloy appears to be more in the form of small entrapped air bubbles than in the form of shrinkage porosity. These voids are usually near the surface of the casting in EM61B alloy since the radiographs of test bars machined from these areas were, in nearly all cases, free of these voids. An extremely coarse grain size and a large amount of massive second phase material in this EM61B alloy in some instances creates a false impression of microshrinkage in many areas of the casting. The other alloys exhibit in some cases a small amount of foreign dense inclusions presumed to be metallic zirconium.

It can be concluded that EK30A, EK30B, EK31A, and EZ33A alloys are similar, as far as castability is concerned. EM61B alloy is somewhat inferior to the others on the basis of the increased number of voids which castings of this alloy possess. Possibly a modification in the gating and/or risering of the mold

woi

WO

cen in sile

the

sess

YS

10

kg

se irof h-

le III is 1e te se h tce nt n 35 ie iy er ne in ie ly ce 25 K-

n

0-1-

n

is

d

TABLE 3-CASTABILITY COMPARISON OF SEVERAL MAGNESIUM-RARE EARTH ALLOYS

	Casting	Pouring	Pouring			Casting Quality
Composition	Number	Temp., F	Time, Min	Vi	isual	Radiographic
EK30A	A-1	1500	0.35	Cope Pit	tting	Good, Pits
EK30A	A-2	1500	0.29		20	99 89
EK30B	A-1	1500	0.37	22 2		59 99
EK30B	A-2	1500	0.37	29 2	**	20 20
EK31A	A-1	1500	0.18	** *	9.9	52 59
EK31A	A-2	1500	0.18	29 2	20	20 22
EZ33A	A-1	1500	0.22	10 1	99	25 92
EZ33A	A-2	1500	0.22	** *		Good, very small amount of fine poros ity and a few pits.
EM61B	A-1	1500	0.44	**	**	Compound segregation. Pits, oxide in clusions and coarse porosity.
EM61B	A-2	1500	0.50	Less cope	e	22 22 22
				pitting		
				above	,	
EM61B	A-3	1500	0.25	10 11		19 10 19 29
EK30A	B-1	1350	0.52	Cope Pit	ting	Good, Pits
EK30A	B-2	1350	0.48	29 91	,	27 19
EK30B	B-1	1350	0.35	** **	*	Pits, possibly very small amount of porosity.
EK30B	B-2	1350	0.35	90 21		11 11 11 11 11 11
EK31A	B-1	1350	0.63	19 19	,	Good, Pits
EK31A	B-2	1350	0.69	10 11	,	11 11
EZ33A	B-1	1350	0.67	10 11		Pits, some porosity.
EZ33A	B-2	1350	0.69	** **		Pits, little porosity.
EM61B	B-1	1350	0.94	Less cope		Pits, coarse porosity and compound
2.11.2012		1550	0.51	pitting		segregation.
EM61B	B-2	1350	0.77	n n	,	99 99 90 99 99
EK30A	C-1	1500	0.73	ОК		Good
EK30A	C-2	1500	0.65	UK		"
EK30B	C-1	1500		OK		
EK30B	C-2	1500	_	OK.		
EK31A	C-1	1500	0.95	ОК		Good
EK31A	C-2	1500	0.93	OK.		COOL
EZ33A	C-1	1450	0.59	ОК		Good
EZ33A	C-2	1450	0.55	UK "		Good
EM61B	C-1	1500	0.86	ОК		Fair, compound segregation and coarse porosity.
EM61B	C-2	1500	0.86	99		Compound segregation and coarse porosity.
EK30A	D-1	1500	0.34	Pitting or		Good, some pitting
EK30A	D-2	1500	0.27	" "		92 99 99
EK30B	D-1	1500	0.36	20 20		20 29 29
EK30B	D-2	1500	0.37	70 07		Good, more pitting than above
EK31A	D-1	1500	0.16	20 19		Pits, small amount of fine porosity
EK31A	D-2	1500	0.16	29 99		22 ES ES 23 25 ES
EZ33A	D-1	1500	0.34	99 99		Good, Pits
EZ33A	D-2	1500	0.36	10 10		n n
EM61B	D-1	1500	0.31	Less pittie	4.7	Pits, porosity in thin and adjacent sections. Fine crack 1½ in. long
FACUE	D 0	1500	0.00	20 10		inside wall.
EM61B	D-2	1500	0.28	19 90		Pits, porosity in thin and adjacent

would have reduced these voids in EM61B alloy, in which case the castability of the five compositions would be more comparable.

Property Evaluation

The analysis and grain size as measured in the center of a 4-in. cube and the grain size as measured in the reduced section of a standard ½-in. diam tensile bar are presented for each melt in Table 4. This information indicates that these alloys are similar in their grain sizes except for EM61B alloy which possesses a considerably larger grain size. Also shown are

the minimum and average tensile properties at room temperature, 400 F, and 600 F of test bars machined from individual castings.

EM61B-T6 is the most brittle alloy with EK30A-T6 and EK30B-T6 following in decreasing order of brittleness. Forty per cent of the EM61B-T6 bars tested at room temperature were so brittle that the stress-strain curve did not deviate 0.2 per cent from the modulus line in the tension test before fracture occurred. No yield strength value could be obtained on these bars. This explains why, in castings B, C, and D, the minimum yield strength values are higher

EK

F.K

EK

EZ:

EM

E7.3

EM

obt

the

the

but

dat

eve

EK

ana

the

dat

var

leve

qui

dra

the

TABLE 4-ANALYSIS & GRAIN SIZE OF MELTS & TENSILE PROPERTIES OF BARS SECTIONED FROM INDIVIDUAL CASTINGS

Compo-	Casting				Analysis				Av	e. Grai	n Diam	١.			T	ensi	le I	rop	erti	es			
sition	No.		Sol.	Inso	l.				_	Test	4-In.		7	0 F			40	0 F			600	F	
		TRE	Zr %	Zr %	Fe %	Mn %	Ni %	W %	Zn %	Bar, In.	Cube, In.		No.of	10 p	000 si	0/ N		10 f ps		0%	No.	10 f p	-
		70	/0	/0	/0	/0	70	70	/0	444.	411.		Bars	-	_			-	_			-	_
EK30A-T6	A-1	3.03	0.28	0.03	< 0.001	0.014	< 0.001			0.004		Min				1/2							
												Aver	28	18	13	2	21	16	11	10	21	12	6
	A-2	3.04	0.22	0.04	< 0.001	0.015	< 0.001			0.005	0.010					1 100							
										0.004	0.000	Aver									20		
K30B-T6	A-1	3.25	0.31	0.04	0.003	0.004	< 0.001			0.004	0.008	Min Aver		23							18	7	
	A-2	9 05	0.31	0.04	0.003	0.004	< 0.001					Min		19				18			10		
	A-Z	3.43	0.31	0.04	0.003	0.004	(0.001					Aver						20			19		
К31А-Т6	A-1	9.86	0.56	0.07	0.002	0.010	< 0.001			0.002	0.004	Min	_								_		
KSIA-10	A-I	4.00	0.00	0.07	0.002	0.010	(0.001					Aver									19		
	A-2	9 86	0.56	0.07	0.002	0.010	< 0.001					Min		22				15			_		6
	14-4	4.00	0.50	0.01	0.002	0.010	0.001					Aver	23	24	16	5	19	17	12	16	19	12	7
Z33A-T5	A-1	2.41	0.69	0.04	< 0.001	0.012	< 0.001		3.00	0.001	0.004	Min	_	19	13	2	_	17	9	6	_	11	7
2.00.1 2.0			0.00	0.0.								Aver	23	21	14	4	19	18	10	17	19	11	7
	A-2	2.41	0.69	0.04	< 0.001	0.012	< 0.001		3.00			Min	_	20	14	2	_	18	9	9	_	11	6
												Aver								18	19		
M61B-T6*	A-1	5.85			0.048	0.99	0.25	< 0.006		0.024	0.030		_					14				14	
												Aver						16			19		
	A-2	6.29			0.040	1.0	0.25	< 0.006		0.030	0.070		_								_		
												Aver						15			19		
	A-3	6.29			0.040	1.0	0.25	< 0.006				Min	-					14 15		1		15	
				0.14	-0.003	0.000	<0.001			0.004	0.007	Aver	_	14				17				10	
K30A-T6	B-1	2.74	0.25	0.14	< 0.001	0.022	< 0.001			0.004	0.007	Min Aver						17			10		8
		0.74	0.05	0.14	<0.001	0.000	<0.001					Min						16			_		
	B-2	2.74	0.25	0.14	< 0.001	0.022	< 0.001					Aver						17			10		
K30B-T6	D 1	9.07	0.96	0.04	< 0.001	0.007	_0.001			0.004	0.010					1/2							
K30D-10	D-1	4.31	0.40	0.01	₹0.001	0.007	0.001			0.001	0.010	Aver				1							
	B.9	9 97	0.26	0.04	< 0.001	0.007	< 0.001					Min				1/2 .					_		
	20-4	4107	0.20	0.02	0.000	0.007						Aver				2					10		
K31A-T6	B-1	3.22	0.54	0.09	< 0.001	0.018	< 0.001			0.002	0.004	Min	-					16		9	_	11	7
					-							Aver	16	21	14	3	10	17	11	13	10	11	8
	B-2	3.22	0.54	0.09	< 0.001	0.018	< 0.001					Min	-	18	14	2 .	_	16	11	11	_	10	6
												Aver	16	21	14		9		11	14	10	11	7
Z33A-T5	B-1	2.28	0.63	0.05	< 0.001	0.004	< 0.001		2.84	0.001	0.004	Min	-	16	13		-		_				
												Aver					10				10		7
	B-2	2.28	0.63	0.05	< 0.001	0.004	< 0.001		2.84			Min	_				_					11	6
										0.00:	0.000	Aver				. 2						7.0	7
M61B-T6*	B-1	6.47			0.038	0.96	0.26	< 0.006		0.024	0.090		-					14				13	
					0.000	0.00	0.00	<0.000				Aver		-				15			10		
	B-2	6.47			0.038	0.96	0.26	< 0.006				Min	16			0 .							
												Aver	10	1.2	15	U	10	13	1.4		10	13	3

*EM61B-T6 bars in which the stress-strain curve did not deviate 0.2% from the modulus line were not included in the yield strength values.

than the minimum tensile strength values. The bar broke before the yield strength was reached. The bars from casting C in EM61B-T6 alloy tested at 400 F are so low in ductility that little or no difference can be seen between the yield and tensile strengths. The coarse grain and excessive amount of second phase material found in EM61B-T6 alloy are largely responsible for this low ductility. In some cases test bars of EM61B-T6 alloy free of all defects including massive compound segregation had ductilities so low that yield strength determination could not be made. Figures 2, 3, 4, 5, and 6 are photomicrographs showing the typical structure of the alloys studied in this investigation.

The two best alloy compositions, as far as tensile properties of bars cut from individual castings are concerned, at each testing temperature are: (1) at room temperature, EK30B-T6 and EK31A-T6, (2) at 400 F, EK30B-T6 and EK31A-T6, (3) at 600 F, EM-61B-T6 and EK30B-T6. It can be noted that EM-61B-T6 retains a greater proportion of its room temperature strength at elevated temperatures than the other alloy compositions.

In order to provide the design engineer with data suitable for his purposes, the tensile results were examined statistically. Cumulative frequency distribution curves for the tensile strength, yield strength, and elongation were plotted on probability paper. Any given location on one of these curves represents the percentage of bars which have a value equivalent to or less than that defined by the point. From these curves the average value (50 per cent point on the curve) and the standard deviation (difference between the 83½ per cent and 50 per cent values) were

TABLE 4- (CONT.)

Compo-	Casting				Analysis	1			Av	e. Grai	n Diam				T	ens	ile 1	Prop	erti	cs			
sition	No.	5	Sol.	Insol						Test	4-In.		7	0 F			40	0 F			600	F	
		TRE	Zr	Zr	Fe	Mn %	Ni %	W %	Zn %	Bar, In.	Cube In.		No.of	10			No.o			% N		100 f ps	
		70	70	70	70	70	,-	,,,	70				Bars				_						-
EK30A-T6	C-1	2.42	0.36	0.05	< 0.001	0.020	< 0.001			0.002	0.009	Min	_	16	12	1	_	15	11	5	_	11	7 !
												Aver			13		12			-	12		7
	C-2	3.36	0.22	0.04	< 0.001	0.016	< 0.001			0.003	0.008						-			-			6
												Aver								4		12	
EK30B-T6	C-1	3.00	0.19	0.04	0.001	0.008	< 0.001			0.004	0.010						-			2		12	
	0.0	0.00	0.10	0.04	0.001	0.000	<0.001					Aver				1.00	12			2		-	9 3
	C-2	3.00	0.19	0.04	0.001	0.008	< 0.001					Aver					12			3			8 5
EK31A-T6	CI	9.05	0.56	0.08	< 0.001	0.014	_0.001			0.002	0.004				15		_			4		11	7 5
K31/1-10	C-1	4.93	0.30	0.03	₹0.001	0.011	(0.001			0.002	0.001	Aver			16			17				12	8 4
	C.2	2.95	0.56	0.03	< 0.001	0.014	< 0.001					Min	_	18	14	1	_	16	12	5	_	10	5 5
	04	2.33	0.00	0.00	0.001	0.011	201001					Aver		19		2	12	17	13	7	12	11	8 4
Z33A-T5	C-1	2.72	0.70	0.05	< 0.001	0.002	< 0.001		2.87	0.001	0.004	Min	_	18	12	2	_	16	9	6	_ !	11	6 1
ALLOUIS AU			0.10	0100	~	410.00	-					Aver	14	19	13	3	12	17	9	11	12	11	7 !
	C-2	2.72	0.70	0.05	< 0.001	0.002	< 0.001		2.87			Min	_	17	12	2	_	16	9	7 .			6 2
												Aver		19			12				12		
M61B-T6	C-1	6.38			0.042	1.02	0.29	< 0.006		0.030	0.060	Min									- !		
												Aver			13						12		
	C-2	6.38			0.042	1.02	0.29	< 0.006				Min			14						_ !		
										0.000		Aver									12		
EK30A-T6	D-1	2.69	0.25	0.05	< 0.001	0.022	< 0.001			0.006		Min					4				4		
		0.00	0.04	0.05	-0.001	0.000	<0.001					Aver					_				_ :		
	D-2	2.69	0.25	0.05	< 0.001	0.022	< 0.001					Aver		17			4		9	9	4		5 4
V 90D TE	- D.1	9.07	0.49	0.05	< 0.001	0.007	<0.001			0.002	0.006		-	-	17	i		19	-	7 .		-	7 5
K30B-T6	D-1	3.07	0.44	0.03	0.001	0.007	₹0.001			0.002	0.000	Aver		-	17	-		20		9			8 5
	D.2	3.07	0.49	0.05	< 0.001	0.007	-0.001					Min	-		17	1		20		6 .	_ !	13	7 3
	D-2	3.07	U. 1m	0.05	0.001	0.007	7,0,001					Aver	30	25	17	3	4	20	15	8	4 1	13	7 4
K31A-T6	D-1	2.79	0.53	0.03	< 0.001	0.017	< 0.001			0.002	0.004	Min	_	19	15	1	-	16	12	7 .	_ !	11	6 3
												Aver	29	23	16	3	4	17	12	8	4	11	6 4
	D-2	2.79	0.53	0.03	< 0.001	0.017	< 0.001					Min	_	20	15	2			11	6		-	5 4
												Aver			16	3	-	16		8			6 5
7.33A-T5	D-1	2.64	0.61	0.05	0.002	0.004	< 0.001		2.87	0.001	0.005				13	-		17				-	6 4
									0.05			Aver	-	21	16	2				15		-	6 6
	D-2	2.64	0.61	0.05	0.002	0.004	< 0.001		2.87			Min		-	13	1		19	-	12 .		11	6 3
					0.005		0.00	.0.000		0.004	0.000	Aver			15	2	-	19		15			7 5
M61B-T6	D-1	6.47			0.038	0.96	0.26	< 0.006		0.024	0.090		29			0		17		2 .			8 1 8 2
		C 48			0.038	0.96	0.96	< 0.006				Aver	29	-	15	0	4			-	4		
	D-2	0 47																					

The minimum tensile properties recorded for all compositions and testing temperatures were not necessarily obtained from the same test specimen.

obtained. These values are listed for each alloy at the three test temperatures in Table 5. Also listed is the number of values included on each determination.

Figures 7 and 8 are examples of five typical distribution curves obtained in this study. In general, the data form a fairly straight line. In a few cases, however, the data are somewhat erratic as shown for EK30B-T6 alloy in Fig. 7. This particular curve was analyzed further by replotting the data according to the separate melts made of this composition. These data are shown in Fig. 9, and demonstrate that a variation in alloy composition influences the property level. Since a certain range in composition is required in any alloy, it is felt that an average line drawn through the data as in Fig. 7 is justified.

In two cases there was some question regarding the advisability of drawing an average straight line. through the plotted data. These specific data can be reproduced by drawing a curved line through the following points on probability paper:

Composition	Test	3% Value	50% Value	97% Value
EM61B-T6 Elong	ation at 600 F	0.25%	3.5%	20.5%
	Strength at 400 F	8700 psi	9400 psi	11,500 ps

In both of these cases the standard deviation was not calculated and the average value listed in Table 5 is the arithmetic average of the individual values.

A comparison of average tensile properties of separately cast bars with bars sectioned from castings for all melts and castings poured is given for the three testing temperatures in Figs. 10, 11, and 12. The properties of the separately cast bars poured on each melt were averaged and plotted against the average

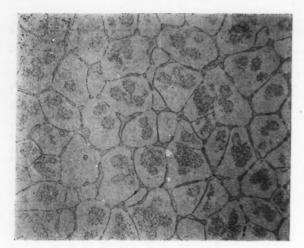


Fig. 2-Structure of EK30A-T6 Alloy. Glycol etchant. 100×.

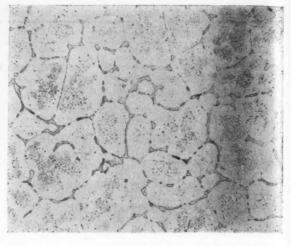


Fig. 3-Structure of EK30B-T6 Alloy. Glycol etchant. 100×.

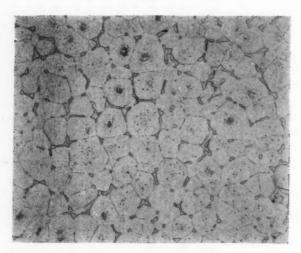


Fig. 4-Structure of EK31A-T6 Alloy. Glycol etchant. 100×.

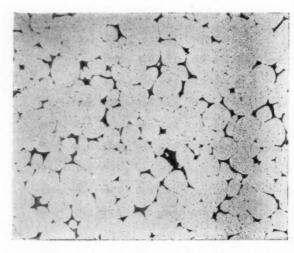


Fig. 5-Structure of EZ33A-T5 Alloy. Glycol etchant. 100×.

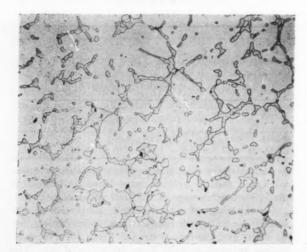


Fig. 6—Structure of EM61B-T6 Alloy. Glycol etchant. $100\times$.

properties listed in Table 5. The properties of the separately cast bars are better than those from the castings at room temperature and are as good as or better than the casting properties at 400 F and 600 F. The outstanding compositions are far as average tensile properties of bars sectioned from all castings are the same as observed earlier in comparing average properties of individual castings with the various compositions.

Samples were taken from similar sections of both A castings in all five alloys for Brinell hardness tests. The average values for each casting shown in Table 6 check each other closely within each alloy composition. Although the differences between them are small, the alloys can be listed in increasing order of hardness as follows: EK30A-T6, EK31A-T6, EZ33A-T5, EK30B-T6, and EM61B-T6.

The 100-hr creep limits of bars from each casting tested at 400 F and at 600 F are shown in Table 7. It was not possible to obtain creep limits on casting

EK

EK

EZ:

EM

stre

PERCENT

Fig

Та

Con EK

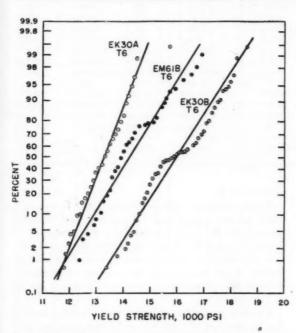
EK EZ

EM

TABLE 5-AVERAGE PROPERTIES AND STANDARD DEVIATION OF BARS CUT FROM CASTINGS

			70 F		4	00 F			600 F		
		TS YS Elon 1000 psi %		Elong	TS	YS	Elong	TS	YS	Elong	
Alloy	Property			%	100	1000 psi		100	%		
EK30A-T6	Average	17.5	13.3	1	16.0	11.0	7.5	11.4	6.6	39	
	Std Deviation	1.8	0.6	1	1.0	0.7	2.5	0.7	0.6	9	
	No. of Bars	175	175	175	93	93	93	81	80	80	
EK30B-T6	Average	21.7	16.0	1.5	19.2	13.4	5	12.6	8.2	40	
	Std Deviation	2.4	1.0	1	1.1	0.6	2.5	0.6	0.6	12	
	No. of Bars	165	164	165	89	88	89	88	87	88	
EK31A-T6	Average	22.1	15.3	3.5	16.7	12.2	12	11.2	7.1	51	
	Std Deviation	1.8	0.8	1	0.6	0.8	5	0.6	0.8	11	
	No. of Bars	165	166	165	89	89	89	89	89	89	
EZ33A-T5	Average Std Deviation No. of Bars	19.8 1.4 166	14.3 1.2 165	2.5 1 166	18.0 0.8 90	9.8	14 4.5 90	11.3 0.4 102	6.7 0.4 102	44 13 102	
EM61B-T6*	Average	14.6	14.2	0	15.0	11.8	1	14.7	8.7	5.5	
	Std Deviation	1.7	1.0	-	1.4	0.9	1	1.4	0.5	-	
	No. of Bars	168	100	168	100	98	99	90	90	90	

*EM61B-T6 bars in which the stress-strain curve did not deviate 0.2% from the modulus lines were not included in the yield strength values.



t.

1e

ie

or

F.

n-

re

re

18

le

i-

re

of

g 7.

Fig. 7—Cumulative frequency distribution curves of the yield strength at 70 F of bars sectioned from castings.

TABLE 6-BRINELL HARDNESS VALUES MG-RE-ALLOYS SAMPLES FROM CASTING A

Composition	Casting	Brinell	Hardness	Number
EK30A-T6	A-1		45.4	1
	A-2		47.5	
EK30B-T6	A-1		54.4	
	A-2		54.6	
EK31A-T6	A-1		48.7	
	A-2		48.6	
EZ33A-T5	A-1		51.3	
	A-2		50.9	
EM61B-T6	A-1		55.1	
	A-2		54.2	

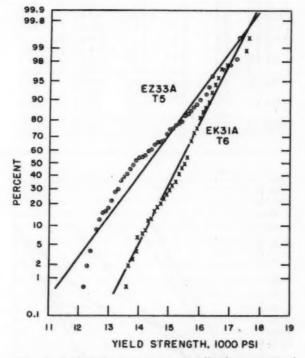


Fig. 8—Cumulative frequency distribution curves of the yield strength at 70 F of bars sectioned from castings.

D because of the geometry of the casting and limitations of testing equipment. The four conventional creep limits presented may be defined as follows:

- 1. Limiting stress to give 0.1 per cent creep extension in 100 hr.
- 2. Limiting stress to give 0.1 per cent creep plus plastic extension in 100 hr.
- 3. Limiting stress to give 0.2 per cent total extension in 100 hr.
 - 4. Limiting stress to give 0.5 per cent total exten-

TABLE 7-100-HR CREEP LIMITS OF BARS SECTIONED FROM INDIVIDUAL CASTINGS

		Range-100-Hr Creep Limits-1000 psi													
				400 F					600 F						
Composition			0.1%Creep Extension	0.1% Creep +Plastic Extension	0.2% Total Extension	0.5% Total Extension		0.1%Creep Extension	0.1% Creep +Plastic Extension	0.2% Total Extension	0.5% Tota				
EK30A-T6	A-1	2	6.4-7.0	6.0-6.9	5.7-6.8	8.2-8.6	2	1.02-1.10	0.95-1.09	1.22-1.31	1.47-1.57				
99	A-2	1	6.6	6.4	5.8	8.5	1	1.00	0.98	1.30	1.54				
EK30B-T6	A-1	2	8.5-9.3	8.3-8.8	7.8-8.2	10.4-11.1	2	0.96-0.97	0.94-0.97	1.08-1.22	1.36-1.47				
99	A-2	2	8.0-9.1	8.0-8.7	7.4-8.4	10.7-10.8	2	1.25	1.23	1.20-1.31	1.47-1.48				
EK31A-T6	A-1	2	9.1-9.3	8.8-9.2	9.1-9.2	10.5-11.3	2 .	1.21-1.40	1.20-1.38	1.37-1.56	1.62-1.80				
10	A-2	1	9.2	9.0	9.0	10.9	2	1.26-1.38	1.26-1.38	1.40-1.54	1.62-1.68				
EZ33A-T5	A-1	2	7.9-8.7	7.5-7.7	6.7-7.0	9.6-10.4	2	0.95-1.06	0.95-1.06	1.16-1.26	1.66-1.69				
**	A-2	2	8.5	7.7-7.9	7.4-8.0	9.5-10.2	2	0.75-0.86	0.75-0.86	0.97-1.13	1.43-1.59				
EM61B-T6	A-1	2	11.0-12.2	8.6	7.8	11.1-11.2	2	2.35-2.85	2.35-2.82	2.80-3.04	3.30-3.70				
29	A-2	2	9.6-11.2	8.0-9.2	7.1-9.0	11.1-11.4	2	2.60-3.18	2.60-3.07	2.66-3.35	3.15-4.00				
29	A-3	2	11.6-12.7	8.4-9.2	7.0-8.1	11.4	2	2.43-2.46	2.30-2.45	2.61-3.10	3.15				
EK30A-T6	B-1	2	6.9-8.0	6.8-7.3	6.7-7.2	9.4-9.6	1	1.41	1.40	1.52	1.74				
	B-2	2	6.0-7.0	5.9-6.9	5.6-6.2	8.3-8.4	1	1.25	1.23	1.50	1.66				
EK30B-T6	B-1	1	8.8	8.1	7.7	9.7	1	0.88	0.88	0.99	1.35				
10	B-2	2	8.4-8.8	8.1-8.4	7.5-8.2	9.8-10.2	1	0.93	0.90	1.08	1.42				
EK31A-T6	B-1	2	7.9-8.2	7.5-7.7	7.4-7.9	9.0-9.7	i	1.08	1.04	1.55	1.78				
19	B-2	2	8.2-8.7	7.7-8.5	7.2-8.2	9.6-9.7	1	1.13	1.12	1.36	1.77				
EZ33A-T5	B-1	2	6.5-8.6	5.5-8.2	6.0-8.6	9.2-9.6	i	0.89	0.83	1.13	1.78				
**	B-2	1	7.5	7.0	7.0	8.8	i	0.87	0.87	1.12	1.67				
EM61B-T6	B-1	2	10.6-10.9	9.7-9.9	9.3-10.0	11.0-11.4	1	2.42	2.42	2.80	3.62				
19	B-2	2	11.4	8.7-11.1	7.8	11.7-11.9	1	2.49	2.47	2.77	3.60				
EK30A-T6	C-1	2		Erratic Data			2	1.29-1.35	1.27-1.30	1.38-1.48	1.65-1.71				
99	C-2	2	8.2-8.4	7.7-7.8	7.6-8.2	9.5		1.46-1.52	1.41-1.50	1.56-1.63	1.85-1.90				
EK30B-T6	C-1	2	8.9-9.6	8.5-9.0	8.2-8.6	12.0		0.95-1.00	0.95-0.98	1.19-1.22	1.48-1.49				
"	C-2	1	9.2	9.1	8.0	10.8	ī	0.95	0.95	1.12	1.45				
EK31A-T6	C-1	2	9.4-9.8	9.0-9.3	8.1-8.5	10.4-10.7	2	1.55-1.61	1.54-1.60	1.68-1.69	1.90-2.03				
n n	C-2	2	8.7-9.2	8.2-8.7	8.0	10.4-10.7		1.40-1.52	1.39-1.51	1.56-1.66	1.87-1.94				
EZ33A-T5	C-1	2	8.6-9.0	7.3-7.4	6.6-6.9	9.6-10.3		0.94-1.30	0.94-1.29	1.26-1.52	1.76-1.97				
10	C-2	2	8.4-8.6	7.4-8.1	6.8-7.0	9.7-9.9	2	1.03-1.47	1.03-1.45	1.39-1.58	1.82-1.94				
EM61B-T6	C-1	2	12.3-12.7	9.1-9.4	8.1-8.2	12.0-12.1	2	2.56-2.89	2.56-2.85	2.83-3.19	3.85-4.10				
"	C-2	2	11.9-12.5	8.8-9.1	8.4-9.8	11.9-12.7	2	2.94-3.00	2.93-2.98	3.30-3.38	3.90-4.10				

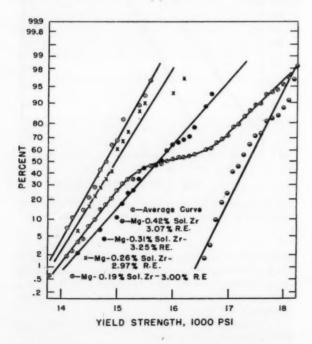


Fig. 9—Effect of composition on cumulative frequency distribution curves of yield strength at 70 F of bars sectioned from EK30B-T6 alloy castings.

sion in 100 hr.

Total extension represents the total deformation which occurs during both the loading and the creep testing period. Creep extension is the extension obtained at constant stress and temperature as a function of time after the specimen is loaded. Plastic extension is the measure of the amount of permanent deformation which occurs during the initial loading of the specimen.

Fig

ELONGATION

TENSILE STRENGTH | YIELD STRENGTH

A sufficient number of bars were available to obtain one to two sets of data for each testing temp rature for each casting. In some cases the results were too erratic to obtain creep limits and were therefore omitted from this table. Based on the results in the table, it can be seen that EM61B-T6 is somewhat superior in creep resistance at 400 F and outstanding in its superiority over the other alloys at 600 F. The table also shows that EK30B-T6 and EK31A-T6 are essentially equal in their creep resistance and are next best to the EM61B-T6 alloy at 400 F. At 600 F the four zirconium-containing alloys are all comparable.

In order to obtain an average or over-all picture of the creep of bars from castings versus alloy composition, the stress for all bars of a given composition was plotted against the percent extension in 100 hr for three parameters of creep: creep extension, creep plus plastic extension, and total extension. Lines ion

4 .47

48

.80

.68

.69

50

.70

90

.03

94

97

94

10

10

ion

ep

ob

nc-

ex-

ent

ng

in

ire

00

ore

he

nat

ng

he

are

F

ar-

of

Si-

on

hr

ep

ies

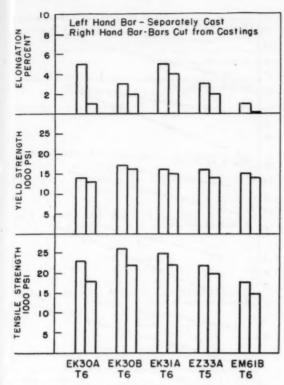


Fig. 10—A comparison of average tensile properties at room temperature of separately cast bars with bars sectioned from castings.

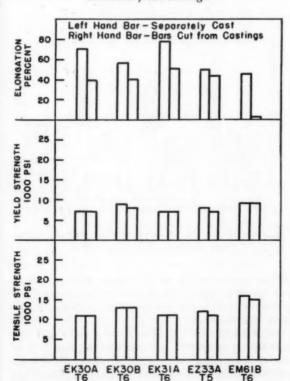


Fig. 12—A comparison of average tensile properties at 600 F of separately cast bars with bars sectioned from castings.

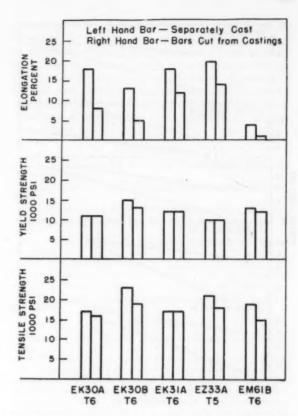


Fig. 11—A comparison of average tensile properties at 400 F of separately cast bars with bars sectioned from castings.

were drawn defining the upper and lower limits of the bands of data. On four curves one point was left outside of the band because something obviously was out of control on the test. Examples of the scatter of data showing total extension at 400 F and 600 F and the bands defining this scatter are presented in Fig. 13 and 14. The data in Table 8 are taken from these graphs showing the scatter of data obtained on the creep tests. By plotting the minimum and maximum values shown in the table it is possible to reproduce these scatter bands.

The range of creep limits of the castings poured in the five compositions are presented for four parameters of creep at 400 F in Fig. 15 and at 600 F in Fig. 16. These ranges are taken from the above mentioned scatter bands. The superiority of EM61B-T6 at 400 F and at 600 F is again apparent.

Conclusion

In order to arrive at a conclusion regarding the relative over-all merits of each of these alloys, one must consider again all of the properties tested. In view of the fact that all of these alloys are vastly superior at elevated temperatures to the other commercial casting alloys of magnesium containing aluminum and zinc, there are many applications where any one of the five alloys tested may be entirely satisfactory. The ease of handling in the foundry and in assembly operations, the cost of the materials in-

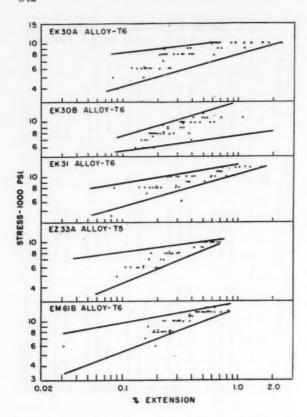


Fig. 13—Creep strength of magnesium-rare earth alloys at 400 F. Total extension in 100 hr.

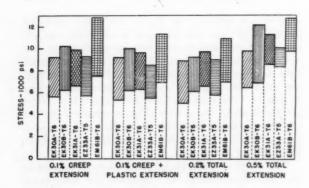
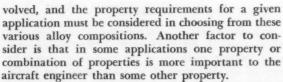


Fig. 15-Range-100 hr creep limits. Bars sectioned from castings. Tested at 400 F.



When the creep strength at 600 F becomes a more important factor than any of the other properties reported in this paper, then EM61B-T6 may be the desired alloy. The extremely brittle nature of this alloy at room temperature and the increased number of voids found in castings of this composition seem to

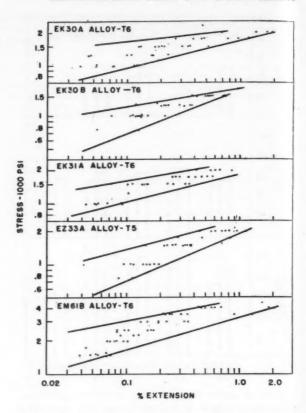


Fig. 14—Creep strength of magnesium-rare earth alloys at 600 F. Total extension in 100 hr.

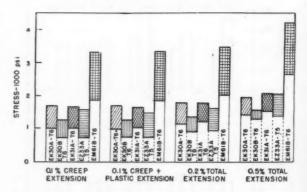


Fig. 16-Range-100 hr creep limits. Bars sectioned from castings. Tested at 600 F.

dictate against its use at the present time. It is conceivable that future foundry developments might improve the room temperature characteristics of this alloy composition.

Alloy EK30B-T6 is good in creep at 400 F and also in tensile properties at all temperatures tested. The use of low cerium Mischmetal in place of the commercial grade of Mischmetal is definitely advantageous. This material is not commercially available today, but it is understood that it could be made commercially when the demand warrants it. This composition compares favorably in castability to the other

STI

K.

Con

EK

EK

EZ3

EM6

thre same ease

T6. mere prop expe

ferio comp conic and with

need The sition

TABLE 8-VALUES WHICH DEFINE THE RANGE OF CREEP STRENGTH SCATTER BANDS FOR EACH ALLOY AT BOTH TEMPERATURES

		Stress	s Range for Given Extension 1000 psi					
	0.1% Ext		0.5% Ext					
Composition	Temp.	, Parameter	Min	Max	Min	Max		
EK30A-T6	400	Creep Extension Creep + Plastic	5.6	9.2	7.6	10.4		
		Extension	5.3	9.2	7.3	10.6		
**	**	Total Extension	4.1	8.3	6.5	9.9		
**	600	Creep Extension Creep + Plastic	1.00	1.70	1.45	1.97		
		Extension	0.99	1.69	1.44	1.98		
90	99	Total Extension	0.95	1.66	1.41	1.95		
EK30B-T6	400	Creep Extension Creep + Plastic	6.3	10.2	7.4	13.7		
		Extension	6.2	10.0	7.2	13.0		
**	**	Total Extension	5.5	7.6	6.9	12.2		
**	600	Creep Extension	0.72	1.25	1.32	1.60		
19	**	Creep + Plastic						
		Extension	0.73	1.26	1.32	1.60		
**	**	Total Extension	0.68	1.21	1.29	1.58		
EK31A-T6	400	Creep Extension Creep + Plastic	6.6	9.9	9.6	12.2		
		Extension	6.3	9.6	9.3	11.5		
**	9.0	Total Extension	5.4	8.7	8.5	11.3		
9.9	600	Creep Extension	1.02	1.66	1.51	2.09		
10	9.9	Creep + Plastic						
		Extension	1.02	1.64	1.51	2.08		
**	99	Total Extension	1.02	1.58	1.50	2.07		
EZ33A-T5	400	Creep Extension Creep + Plastic	5.7	9.3	9.9	12.6		
		Extension	5.5	8.5	9.2	10.7		
**	**	Total Extension	4.3	8.2	9.2	10.7		
	600	Creep Extension	0.72	1.56	1.41	2.05		
92	99	Creep + Plastic Extension	0.73	1.47	1.42	2.09		
20	**	Total Extension	0.70	1.35	1.39	2.05		
EM61B-T6	400	Creep Extension	7.5	12.8	12.5	15.0		
II III	. "	Creep + Plastic						
		Extension	6.9	11.2	9.7	13.6		
9.0	22	Total Extension	5.3	9.6	9.8	12.9		
**	600	Creep Extension	1.87	3.6	2.8	4.2		
**	29	Creep + Plastic						
		Extension	1.84	3.3	2.8	4.2		
12	9.0	Total Extension	1.65	3.0	2.65	4.1		

three alloys containing zirconium and it possesses the same advantages as EK30A alloy when considering ease of alloying in the foundry.

EK31A-T6 is nearly equal in properties to EK30B-T6. The raw materials involved are available commercially at this time, and its use for extra high property applications is suggested when the additional expense due to the high zirconium is warranted.

d

n-

n-

15

SO

ne

n

0

n

0-

er

The properties of alloys EZ33A-T5 and EK30A-T6 are essentially equal to each other, and slightly inferior to EK30B-T6 and EK31A-T6. The two alloys compare well in castability although the higher zirconium content of EZ33A alloy increases both its cost and handling problems in the foundry as compared with EK30A alloy.

A study of this type cannot be considered final and conclusive. More foundry and service experience is needed in order to completely evaluate these alloys. The techniques of alloying and handling these compositions are becoming more effective and practical, and

continued improvements in foundry practice can be expected. Future developments in magnesium alloys intended for elevated temperature applications might be in the direction of a low cerium Mischmetal-containing alloy with a higher zirconium content than EK30B alloy. Another potential approach toward improved properties might be along the lines of a modification of EM61B alloy to gain added room temperature ductility.

Bibliography

- K. E. Nelson and F. P. Strieter, Transactions, A.F.S., vol. 58, p. 400 (1950).
- T. E. Leontis, Trans. AIME, vol. 185, December 1949. (TP 2726E).
- 3. J. W. Meier and M. M. Martinson, Presented at the Annual Meeting, Canadian Institute Mining & Metallurgy, April 18, 1050
- 4. K. Grube and L. W. Eastwood, Proceedings ASTM, vol. 50 (1950).
- 5. C. J. P. Ball, Metal Industry, vol. 75, p. 152 (1949).
- 6. P. F. George, Transactions, A.F.S., vol. 57, p. 133 (1949).
- 7. C. E. Nelson, Trans., AIME, vol. 159, p. 392 (1944).
- A. A. Moore and J. C. McDonald, Proc. ASTM, vol. 46, p. 970 (1946).

DISCUSSION

Chairman: M. E. Brooks, The Dow Chemical Co., Bay City, Mich.

Co-Chairman: W. D. Danks, Howard Foundry Co., Chicago. A. C. Jessup (Written Discussion): ¹ I have read this paper with considerable interest. Magnesium Elektron Ltd., the pioneers in Britain of the zirconium-containing alloys of this class, have a considerable background of experimental work and practical experience the results of which do not accord with many of the conclusions reached by the authors.

Dealing first with the comparison of castability, unfortunately I have not seen the photographs illustrating the castings used by the authors, but, from results quoted in the paper, it appears that these were of a relatively simple type. Experience in this country with an alloy of the type described by Grube and Eastwood indicates that attempts to make intricate castings would have met with great difficulty from hot tearing. From our experience we should list the alloys in order of decreasing castability as follows:-EZ33A, EK31A, EK30B EK30A, EM61B. The alloy termed EZ33A is comparable to the Elektron alloy ZRE1, although the latter contains 2.5 per cent zinc and 2.7 per cent RE metals, and in practice the difference in castability between this and EK31A (equivalent to Elektron MCZ) is quite marked. ZRE1 was in fact introduced to overcome certain casting diffculties experienced with MCZ, but both of these alloys are superior to those having a low zirconium content.

The methods used by the authors for the introduction of zirconium also require comment. The accumulated experience of Magnesium Elektron Ltd., proves conclusively that the highest effective zirconium contents cannot be achieved by the use of metallic "pure" zirconium, although chemical analyses may indicate high "soluble" zirconium contents.

It is therefore probable that the effective zirconium contents reported for EZ83A, which was made from metallic zirconium sponge, are misleadingly high. The presence of microporosity in EK33A castings A2 and B2 tends to confirm this since experience with Elektron ZREI has shown that the presence of microporosity in even the most difficult casting is a very rare phenomenon, whereas microporosity readily occurs when the effective zirconium content is low.

The information quoted in the report does not enable an assessment of the effectiveness of the "zirconium hardener" used for EK31A to be made. In comparing these two alloys it would have been preferable to use a constant method for alloying the zirconium.

With regard to tensile properties, the experience in this country shows that ZREI is slightly superior to MCZ. The reported superiority of EK31A, with a zirconium content of 0.55 per cent, over EZ33A with a zirconium content of 0.70 per cent is quite

¹ Chief Metallurgist, Magnesium Elektron Ltd., England.

K.

an

917

an

all

pe

COI

em

size

wa

hai

bil

pos

Ne

pap

WO

lot

var

litt

acti

tha

pro

pro

cast

sam

ing side

Ir

allo

ferie

with

the

pare

cast

four

alloy

show

Allo

EZ 33

rang

0.6 :

and

to th

have

temp

2.9 pe

tion

in th

8 P

Ltd.,

On

TH

contrary to experience with Elektron alloys and a further confirmation that the analyzed zirconium contents in the EZ33A

alloys were deceptively high.

The values for creep strength obtained for the alloy EZ33A are somewhat lower than those obtainable from ZRE1 alloy. This is probably due at least in part to the zinc content of the EZ33A being higher than that for ZRE1, since creep strength begins to decrease if zinc is increased beyond 2½ per cent. The comparison of creep strength between the various alloys examined by the authors is probably also affected by the difference in levels in rare earth metal content. In particular it should be noted that the average rare earth metal content for the EZ33A castings is appreciably lower than that for the other alloys of nominally similar content.

K. E. Nelson (Authors' Written Reply to Mr. Jessup): In respect to Mr. Jessup's comments on castability, we have had further experience with EM61B since our paper was written. Even more difficulty was encountered than in the original work so that we can say we are in substantial agreement in the question of the castability of this alloy. In respect to the British alloy ZRE1, we wonder whether their definition of castability is the same as ours. We have thoroughly investigated the alloying of zirconium and have successfully introduced it into magnesium rare earth alloys in several ways. There has been no essential difference in the soluble zirconium content, the tensile properties, or in the castability as defined in the paper, of alloys made with various zirconium alloying agents.

The remarks about the tensile properties of ?RE1 compared to EK31A are probably not strictly applicable. They are actually comparing ZRE1 to another British material, MCZ, which would correspond to EK31A-T5, that is, aged only. The material reported on in our paper was EK31A-T6. This material has

superior tensile properties to EK31A-T5.

The effect of varying rare earth content in alloy EZ33A has been checked. No effect was discerned as the rare earth content was increased. We have not checked the effect of varying zinc and so can not comment further on Mr. Jessup's remarks in

that regard.

L. W. Eastwood (Written Discussion): ^a The paper comparing the casting characteristics and properties of several magnesium rare earth sand casting alloys by K. E. Neison and P. F. Strieter is based on the results of a great deal of work and represents a real contribution to the light-alloy foundry technology. There are several points which I would like to make regarding this

рарет:

1. The present operating temperature for light-alloy parts in aircraft motors is approximately 400 F, as the authors point out. Alloys which provide good load-carrying capacity at this temperature present no real problem. The authors failed to point out that the alloy referred to in their paper by the ASTM designation EM-61B-T6 was designed primarily for operating at higher temperatures up to 600 F. We would not recommend it if the operating temperature were only 400 F. An alloy similar to EM-61B-T6 will be of interest when the operating temperatures reach 500 to 600 F. As the authors observed, this alloy has higher load-carrying capacity at 400 F than the other alloys, but real superiority is markedly evident when the service temperature is 600 F.

2. The gas porosity and trapped gasholes found in the EM-61B-T6 alloy castings and the unsoundness in the separately cast test bars indicate that both melting practice and gating practice require improvement. No information is given on gating or pouring practices or on pouring temperatures. Melting and gating practices which are suitable for one alloy may not be suitable for another. Thus, foundry practices suitable for aluminum with 5 per cent silicon might be quite satisfactory, but the same practices would be disastrous for an excellent alloy such as aluminum with 10 per cent magnesium. We have had considerable success in melting magnesium alloys containing rare earths by doing it under a nitrogen cover but without using fluxes. This has been done on relatively small melts only. Nitrogen is easy to use, it is cheap, and there is a good possibility that the method would be practicable for large-scale melting operations. Although no large castings were made from melts prepared in this manner, several different castings were made without the occurrence of appreciable segregation or other found r_{γ} defects.

Also related to foundry practice, particularly melting methods and pouring temperature, is the considerable amount of segregation of the magnesium-cerium compound in the castings made by the authors. The evolution of gas during solidification may accentuate the segregation and may account for the segregation obtained. This defect may be the cause of the lower tensile properties and lower ductility obtained in the Dow laboratories, as compared with the results obtained at Battelle and in other laboratories.

The authors stress the low elongation obtained in the tension tests. Because the tensile elongation is low, they consider the alloy to be inferior. It is interesting to observe that the outstanding light alloy employed for elevated-temperature service is the one known as "Y" alloy, which has been in service for over 30 years. This alloy, like many others of the outstanding heat-treated and aged aluminum-base alloys, has a tensile elongation too low to measure. If the same reasoning were applied to "Y" alloy, it would have been discarded before it was put into production.

In elevated-temperature service, resistance to failure by creep is of outstanding importance. The greater load-carrying capacity of the EM-61B-T6 alloy, ac illustrated by Fig. 16 in the authors' paper, does not appear to have been emphasized.

K. E. Nelson (Authors' Written Reply to Dr. Eastwood): Comments such as Dr. Eastwood's must be seriously considered when a new composition is being developed as a production sand casting alloy. There should be adequate reasons given for con-

sidering any one composition over another.

It is true that this evaluation was primarily for alloys considered for use at 400 F with less emphasis on their load-carrying capacity at 500 or 600 F. If load-carrying capacity were the only factor to be considered, then, without a doubt, compositions of the type EM61B-T6 would be desired at 400 or 600 F as indicated in Fig. 11, 12, 15 and 16 and in the text of the paper. As previously presented, and as we shall show in more detail in this discussion, other factors must be considered.

It is possible that a modification of gating and risering and/or melting practice might reduce or eliminate the voids found in the EM61B alloy. This very interesting phase of work was not a part of our program. Although specific details were omitted, it was stated in the paper that gating and risering of the casting as well as pouring temperature "was the same for each composition and was the one used by the foundry in pouring EK30A alloy production castings." The pouring temperature and pouring time for each casting was reported in Table 3 of the paper.

The authors do not believe that the gas referred to by Dr. Eastwood affected the segregation of compound and thus the ductility of alloy EM61B. Metallographic studies of this alloy indicated areas of highly segregated compound in porosity-free areas. The authors feel that the gas referred to above has been entrapped by the turbulence of the metal stream and is not a gas coming out of solution during solidification.

The authors believe it shou'd be pointed out that, as reported in the paper, there were bars radiographically free of defects which were so brittle that yield strength values could not be obtained. Another fact which should be mentioned is that the authors have, subsequent to the original program, again attempted to cast EM61B alloy in the thin-walled casting used in this evaluation. In four attempts at casting this part, the authors have encountered cracked castings, formed during solidification, three times. The fact that the other compositions were not as susceptible to cracks indicates EM61B alloy is less castable.

In arriving at our present suggested alloy system of Mg + 3% rare earth + zirconium the authors have been aware of improved properties in the higher rare earth-manganese alloys. They have attempted to pour production castings in many alloy compositions including the following:

Mg + 6% Rare Earth + 2% Mn Mg + 6% Rare Earth + 2% Mn + Zr Mg + 6% Rare Earth + 1% Mn + Zr Mg + 6% Rare Earth + 0.25% Zr Mg + 3% Rare Earth + 0.25% Zr

Colum-It has been observed that reducing the Mn content and adding a small amount of Zr decreased handling problems in the foundry

 $^{^{9}\,\}mathrm{Supervisor},$ Nonferrous Metallurgy, Battelle Memorial Institute, Columbus, Ohio.

de

a

Ott

ile

es.

ier

oa he

at-

is

ver

at-

on

ro

TS

d):

ed

nd

n.

id

ng

aly

of

di-

As

his

or

in

t a

it

ng

0A

ur-

er.

Dr.

he

loy

ree

een

ta

ted

rets

be

the

in

he

di-

ere

st-

of

and increased the ductility of the part by decreasing its grain size. In this operation, however, the elevated temperature yield and tensile strength values and the creep properties of these alloys were reduced. Lowering the rare earth content from 6 per cent to 3 per cent, omitting the Mn, and increasing the Zr content to 0.25 per cent resulted in a reduction in the amount of embrittling compound which was present and reduced the grain size of the alloy considerably. It is realized that a compromise was made between properties and the ability of the alloy to be handled in the foundry and assembly operations. The practicability of a production foundry producing parts in a brittle composition such as EM61B alloy is not encouraging.

M. W. MARTINSON AND L. BADONE (Written Discussion): Mr. Nelson and Mr. Strieter are to be congratulated on an excellent paper which should be invaluable to aircraft engine designers working with the newer high-temperature magnesium alloys. A lot of data has been published for the rare-earth alloys, based on various types of separately cast test bars, but until this paper very little data was available on the properties to be expected in actual castings.

Table 4 in the paper illustrates the well-known foundry fact that each casting is a problem unto itself. For instance, castings B and C, on all alloys, give slightly lower room temperature properties than castings A and D. Thus any statements on the proportion of separately cast test bar properties obtainable in castings should always be strongly qualified, as they can never be anything but generalizations.

The five alloys dealt with in this paper do, however, show the same trend within any one type of casting, and referring to Fig. 10 one sees that the overall trends are the same when either casting properties or separately cast test bar properties are considered.

In Renfrew we have been working for some time with these alloys, and in particular with (EZ33A-T5). The slightly inferior showing of this alloy in the present work, when compared with EK30A, EK30B and EK31A, we believe, is due in part to the lower rare earth content used. The following table compares the average. I room temperature properties on separately cast test bars from some of our production melts of EZ33A with the properties taken from Fig. 10 shown for similar bars for the four manganese-free alloys. The rare earth analyses for the alloys dealt with in the paper, have been averaged and are also shown for comparison purposes.

Alloy & Condition 1	R.E. %	U.T.S.	0.2% P.S.	E1 %
EK 30A-T6	2.88	23.0	14.0	5.0
EK 30B-T6	3.07	26.0	17.5	3.0
EK 31A-T6	2.95	25.0	16.0	5.0
EZ 33A-T5	. 2.51	22.0	16.0	3.0
EZ 33A-T5 (LAL Prodn	3.1	25.0	16.0	4.7

The averaged production melts fell within a narrow analysis range of R.E. 3.1 ± 0.1 per cent; Zr 3.1 ± 0.1 per cent; Zr (S) 0.6 ± 0.05 per cent, and at this level of rare earths the creep and elevated temperature properties are probably much closer to those of the other three alloys.

On a casting similar to type B used in the investigation, we have obtained, on bars cut from the casting and tested at room temperature, U.T.S.—19.9; 0.2 per cent P.S.—13.1; E1, per cent—2.9 per cent, the sectioned castings coming from normal production runs. These properties come very close to the ones given in the paper.

As regards foundry characteristics, these alloys are definitely inferior to the AZ alloys. They are more turbulence prone, and this necessitates careful gating if the cope pitting mentioned by the authors is to be avoided. Feeding on the alloys is good but solidification shrinkage is dependent on composition, and where close dimensional tolerances have to be met close control of composition is necessary. Work with fluidity test spirals has indicated that zinc is beneficial. It has also been noted that property levels tend to be higher when scrap is being re-used than when charges are entirely composed of new metal.

In evaluating the rare earth alloys it must be remembered that they all have far superior elevated temperature properties to the older alloys and the differences between EK30A, EK30B, EK31A and EZ33A are minor when compared with, say AZ91. Considered from this point of view, the slight sacrifice in elevated temperature properties incurred by the use of EZ33A is well justified when a high temperature solution treatment is avoided.

We would like to query this solution treatment as on castings of type B distortion must be an ever present danger and inevitably lead to slower rates of production.

MR. NELSON: I am happy to see the interest of Messrs. Martinson and Badone in our paper and to have the opportunity of this exchange in comments.

It is a fact that the final evaluation of alloys must be made in actual service tests, or, second best, on bars machined from castings. Equipment for service testing is not readily available and in order to facilitate our comparison of alloys we selected the second method.

The comparison made by Messrs. Martinson and Badone shows the yield strength values of separately cast bars of both Dow and Light Alloys Limited production EZ33A-T5 to be the same but with higher elongations and tensile strengths in the higher rare earth content alloy. We have not obtained elongations and tensile strengths in separately cast bars of this composition with equal rare earth contents as high as they have quoted. As stated in their discussion their data on bars sectioned from castings compare exactly with our results with our somewhat lower rare earth contents.

The observation that highway property levels are obtained on re-used charges than on virgin charges is interesting.

In answer to the query on distortion during heat treatment we find the distortion of EK30 type of alloy heat treated at 1050 F is not serious, nor is it any different than obtained on solution treating Mg-Al-Zn alloys at their normal solution treating temperatures of 730 to 780 F.

W. E. SICHA: 4 Have the authors made any effort to determine the importance of grain size in the good resistance to creep that they obtained with EM61 alloy? How much is attributable to grain size and what portion to alloy composition?

MR. NELSON: The general belief in regard to creep resistance of alloys as a function of their grain size is that the coarse grained alloy is stronger than the fine grained alloy. This factor has some bearing on the improved creep strength of EM61B alloy over the other alloys reported in this paper. However, in addition to this we have seen in our work at Dow that the higher rare earth content (6 per cent over 3 per cent) contributes to a greater degree to the good creep resistance of this type of alloy than does its coarse grain.

In the magnesium-rare earth-zirconium alloy system we have found essentially no variation in creep strength between 0.0 per cent Zr and 0.6 per cent Zr in spite of a considerable refinement in the grain size and improvement in the tensile properties as the result of the Zr addition.

⁸ Production Manager and Plant Metallurgist, respectively, Light Alloys, Ltd., Renfrew, Ont., Canada.

⁴ Chief, Cleveland Research Div., Aluminum Co. of America, Cleveland.

LADLE REFRACTORIES AND PRACTICE IN ACID ELECTRIC STEEL FOUNDRY

By Clyde H. Wyman*

Introduction

IN THE CASTING OF METALS the foundryman endeavors to obtain the cleanest metal possible at the ladle spout. This can be accomplished only by maintaining ladles free from furnace and ladle slags. From a pouring standpoint, a fluid slag is the Cleaning Room's worst offender. In addition to slag surface defects, there is also a great possibility of trapping silica particles in the steel thereby creating what is known as an accidental nonmetallic inclusion.

The silicious material known as ganister is commonly used in ladle linings in the acid electric steel foundries. This material is not processed by the refractory manufacturers but is used in the raw state and prepared and graded for the various service conditions.

An understanding of the reactions of silicious materials as refractories revolves around a study of silica, the basic constituent of this material. Quoted from bulletin on Massive Quartzite by A. R. Blackburn of the Ohio State University, silica is polymorphic having two or more crystalline forms and exists in more than one mineral form. These are alpha and beta quartz, and alpha and beta crystobalite and alpha and beta tridymite. At atmospheric temperatures these minerals exist only in the alpha form but invert rapidly to the beta form when heat is applied. Any one of these minerals may be converted into either of the others by proper heat treatment. The permanent expansion of silica when heated and the reversible expansion with heating and cooling are due to these inverting properties. With normal service, the quartz inverts to both crystobalite and tridymite and will consist of all three silica minerals plus silica glass. If the temperature does not exceed 2678 F, it will be composed of tridymite silica glass and impurities. If the temperature does exceed 2678 F, it will become crystobalite with silica glass and impurities. These inversions only take place on exposed surfaces.

Chemical analysis of ganister is as follows:

Silica,	SiO ₃	98.82%
Ferric Oxide,	Fe ₂ O ₃	0.09%
Alumina,	Al ₂ O ₃	0.73%
Magnesia,	MgO	0.01%
Titania,	TiO,	0.02%
Total Alkalies		0.03%
Ignition Loss		0.30%

Materials

The greater proportion of ganister that is consumed by the foundry industry originates from three different districts—Eastern, Middle West, and Western. The material described in this paper is South Dakota ganister. According to geological reports, the material is bed rock belonging to the Sioux formation and outcrops along the valley of the Big Sioux River. The exposures are very hard and are practically pure quartz.

Several years ago, the author conducted an intensive investigation upon the effect of grain size and grain distribution of the material used in a monolithic rammed ladle lining. The principal objective was to obtain the highest degree of refractoriness with a minimum of erosion which is the determining factor for maximum life and clean ladle walls. Our approach was from two distinctive phases; first, density and second, the correct silica and alumina-silicate or clay ratios. The ratio will determine the degree of vitrification or a fusion zone. It is our belief that the closer you can come to a solid mass on the exposed ladle wall, the greater efficiency you can expect from the operation of the ladles. In order to obtain this degree of density from the ganister mix, it is necessary to maintain a definite grain size with proper grain distribution which will create the desired interlocking effect of the grains and the absence of voids to resist and prevent a starting point for undercutting and erosion of the wall.

After thorough investigation we have adopted the following fineness specification which has given satisfactory service:

^{*} Chief Metallurgist, Burnside Steel Foundry Co., Chicago.

n-

ree est-

th he

on

er.

en-

nd

10-

ve

th

ac-

ur

ns-

ite

of

at

X-

ect

in

is

er

er-

ds

ng

he

is-

-64

REPRESENTATIVE FINENESS

Sieve No.	Per cent
20	11.6
30	5.7
40	8.2
50	17.9
70	19.3
100	13.8
140	5.7
200	3.5
270	1.4
Pan	3.9

The above fineness is classified as "1/16 and down" material; the wide distribution is appa.ent. Taking the A.F.S. fineness numbers into consideration, over a period of time, the cars of ganister will average from 54 to 60. To obtain the desired density and permeability, we request the material to run 40 per cent reclaimed on a 50-mesh screen and the balance to pass through to the pan. Our ladle performance records of the number of heats per lining clearly show that an average allowance of 3 per cent plus or minus at the 50 per cent desired graduation point will produce satisfactory results from a density standpoint.

In steel foundries where various type of ladles are employed, which include teapot and bottom pour and covered monorail ladles, it is essential to know through experimentation, the thermochemical reaction and the eutectic point of the ladle mixtures. This must be accomplished to obtain the desired homogeneity resulting in the desired monolithic structure. This depends on the subjected temperatures and service on the exposed surfaces of the ladle wall.

Four Ladle Mixtures Tried

Through observation and experimentation, we have adopted four different ladle mixtures depending on the application and the heat effected zones. The use of these various mixtures is determined by the degree of temperature where the particular location of the ladle is subjected to, and also the rate of heat conductivity from that particular location. High refractoriness is not soley important in ladle linings but, to arrive at the proper silica-clay ratio in combination with the fluxing ingredient, it is necessary to reach the eutectic point of the mix depending upon the temperature range to create thermochemically the degree and depth of the vitrification or fusion strata. For illustration, we carried on an investigation of eight different materials used in the spout of the teapot ladles. Four of these mixtures were milled in our own plant and the balance were purchased items. Following the day's operation, the inside diameter of the down and in spout was measured to determine the degree of erosion and also the condition of the wall surface. Taking all the different materials into consideration under approximately the same service conditions, the total increase in area ranged from 28 per cent to 340 per cent. The material that showed the most homogeneous mass and the deepest vitrification zone had the lowest percentage of area increase in the spout opening.

Properties and Mixes

The service testing of the ganister was carried out with reference to its properties when mixed with the binding materials and the suitability of these mixtures to ladle service. As stated before, we have adopted four ladle mixtures, each designed for a specific purpose depending upon the service requirements of the ladle.

The writer calls your attention to the fact that he uses a larger percentage of fire-clay in the mixes; first, to maintain a closer control of the permanent expansion of the silica and, second, the alumina-silicate in combination with the silica to obtain the desired eutectic mixture. A small percentage of bentonite is added to increase the green strength with also a slight effect upon the hot strength and added plasticity without much detrimental effect upon the permanent expansion. About 1 per cent of iron oxide and fly ash is introduced in the mix as a fluxing and glazing agent to accelerate and act as a mineralizer to insure thermochemical reaction and deepen the vitrification zone.

The sidewall mix used in the teapot and the small bottom pour ladles consists of 89.3 per cent ganister, 9.5 per cent bonding materials, and 1.2 per cent fluxing agent.

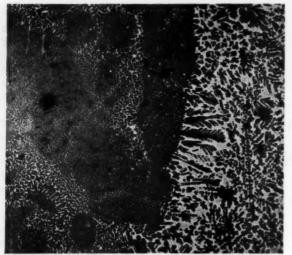
MIXTURE FOR SIDEWALL OF 5-TON TEAPOT LADLE

Ganistet	600	lb	89.3%
Fire Clay	41/2	gal-43.2 lb	6.4%
Iron Oxide	3/4	gal- 8.1 "	1.2%
Bentonite	11/2	gal-10.8 "	1.6%
Fly Ash	/-	10.0 "	1.5%
Glutrin	134	gal	

Properties derived from the above mix are as follows:

Moisture	6.7%
Permeability	40
Green Compressive Strength	11.1 psi
Dry Tensile Strength	105 "
Hot Strength at 2000 F for 12 min	690 "
Hot Strength at 2500 F for 12 min	32 "

The temperature exposure and the erosion is greater in the spout of the teapot ladle than at any other location. The factor of heat gradient is low due to the fact that only a refractory wall separates the spout from the main body of the ladle. In this mixture, due to the extreme service, we use the silica in a very fine grain size to obtain a high degree of refractoriness and also the maximum degree of density. The silica content is 87.8 per cent, bonding material 11 per cent, and the fluxing ingredient 1.2 per cent. Through experimentation, we run the spout for approximately 25 heats before slagging commences, then the spout is replaced. In this period of time the vitrification strata will extend from 1 to 11/2 in. deep with no visual indication of a crystalline structure, but a perfectly blended solid mass.



Hot Face Zone Su.tace Adhering Slag

Hot Face Zone. There was progressive alteration of the
quartz to cristobalite toward the hot face zone. In the latter, the
former quartz grains were still identifiable, but were completely

altered. The matrix contains a pyroxene glass.

Adhering Slag was ½6 to ½ in. thick, dendritic cristobalite and pyroxene.

Fig. 1-Sidewall refractory. The unaltered material as well as the raw ganister were quartz. Mag. 60x.

The third mixture is applied in the form of a slurry approximately 3/4 to 1 in. thick as a coating over the brick lining in the large bottom-pour lad'es. This affords good protection for the brick and normally has a life of from 30 to 40 heats before chipping out and recoating. The bottoms and the wall surrounding the nozzle are rammed in with the same mix we use in the side walls of the teapot ladles.

The coating mix carries practically the same ad-

ditions as the spout mixture with the exception of a larger amount of fireclay. Due to the manner of application, it is necessary to increase the cohesiveness and plasticity of the material.

MIXTURE FOR SPOUT OF TEAPOT LADLE

Silica Sand, Washed A.F.A. 60		450	lb	65.89
Silica Flour, 250 Mesh		150	20	22.00
Fire Clay	51/2 gal -	52.8	**	7.79
Bentonite	13/4 " -	12.6	23	1.80
Iron Oxide	3/4 gal -	8.1	919	1.20
Fly Ash	1 gal -	10.0	**	1.50
Straight Glutrin	1 gal			

Properties

Moisture	6.3	%
Permeability	30	, .
Green Compressive Strength	9.6	psi
Dry Tensile Strength	120	22
Hot Strength at 2000 F for 12 min	850	27
Hot Strength at 2500 F for 12 min	110	22

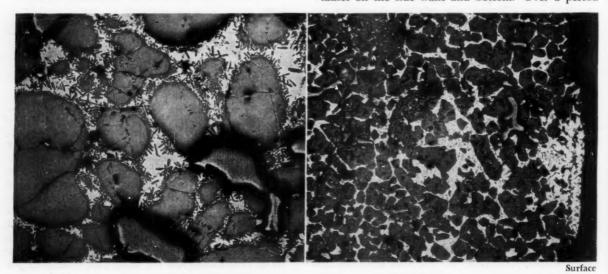
SLURRY MIXTURE FOR 5-TON BOTTOM-POUR LADLE

Silica Sand, Washed A.F.A. 60	45	0 11	64.5%
Silica Flour, 250 Mesh	15	0 "	21.5%
Fire Clay	7 gal -	67.2	9.6%
Bentonite	13/4 gal -	12.6	1.8%
Iron Oxide	3/4 gal -	8.1 "	1.2%
Fly Ash	l gal -	10.0 "	1.4%
Glutrin	1 gal		

Propertie

Moisture	6.90	%-As Milled
Permeability	25	
Green Compressive Strength	9.9	psi
Dry Tensile Strength	100	27
Hot Strength at 2000 F for 12 min	880	29
Hot Strength at 2500 F for 12 min	130	**

The fourth mixture is used in the 250-lb monorail ladles on the side walls and bottom. Over a period



Black Zone, 1 to 2 in. thick, quartz grains surrounded by a glass containing tridymite and a fine-grained pyroxene.

Gray Zone, 1/8 to 3/8 in. thick. All of the quartz grains have been altered to smaller cristobalite grains. The matrix contains cristobalite and pyroxene. Near surface, magnetite is present.

Fig. 2-Spout refractory. Unaltered refractory was predominantly quartz. Mag. 80x.

of a

ap-

1055

.00

of several years, we conducted an investigation with ladle linings for the sole purpose to find a mix which would minimize slag formation accumulated in the monora'l ladle in pouring of the lighter castings. Through this investigation, we used all types of silicious materials and even experimented with some basic linings. Grain size and grain distribution should be emphasized. The following permeability and hot strength figures which we obtained in this mix are very desirable. Also, the eutectic point of the different compounds is well within the pouring temperature range of the metal, thus setting up a definite vitrification zone which is also necessary to maintain a low slag resistant value.

MIXTURE FOR 300-LB MONORAIL LADLES

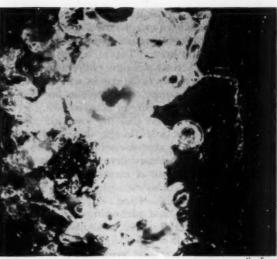
Ganister						2	300	lb	48.49
Car-O-Line			1			9	300	22	48.49
Fire Clay				1	gal	-	9.6	2.5	1.60
Fly Ash				1	gal	-	10.0	**	1.69
Glutrin			1	1/2	gal				
	Pr	ope	erti	es					
Moisture								6.7	1%
Permeability								40	
Green Compressive S	itre	ngth	1					9.5	B psi
Dry Tensile Strength		-						95	21
Hot Strength at 2000	F	for	12	m	in			740	9-9
Hot Strength at 2500	F	for	12	m	in			115	**

In all the mixtures 1 to 11/2 gal of glutrin is added as a tempering and binding agent. This liquid material has a great effect upon the green and hot properties and also enhances the refractory values of the

Installation

The practice of building the ladles is rather universal throughout the small casting industry. We employ specially built steel forms on the teapot, small bottom-pour, and monorail ladles. This form is designed with a slight taper depending upon the capacity of the ladle to permit the ladle man to withdraw the form after ramming and avoid any tearing or disturbance upon the ladle wall. The degree of ramming to obtain the expected monolithic mass is determined by the depth of the impression of the impact of the pneumatic rammer. This impression, when the degree of ramming is adequate, shows just a slight indication of the ramming head. To further insure soundness after the form is drawn on the teapot ladle, the operator will again ram the side walls in a horizontal position. It is quite surprising the soft spots one will encounter even after taking painstaking efforts on the first ramming operation. These soft spots usually occur at the parting lines of the different additions of the mix, even though the ram surface between additions is thoroughly scratched to allow the blending of the material.

For added protection, two rows of straight fire brick are laid in with cemented joints on the bottom of the teapot ladle, and the balance of the bottom and



Altered Zone. The decarburized zone is 1/4 to 1/2 in. thick. The glazed surface is 1/16 in. thick. The latter is a vesicular siliceous glass containing Al2O3, CaO and MnO. Conical illumination. Mag. 40x.

Fig. 3-Monorail refractory. Unaltered material was quartz with carbon and some clay-like material.

side walls and spout are rammed in place. On the small bottom-pour and monorail ladles, both the bottom and side walls are made up of a monolithic lin-

The practice on the large bottom-pour ladles is to lay two courses of brick on the bottom, and one course on the side walls. Approximately a 5-in. rammed bottom is installed depending upon the height of the nozzle in order to allow for proper drainage. As explained before, a layer of slurry is applied to the side

All types of ladle shells are perforated with 1/4 to 3/8-in. vent holes placed on from 2 to 5-in. centers depending upon the size of the ladle.

Drying and Preheating

We all strive to produce a quality product. Many a good heat of steel tapped into a so-called "wet" ladle has resulted in inferior and unsalable castings. Whenever pear-shaped holes or cavities appear surrounding the riser and also on the cope surface, most of the time, you can trace it down to a wet ladle. When the metal is poured into insufficiently dried ladles, water vapor is formed from the excess moisture, which in turn dissociates into hydrogen and oxygen and may be trapped in the casting surface forming deep-cratered cavities.

We are firm believers in ample drying and proheating facilities in order to avoid the possibility of a green or wet ladle. Our procedure to obtain a thoroughly dried ladle is taken in three steps. When the lad'e is newly lined, the first step is to apply heat from a forced draft, pre-mix gas combustion chamber, which circulates the heat efficient'y. This preliminary drying is carried on for an 8 to 10-hr period at a temperature of from 500 to 600 F. In the second step, the ladle is placed in a horizontal position

against the ladle wall or in a vertical position using a gas burner with a low air pressure, and the flame is set to maintain a temperature of from 1000 to 1200

F for a period of 8 hr.

Following the second preliminary drying operation, the ladle is now ready to be put in use. Before the ladle receives its first heat, we allow 3 hr for the final preheating at a temperature of between 1600 to 1800 F. It is well established that the lower the density maintained in the ladle linings, the more caution must be exercised in also maintaining a definite rate of heat in the preliminary drying of the ladles. If this three-step program is carried out as outlined, the control of permanent expansion is within bounds and no indication of buckling or spalling occurs.

Our ladle drying and preheating facilities consist of a low-temperature gas-fired circulating-type heater. To accommodate the teapot and large bottom-pour ladles, we have three-wall preheaters with a ladle truck on tracks in front of each wall in order to maintain the proper space between the ladle proper and wall. From a standpoint of heat conservation, this opening is important. On the small bottom-pour ladles, we have practically the same wall setup only on a smaller scale. This wall is continuous with three gas burners used for preliminary drying and final preheating. The monorail ladles are preheated in a vertical position with room for six ladles on a roller conveyor. Each ladle is supplied with an individual burner and as the front ladle is p'aced in service, a new lad'e is placed on the line thus allowing sufficient time for the drying operation. An average of two ladles is used on each heat so the cycle permits a drying time of approximately 4 to 41/2 hr. Allowances are also made for air drying of these ladles before subjected to heating to avoid any damage to the ladle wall.

Distribution

The distribution and handling of metal in the steel casting industry determine the type of ladle to be used. This is governed by the type of mold and casting design which, through foundry experience and technique, determine the proper rate and speed of pouring at a temperature with a given flowa' il'ty to produce a casting with a minimum of the usual foundry defects caused from the pouring operation.

The average jobbing electric steel foundry producing castings weighing from several ounces to several thousand pounds should be equipped with various types of ladles. It is quite obvious that the lighter weight casting should be poured with hand shank or monorail ladles. Proceeding in the next weight class, it is universally conceded among foundrymen that the bottom-pour ladle is superior from a cleanliness and

sharp casting detail standpoint.

Usually monolithic ladle linings are prepared, refractory mixes blended and milled in the foundryman's own plant. Practically all refractories used in a bottom-pour ladle are manufactured items. The selection of brick, sleeves, nozzles and stopper heads is based upon the performance of the ladle operation. This requires a careful study of the composition and design of the refractories which are best suited for one's pouring practice.

There is a difference of opinion as to the design and shape of the nozzle and stopper head. Since the advent of the bottom-pour ladle, almost every conceivable design of stopper head, sleeve brick and nozzle has been tried in the steel industry. It is a known fact that U. S. patent files alone contain some 90 claims for variations in these parts.

Bottom-Pour Ladle Performance

For successful bottom-pour ladle performance, it is essential that the ladle man be a sincere conscientious individual. Not the slightest detail can be missed or overlooked in the maintenance and preparation of the well, together with the stopper rod set-up of the ladle.

Stoppers and nozzles are not complicated in design; their function is to shut and open the ladle metal pouring outlet. Like any valve, the functioning of the stopper head-nozzle combination depends upon one harder surface impinging on a softer one.

Experience has shown that the best nozzle brick is made of third grade fire clay with a softening point of 2600 to 2700 F. The nozzles provide the seat for the stopper head, therefore, it should soften slightly at steel pouring temperatures. There exists a difference of opinion as to the proper design, which is usually based upon one's pouring practice and size of heat. Practically all nozzles are designed with a streamlined orifice, curving from top with an ample radius at the seating surface. In order to obtain a good circular contact, it is well to place the stopper head upon the nozzle and turn completely around to observe if the marking from the graphite head has formed a circle upon the seat of nozzle. For high temperature pouring, a high seat is desirable and will minimize sticking difficulties.

Stopper heads take the brunt of the pouring load, must resist high temperatures and be able to maintain sufficient compressive strength at these temperatures. Through years of experimentation, stopper head manufacturers have found that natural graphite with suitable clay binders seems to be the best material from a service standpoint. The high rate of heat conductivity enables the head to resist the thermal shock when it comes in contact with the molten steel. The size of the ladle and pouring temperature determine the composition of the head. Some operators prefer a head with a lower carbon content, which lowers the rate of heat conductivity to other critical parts of the assembly. Depending upon the burning temperature of the heads, the carbon content determines the refractory point, the hardness and the resistance to thermal shock.

The head is assembled with the rod by means of a bolt or pin as shown in the stopper and nozzle assembly (Fig. 4). In order to avoid contact between the walls of the stopper head and the stopper bo't, some operators, as an extra precaution, used a silicious clay around the stopper bolt to form an insulating layer to reduce the rate of heat transfer to the vital bolt.

gn

he

n

nd

ne

is

us

of

le

1:

al

of

n

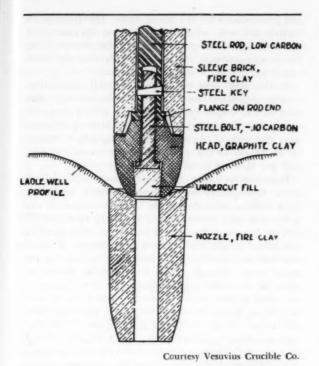


Fig. 4-Stopper and noz 'e assembly.

After the pin and head are keyed in the rod, care should be exercised not to drive the pin too tight in order to avoid damage to the shoulders of the head.

The next step in the rod assembly is to plug the undercut in the stopper head. Considering the extreme exposure of this area in service, a high refractory, low shrinkage material should be used. Stopper head manufacturers produce a plug mix using a graphite-bonded mixture which has proven satisfactory. For economy, some operators grind used graphite stopper heads and use a ratio of two-thirds of ground heads and one-third fire clay, which produces a good plug mix. In the ramming of the plug mix in the undercut area of the head, it is good practice to work the mix in well with fingers and then supply sufficient material to complete ramming in one operation. This precaution will produce a homogeneous plug free of laminations.

With the head attached to the rod, it is now ready for the upper protection with a series of clay sleeves covering the stopper rod. Sleeves like nozzles are usually made of third grade fire-clay. Sleeves must be able to withstand thermal shock together with low linear thermal expansion. In addition to the above desired properties, the resistance to spalling, cracking and slag erosion with a high pyrometric cone value are all necessary requisites in order to maintain satisfactory ladle operation. Considering the stopper assembly, the sleeves carry the least burden from a mechanical load standpoint and only serve as a protective covering of the stopper rod.

Designs of sleeves are fairly well standardized, most

commonly adopted of which consist of two types called the bell or bottom, and straight sleeves. The bell sleeve is designed to fit the shank of the stopper head with a skirt extending around the upper periphery of the head. This feature is added protection and serves as an obstruction in preventing the molten metal from reaching the vital stopper bolt.

Stopper Rod

The size or diameter of the stopper rod determines the size straight sleeve to be used. For example, in using 13/4-in. rod, it is common practice to use a 2-in. i.d. and 4-in. o.d. sleeve, thus allowing a 1-in. refractory wall surrounding the rod.

In the stopper rod assembly, it was previously stated that the head had been attached to the rod. The next step is the placement of the required number of sleeves on the rod to just below the threaded section of the rod. Following this operation, the rod is suspended by resting on the washer and nut, which is placed at the bottom of the threaded section.

It is important that the shank or male of the stopper head and the female part of the bottom or bell sleeve be of the same design so that there is little clearance between them. This particular phase is the heart of the assembly. On the smaller rods, the loose sleeves are lifted while the bottom sleeve is mudded into the head. It is essential that a high fusion and low shrinkage mind mixture be used between the joints. Each succeeding sleeve is joined with the same careful consideration.

There exists a small opening between the rod and the sleeves. It is considered and practice to pour loose silica sand in this gap, thus reducing the rate of heat conductivity to the rod.

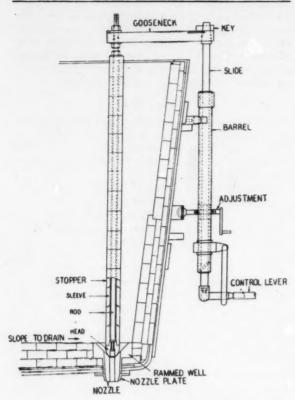
In order to assure oneself, when the joining of the sleeves is completed, the nut arrangement with the bottom washer is slightly tightened to exert a pressure on the sleeve assembly, thus assuring a compact snug fit between the joints.

Drying Operation

Before the drying operation, the nut should be backed off slightly to permit expansion of the sleeve assembly. Too much emphasis cannot be placed upon the drying operation. Like in all refractory and steel assemblies, a slow drying procedure is absolutely necessary to prevent damage to the refractories through the expansion of the materials.

Ample time at low temperatures is an important factor to obtain a moisture free assembly.

Figure 5 shows a cross-section of the pouring ladle with the stopper rod assembly. The goose neck is keyed in the slide operating mechanism and the adjusting screw is moved for proper placement of head in relation to the nozzle. In setting the rod in the ladle, it is set so that the head is slightly off center towards the near side of the ladle. This slight kick is necessary to compensate for the buoyancy of the steel and the expansion of the rod so that when the ladle is filled, it brings the head directly over the nozzle. When setting the rod assembly, at least ½6 in. per ft of sleeves should be allowed for expansion



Courtesy Vesuvius Crucible Co.

Fig. 5-Pouring ladle.

to prevent cracking of the sleeves and undue pressure on the stopper head.

To further test the contact point between the head and nozzle seat, a pan should be placed directly under the nozzle opening and dried sand sprinkled around the head to determine if any sand grains seep or pass through the contact area. This practice is not an absolute cure-all against leakers, but will give one a good indication of the setup.

A great deal of investigation has been carried out on the thermal distribution of stopper rod assemblies. For efficient bottom-pour ladle practice, every precaution must be taken to minimize the rate of heat conductivity to the stopper rod and bolt.

The refractory bottom of the ladle should be sloping to the nozzle and the surrounding well. This well should be broad and shallow in order to prevent chilling of the metal around the stopper. In the setting of the nozzle, first examine the condition of nozzle plate which must be perfectly clean with a perfectly flat surface in order to maintain the nozzle in a perpendicular position. For added protection to insure perfect alignment, a mandrel with a head fitting the nozzle and a shaft extending into the nozzle opening should be used; the opposite end should be threaded like the stopper rod. This mandrel should be assembled exactly like the stopper rod assembly and connected with the goose neck. The practice in making the well, which is a ramming operation, requires a rigid setting in order to hold the nozzle in the proper position to meet the radius of the head which constitutes a valve arrangement.

In the acid steel foundries, the well material is usually a ganister mix similar to the sidewall mix used in the teapot ladle. After the well is rammed, the surface should be troweled to a sloping effect to meet the sides of the nozzle. Care must be exercised in the drying of the well material to avoid cracking and any thermal shock to the nozzle.

Bottom-pour ladles are usually lined with a high grade first quality hard burnt fire brick, including both bottom and side walls. In the laying of the brick, it is advisable to imbed the brick in a layer of wet mud next to the shell of the ladle. This practice will afford a tight seal between the joints of the brick and will add another layer of protection in case the metal seeps through between the brick. From experience, to obtain a prolonged ladle life, a slurry of wet mud approximately 3/4 to 1 in. thick on the surface of the side wall brick will protect the brick from the erosion of the metal and slag. In case of the bottom, a rammed ganister lining is put in place from 4 to 5 in. in thickness. Based upon ladle performance, the addition of the slurry and the bottom ganister has greatly increased the life of the brick.

Other Factors

The life expectancy of the ladles aside from the effect of slag is determined by other factors, such as, type of steel, deoxidation products and time consumed in pouring heat. From experience, we all fully realize that the pearlitic manganese steel series have an erosive action upon the acid-lined ladles probably caused from a metal and slag reaction. Secondly, the calcium deoxidation products react with the acid lining, causing a chemical reaction which produces the undesirable thin fluid slag which eventually lodges in the cope surface of our castings and adds to the Cleaning Room cost. Third, the holding time, or time of exposure at high temperatures beyond the softening point of the ladle lining, will have a pronounced effect upon the life of the installation.

Conclusion

In the steel foundry industry, the study of the behavior and characteristics of refractory materials of monolithic linings is interesting and also mystifying. With supposedly the same identical service condition and the same refractory materials, you se dom see where two foundries can accomplish the same results. It is quite obvious there is an unknown factor which enters into the picture. Some say the raw materials are not consistent; others remark it is in the preparation of the materials, or the method of installation might be responsible; still another phase, which we at times are not aware of, is the thermochemical reaction which takes place and causes undue deterioration.

Side stepping all the variables which might be encountered, the author approached the problem to den

d

is

l,

0

d

CF CS

g

e

termine the effect of grain size and grain distribution of the ganister in order to increase the life of the ladles, which means a reduction in ladle cost and, more important, cleaner metal at the ladle spout. This investigation was carried through over an extensive period of time starting with the coarse ganister specified as 1/4-in. and down and eventually decreasing the grain size to 40 per cent reclaimed on a 50-mesh screen, and the balance through to the pan. This material could be classified as 1/16-in. and down.

Results indicate, from a service standpoint, that the size of the grain aids in the promotion of the desired density. Based upon the performance of the teapot and small bottom-pour ladles, we have more than doubled the life of the ladle linings. When this investigation was started, the author was fortunate to obtain 40 to 60 heats from a lining. Today, as the result of our findings, we have raised our average from 200 to 250 heats, and also an improvement from a cleanliness standpoint.

DISCUSSION

Chairman: R. H. STONE, Vesuvius Crucible Co., Pittsburgh. Co-Chairmen: R. P. SCHAUSS, Illinois Clay Products Co., Chicago; R. A. WITSCHEY, A. P. Green Fire Brick Co., Chicago.

V. E. ZANG: 1 What size ladle is your teapot ladle? What is the thickness of your lining and what is your holding time?

MR. WYMAN: It is a 5-ton ladle. The thickness of the wall is approximately 5 in. and the holding time varies from 35 min to 1 hr and 10 min.

MEMBER: Do you have a specific length of time for mixing your refractory mix and wnat type of equipment do you use for this operation?

MR. WYMAN: This operation is done in muller-type mixer. The ganister is added first, then the binding materials, and they are mulled for about 12 min. Then there is a mixture of the glutrin, which we temper with a mixture of 1:1 glutrin and water, to 6 per cent moisture. A moisture test is made on every batch.

The life of our ladles has increased four times since we started this investigation.

C. D. SEARS: 2 Do you find moisture content to be a critical

MR. WYMAN: Yes it is. When moisture content is too high, you get that sliding effect from your ramming operation and the material does not stay put.

Mr. SEARS: Could you get a too dry coating the other way? MR. WYMAN: Then it will also affect your ramming operation.

J. H. RICKEY, JR: 8 We have always recommended the use of the form as the author did in the paper. I would like more information about the author's steel forms?

MR. WYMAN: Your initial cost is higher but over a period of time it would be advantageous to use a steel form rather than a wooden form because you must at times resort even to a sledge hammer to get them out.

MR. RICKEY: Out of what do you construct them?

MR. WYMAN: We make them out of plate material in the boiler or plate shop.

MR. RICKEY: In regard to drying of ladles, the author mentioned the use of forced draft in a pre-mix draft chamber. Could it be possible to dry those ladles in a core oven type drier and still get satisfactory results?

MR. WYMAN: Then you would be exposing the gear mechanism to the heat.

MR. RICKEY: Does that affect the lubricant and gear mechan-

MR. WYMAN: I think it will. We use a pre-mix ladle drier which circulates the heat well. It is possible to maintain constant temperature.

steel? What is the fusion point of your mixture containing 98 per cent silica? MR. WYMAN: I do not know what the fusion point of that mixture is. We never made a pyrometric cone (P.C.E.) test on

MEMBER: What is the pouring temperature of your molten

it. The pouring temperature of the molten steel will average 3175 F using an immersion couple.

MEMBER: Fusion point of the refractory is not so important any more. Mr. Wyman has proven that if you form the glass on the surface of the refractory that is worth more than to have a higher fusion point.

CO-CHAIRMAN SCHAUSS: We also have much experience with these glaze formations which set up a high resistance to slag or metal attack or an erosive attack, resulting in longer service life of the refractory.

MR. WYMAN: When we come to the decision to take our ladles out of service, we still have from 4 to 41/2 in. of ladle wall remaining. The bond in the refractory back of the glazed surface will be all burned out. It just crumbles. The bond is absolutely gone, but the thickness of the wall does not vary much from the original wall placed in the ladle. But when you hit that wall with a hammer, it just crumbles. That loss of bond is progressive from the glazed surface toward the ladle shell.

MR. RICKEY: The author mentioned the importance of the grain size of the refractory. We have always recommended the addition of a 1/4 to 1/16 in. ganister with the fines taken out. Do you think that is necessary or is it permissible to include the fines in adding the ganister?

Mr. WYMAN: You need the fines to fill the voids. D. R. HAYES: 4 Are the ladles kept hot at all times?

Mr. Wyman: Yes.

MR. HAYES: How do you control the ramming technique and density of the wall? Do you use a pneumatic hammer for ram-

MR. WYMAN: Yes, we do.

MR. HAYES: How do you avoid air pockets and layer effects? Mr. WYMAN: You put in a certain amount of material and ram it down. Then you scratch or roughen the surface of this material to get a better bond for the next layer. We take extra precaution and after the form is drawn we ram in a horizontal position to remove the soft spots on those layers between the additions of the material.

MR. HAYES: How cool do you allow the ladles to get?

Mr. WYMAN: The shell is absolutely cool.

MR. HAYES: Once you put the ladle into use do you keep your ladies warm at all times for the life of the ladle, or do you allow them to go down to room temperature, then up to \$100

Mr. WYMAN: The ladles are in continuous use, except for about 12 hr during the night when the ladle cools.

C. W. THOMPSON: 5 What method do you use for determining density? Density seems to be an important factor in any refractory. It is especially true in this case. Is there some simplified

method of testing for density?

MR. WYMAN: We use the density recorder from our sand testing equipment. Grain fineness is a better indication of your

F. RUTHERFORD: 6 In ramming the refractory you probably bring some of the fines to the surface, depending upon the moisture content. In that way you probably form harder material at the surface than beneath the surface.

Mr. WYMAN: The theory sounds reasonable.

MEMBER: Would it not be advisable, when initially starting up the ladles, to put some low-fusion material, such as water glass, on the face of the lining to set up the glassy surface? Would that not speed up the process?

MR. WYMAN: I had no success with water glass although many foundrymen use it.

The secret of any ladle lining is the surface vitrified zone

which in our case is over 1 in. thick. W. R. Jaeschke: ⁷ Has a study been made of those types of liquid glass or glass formations in the matrix that are referred to as a bond? In some types of services, is it possible that that glass might have such a low melting point that it might flow out

Vice President, Unitcast Corp., Toledo, Ohio.

² Foundry Metallurgist, International Harvester Co., Indianapolis, Ind.

³ Sales Mgr., Ironton Fire Brick Co., Ironton, Ohio.

⁴ Asst. Supt., Met., Pratt & Whitney Aircraft, E. Hartford, Conn.

⁵ Sales Engr., New Jersey Silica Sand Co., Millville, N. J.

⁶ Sales Engr., Refractories Engineering, Hamilton, Ont., Canada.

⁷ Consulting Engr., Whiting Corp., Harvey, Ill.

in service and carry some of the grog material with it, and in other cases might it be of such a nature that it is not fluid at all but semi-plastic and does act as a bond? It seems to me that in many cases refractory failures are due to the formation of a liquid glass which is carried away by the movement of the metal and causes the failure of refractories. In this case, evidently it seems to act as a bond and protective medium.

Co-CHAIRMAN WITSCHEY: I want to discuss the term P.C.E. which is used by the refractory industry. P.C.E. is not a definition of the melting point of a refractory. It simply explains the softening behavior of refractories. It is a test in which a cone of given material to be tested is compared in heating behavior with a series of standard cones. In other words, a cone is subjected to a temperature cycle and how it compares to a standard cone is given as an index of melting and softening behavior. However, it is a valuable help to refractory producers and users, but it should not be misinterpreted.

With regard to the formation of a glassy phase on refractories, the other service conditions to which a refractory is subjected will largely determine what effect that glassy phase will have in its ultimate behavior. If that refractory is part of a structure which is subjected to excessive load and the loading of the refractory proceeds progressively and the structure has to resist load, then the formation of the glassy phase definitely hastens early failure.

In the case of ladles, the refractory does not have to resist a static load as such. It comes as one of washing and erosion which is resisted by that viscous formation that you get in these

ladles

MR. JAESCHKE: I have seen brick linings in teapot spout ladles. They were used for metals at very high temperature and there was definite flow action down the spout side. That lining would cut or channel very noticeably, showing the movement of the metal was taking something out of the brick that was soft. That may have been a glass face of the refractory also. There must have been something different in this case. What is the differences in the glasses and their characteristics? What makes one beneficial and the other harmful?

CO-CHAIRMAN WITSCHEY: Our discussion of glass is really a discussion of slag reaction. Slag reactions involve the complexities of composition, viscosity and other things which make it difficult to answer specific questions. It is true that the nature of the glasseous phase or slag will vary in composition and these variations in composition will affect the viscosity and the physical properties of that slag to a varying extent. Some slags would be conducive to permitting the aggregate or the coarse particles of the entire mass to float away and others would resist it.

Member: The chemistry of both glasses and slags vary considerably. As far as glasses in brick is concerned, it depends upon what type of brick you are using. In this case, I suppose,

it is silica brick.

MR. JAESCHKE: No, it seems to me that refractory materials that are strictly of the usual fire clay mixture, roughly 50 per cent silica, and 40 to 50 per cent alumina, form a type of glass that washes away. The glasses herein discussed in both the monolithic mixture for this ladle lining and the blow linings in the cupola are a glass formed in a high silica base material. There it seems like a glass formed proves beneficial but in alumina clay mixtures it seems to be harmful. I was referring to a high quality kaolin brick. I have seen that type of brick used in a teapot spout ladle. That lining would channel, cut or wash out very noticeably just from the movement of tilting the ladle, and on the side on which the metal flowed out through the spout showing it was taking something soft out of that brick. This cutting action seems to be in the fireclay mixture and not the high silica base mixture.

MEMBER: As I understand it the glasses formed into monolithics evidently are of such a nature that they resist both temperature and corrosive attack. You feel it is due more to the

higher silica content of the glass.

MR. JAESCHKE: It is related to the composition. I have seen the same condition in high silica base materials, even in firestone. You take firestone brick and cut or saw it and it will make a satisfactory ladle lining. You can use the ladle at high temperature for 8 hr a day and there will be no appreciable amount of slag formed all day. When we use a fire clay mixture the lining cuts and slags, but when we use high silica or high alumina materials we do not have that trouble. Somewhere between the high alumina and high silica, the middle point seems to give us trouble. The glass must be different.

Member: Density or permeability of the refractory must also be considered. In monolithic linings with the same chemistry and more or less the same permeability, the effect of metallic oxide penetration into them becomes practically negligible. In other words, you expose less surface of the refractory to the corresive action of the metallic oxide. I think the silicas, ganisters, and silica sands are prone to be attacked by low fusion oxide of the metallic irons and if you can keep that penetration down to begin with, you can pour many heats before you obtain a concentration of metallic oxide which will prove harmful and flux your ladle. I think the author has definitely accomplished that. That glazed surface prevents further penetration.

N. W. ROUDABUSH: 8 One of the things that was discussed was refractories in ladles. I am more familiar with the steel industry than the foundry industry, but the ladle brick does not have a very high P.C.E. They have inclusions of elements that give off gases that cause the brick to expand. When this happens, you get a higher opening porosity in the brick lining that amounts to practically the same as the rammed lining with the acid refractory. If you look at the silica-alumina-calcium diagram you find that silica can absorb more calcium without losing refractoriness than can the silica-alumina or high alumina brick. So that a cupola lined with silica has a relatively greater resistance to high lime slag than a silica-alumina brick. The same applies in the ladle. I think a highly siliceous ladle, disregarding the physical characteristics of the liquids in the reaction, does have a greater resistance to the lime slags than does the silicaalumina, because the curve in the phase diagram drops off greater when going toward a 100 per cent calcium oxide at the fire clay level than it does with pure silica.

W. Harbrecht:

We have been lining ladles, from 50-lb hand

W. HARBRECHT: ⁹ We have been lining ladles, from 50-lb hand shank size up to 2,500-lb teapot ladles with magnesia or magnasite bond to one with sodium-silicate and water. They are rammed in as the author explained. We get no reaction with metal or slags. We are using basic slags and perhaps one of the reasons for this is that the basic material has a high melting point, higher than the ganister or fireclay bricks. I believe a magnasite lining is excellent for extremely high temperature

application.

MR. JAESCHKE: Drying of refractories is also very important. One foundryman said that no refractory is any good unless it is thoroughly dried before it is put into service. He said that he used different types of rammed linings in converters and the same refractory mix is used in ladles. He doubled the life of the linings of converters by thoroughly drying with a hot air drier before starting preheating. If the converters went into service about half baked, the length of service was cut in half.

CO-CHAIRMAN WITSCHEY: I am happy that Mr. Jaeschke, who is a foundry engineer, brought up that important point. Refractory manufacturers have become lax in stressing that. Ladle life is predicated on adequate drying and burning in. Drying of refractories certainly has as much effect on the performance of ladle refractories as charging does on cupola operation.

8 General Refractories Co., Philadelphia.

0000000

Research Laboratories, Union Carbide & Carbon Corp., Niagara Falis, V. Y.

SUBJECT INDEX

en reill gh ole regh rent

so ry lic In ne son in in ind

as ry aff u ts d n g k. e- e- l- s - eff e

di-ehe gae

Α	nodular, Effect of phosphorus content on mechan-
Absorbing the technical trainee	ical properties of a 501-500
Absorption of gases in metals 201	Oxidation-reduction principles controlling com-
Acid electric steel foundry, Ladle refractories and	position of molten
practice in	Cast steel, Mechanical properties of, as influenced by
Acoustical or ultrasonic tests	mass and segregation
Air furnace melting	Charging magnesium alloys in furnace for melting 140
Air oxidation during induction melting of gray iron 476	Charles Edgar Hoyt lecture
Alloyed white iron, Isothermal transformation charac-	Chemical composition and structure of gray iron, Some
teristics on direct cooling of	effects of temperature and meeting variables on . 472-48
Aluminum bronze, Time required for, to solidify 395 Aluminum melting	Chill:
Aluminum-silicon alloy composition	amount of, Effect of composition of iron on 86
Modification of	depth, Measurement of
A.F.S. Committee reports	depth of, Influence of melting conditions on 88
A.F.S. Fineness Number of Sand	tests and the metallurgy of gray iron
A.F.S. Research reports 67	Chill tests for gray iron
American Society for Engineering Education 530	Chromium, Effect of, on annealing rate of malleable iron 244
Anneal of black-heart malleable iron, Relative effects	Chromium-mclybdenum steels, Effect of vanadium on
of chromium and silicon contents on rate of 244	properties of cast
Annealing furnaces and atmosphere for malleable iron 494-500	Coke Ignitor, Electric 6
Arbitration bars for gray iron testing	Coke properties
Austenite dendrite formation in gray iron 428 Austenite transformation data for alloyed white iron 513	Cold chamber die casting machine
Atmosphere:	Cold shuts in malleable castings
and annealing furnaces for malleable iron 494-500	College graduate, What the, can do for the Ameri-
for annealing malleable iron	can foundryman
gas samples of, over molten gray iron 474	Composition studies of molding sands 291-295
Influence of, on molten gray iron 474	Control, Slag, in basic cupola operation 323-331
	Cooling curves of high purity 85-5-5-5 alloy 366
	Copper alloys, Equipment for melting
Baking, Core, oven	Copper alloys, Refining secondary
Baking, Dielectric core	Copper-base alloy, Effect of 2% nickel on
Barrel-type furnaces for melting aluminum 142	grain size of cast
Basic cupela melting	Copper, Density of pure
Basic cupola operation	Core baking, Dielectric
Slag control in	Core baking oven
Binders, Resin core	Core binders, Resin
Black-heart malleable iron, Relative effects of chrom-	Core blower
ium and silicon contents on rate of anneal of 244	Bench-type
Boiler drop plugs, Manufacture of bronze 296-299	Diaphragm-type 217
Bottom-pour ladle performance	Draw-type
Bronze, aluminum, Time required for, to solidify 395	Core blowing machines, Design of core boxes and
Bronze boiler drop plugs, Manufacture of 296-299	driers for use on
Bronze, Temperature determination of molten 350	Core blows
Bronze, Time required for, to solidify in sand and	Core boxes and driers, Design of, for use on core blow-
insulated cavity	ing machines
Bronze, 88-8-4, Fracture characteristics and melt quality of 67 Bubbling air through molten iron	Core mixes Resin
Buckling in dry sand	Core mixes, Resin
Buckling in green sand	Core practice as related to malleable foundry losses 300-305
bucking in green saint	Core room mechanization
C	Core sand mixing procedure
Calcium carbide, Desulphurization of gray cast iron with 286	Core sand mixtures
Carbide stability in gray iron, Effect of alloying elements on 196	Core sand practice
Carbon control in cupola	Core scabs
Carbon equivalent in gray iron 431	Cores, drying of
Carbon pickup in cupola 23	Cests, Evaluating, in jobbing foundries
Carbon solubility in molten iron	Crucible melting of magnesium
Carburization in cupola operation	Crucibles for melting aluminum
Castability and property comparison of several mag-	Crystallography and metallography of graphite 182 Crystobalite in slag
nesium-rare earth sand casting alloys	Cupola, Basic, melting
Casting finish, Effect of sand grain distribution on 528-528 Cast iron:	Cupola, basic, operation, Slag control in 323-331
Freezing of white, in green sand molds 488-493	Cupola charge
gray, Machinability of high-phosphorus 282	Cupola, charging, Arrangement for 64
Kinetics of graphitization in	Cupola dimensions, Production
and the same of th	Constant Matellianisal blest

Cupola slag analysis	Stationary, pot-type
D	annealing malleable iron
Defect, Scab, on gray iron castings 99 Definitions	heating sand test specimens
Thermal conductivity	Furnaces for melting aluminum: Barrel type
Riser efficiency	Fuel-fired:
Degassing copper-base alloys with nitrogen	Lift-out type
Density of pure copper	Stationary, pot-type
Dephosphorization in cupola operation	Tilting, pot-type 140
Design of core boxes and driers for use on core blow-	Low-frequency induction
ing machines 216	High-frequency induction
Design of furnaces for melting aluminum 149	Reverberatory
Desulphurization in cupola operation	Tilting reverberatory
Desulphurization of gray iron with	C
basic-lined cupola	•
Calcium carbide	Gas samples of atmosphere over molten gray iron 474
Development program, U. S. Navy 348-352	Gases, Injecting, in molten metals
Diaphragm-type core blower	Grain distribution, sand, Effect of, on casting finish
Dielectric baking equipment	Grain size of cast copper-base alloys, Effects of certain
Dielectric baking of cores	elements on
Economics of	Graphite, Crystallography and metallography of 182
Dilatometer tests on sand	Graphite distribution in gray iron 484
Dilatometric thermal analysis of cast steels	Graphitization in cast iron, Kinetics of 181
Dimensional checking and pressure testing of gray	Graphitization of cementite, Subcritical 194
iron castings	Gray iron:
Dimensional checking of unmachined castings, Fixture for 305	Air oxidation during induction melting of 476
Dimensional variations in castings 302	castings. Dimensional checking and pressure testing 304-308
Direct cooling of alloyed white iron, Isothermal trans-	chill tests
formation characteristics on 509-522	desulphurization with calcium carbide 286
Dirt in malleable castings	Effect of:
Distribution, Effect of sand grain, on casting finish 523-528	coke in melting
Dolomite, Calcined, in basic cupola 24	iron oxide additions on composition of 478
Draw-type core blower	melting, under CO 482
Drier, Sand	melting, under slags
Driers and core boxes, Design of, for use on core	refractory type in melting 476
blowing machines	steam treatment of molten 482
Driers and core plates for dielectric baking 164	foundry, medium, Mechanical equipment for 56
Drying of molds and cores	Graphite distribution in
E	Inclusions in
Electric coke ignitor	Machinability of high-phosphorus
End effects on soundness of cast steel bars	melt, Effect of rust on analysis of 479
Equipment and methods of straightening and dimen-	Metallographic examination of 404-410
sional inspection of malleable iron castings 418-424	microstructure
Equipment for melting	Solidification of, in sand molds 425-434
aluminum alloys	Some effects of temperature and melting variables
copper alloys	on chemical composition and structure of 472-487
Equipment for nondestructive testing 309-314	Greco-Latin square
Equipment, Mechanical, for medium-size gray iron foundry 56	Green sand molds, Freezing of white cast iron in 488-493
Eutectoid reaction in gray iron	Gypsum riser sleeve, Use of
Exchange papers	
A.F.S. to IBF	н
IBF to A.F.S	Hardness of magnesium-rare earth alloys 539
Australian Institute of Foundrymen to A.F.S 28	Heat and mass flow analyzer, Thermal studies with 13
Swedish Institute for Metal Research to A.F.S 35	Heat flow in moist sand
A.F.S. to French Technical Foundry Association 56	Heat problems in the foundry 9
Exothermic riser compounds	Heat transfer, Fundamentals of
F	Heat transfer theory, Practical consequences of 8
	Heat transmission to sand mold
Fatigue data summary	Heat treatment of magnesium-rare earth alloys 533
Fayalite in slag	High and trade schools, Development of foundry
Feeding of nonferrous castings	courses in
Fineness number of sand	High-phosphorus gray cast iron, Machinability of 282
First-stage annealing of malleable iron	Hiring of personnel
Fixture for dimensional checking of unmachined casting 305	Hot chamber die casting machine
Foundry courses in high and trade schools 332-336	Hot tears in castings
Foundry losses, Core practice as related to malleable . 300-303	Hot tests on sand
Foundry training course outline	Hot toughness of green sand
Foundryman, What the college graduate can do for the 529-531	Hydrogen, Influence of, on solid steel 203
Fracture characteristics and melt quality of 85-5-5	
and 88-8-4 alloys 67	l .
Freezing of white cast iron in green sand molds 488-493	Ignitor, Electric coke
Fuel-fired furnaces for melting aluminum:	Immersion test for steel penetration into molding sand . 37
Lift-out type	Impact, Testing of sand under

Inclusions in gray iron	
inclusions in gray iron	Metal penetration
to describe analytical many from Ata antidation destruction Africa	Metal penetration
Induction melting gray iron, Air oxidation during 476	
Induction melting of cast iron	Mechanism of
Industrial research, Management of	Metallographic examination of gray iron 404-410
Industry participation in secondary school training	Metallography and crystallography of graphite 182
program for the foundry	Metallurgical blast cupola
Ingot remelting, Aluminum	Methods of straightening and dimensional inspection
Inoculation of gray iron	of malleable iron castings
Inspection of malleable iron castings, Equipment and	Michigan dune sand 412
	M.cro-examination of cast steels
methods of straightening and dimensional 418-424	Microstructure of:
Insulating riser sleeves	
Investment casting, Precoat materials for 315-322	alloyed white iron
Iron, cast, Kinetics of graphitization in , 181	gray iron
Iron castings, Silicon-chromium alloy in complicated . 401-411	magnesium-rare earth alloys
Iron oxide additions. Effect of, on composition of gray iron 478	Mo-Ni-Cr cast iron
Isothermal transformation characteristics on direct	nodular cast iron
cooling of alloyed white iron 509-522	Misruns and cold shuts in malleable castings 303
	Modification of aluminum-silicon alloys 28
1	Molasses-bonded sands 454
	Molding equipment for gray iron foundry 59
Jobbing foundries, Evaluating costs in	Molding materials, Nonferrous
K	Molding sands, Compaction studies of
Keel block	Molds, Drying of
Keel block	Mo-Ni-Cr cast iron, Structure and mechanical
Kinetics of graphitization in cast iron	properties of
1	
the state of the s	N
Ladle refractories and practice in acid electric steel	Nickel, Effect of 2 per cent, on 5 Pb-Cu alloy 363
foundry	Nitrogen, Degassing copper-base alloys with 74
Losses, Core practice as related to malleable foundry . 300-303	Nodular cast iron
	Effect of phosphorus content on mechanical prop-
Low-frequency induction melting furnace 143	
M	erties of
IVI	Photomicrographs of
Machinability of high-phesphorus gray cast iron 282*	Nodule counts in malleable iron 247
Magnesite brick in basic cupola 24	Nondestructive testing, Equipment for 309-314
Magnesium alloys:	Nonferrous development program, U. S. Navy 348-352
Die casting	
Melting of	0
	Oil-bonded sands
Magnesium melting:	Oil, Core, evaluation method
and refining	
in crucibles	Open-pot method of magnesium melting 147
in reverberatory furnace	Outline for foundry training courses
Open-pot method of	Oxidation reactions in melting iron
Magnesium-rare earth sand casting alloys, A castability	Oxidation-reduction principles controlling composi-
and property comparison of several	tion of molten cast irons
Magnetic particle tests	
Magnetite in slag	P
Malleable cast iron—annealing furnaces and atmosphere 494-500	Donatrant tests
	Penetrant tests
	Devetors in and much muface.
Malleable foundry losses, Core practice as related to . 300-303	Penetration and rough surfaces
Malleable iron castings, Equipment and methods of	Penetration in sand molds, Determination of metal 253
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424	Penetration in sand molds, Determination of metal 253 Penetration of steel:
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and sili-	Penetration in sand molds, Determination of metal 253
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424	Penetration in sand molds, Determination of metal 253 Penetration of steel:
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and sili-	Penetration in sand molds, Determination of metal
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244	Penetration in sand molds, Determination of metal
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart	Penetration in sand molds, Determination of metal 253 Penetration of steel: Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53 Penetration of steel, Effect of:
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: 50 binders and admixtures in sand 45
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal 253 Penetration of steel: Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: binders and admixtures in sand 45 chemical reactions at steel:sand boundary 48
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53, 54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: 50 binders and admixtures in sand 45 chemical reactions at steel:sand boundary 48 composition of the molten metal 44 grain size and grain distribution of the sand 44 moisture content of the sand 46 mold washes on 46 permeability of the sand 46 preheating of the sand 47 ramming density 46
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244 Management of industrial research 1 Mass, Mechanical properties of cast steel as influenced by 223 Mcchanical properties and structure of a Mo-Ni-Cr cast iron 353-359 Mechanical properties of a nodular cast iron, Effect of phosphorus content on 501-508 Mechanism of gamma-range graphitization of gray iron 184, 185 Mechanism of metal penetration 114 Mechanism of steel penetration in sand 53, 54 Mechanization of gray iron foundry 56 Melt quality and fracture characteristics of 85-5-5-5 and 88-8-4 alloys 67 Melting: aluminum and magnesium-base alloys 139 and refining magnesium 398 copper alloys, Equipment for 465 gray iron, Mechanization in 663 gray iron under CO. Effect of 482	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244 Management of industrial research	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53, 54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: binders and admixtures in sand 45 chemical reactions at steel:sand boundary 48 composition of the molten metal 44 grain size and grain distribution of the sand 46 moisture content of the sand 46 mold washes on 46 permeability of the sand 47 ramming density 46 sintering point 36 static pressure of molten metal 36, 43 temperature of molten metal 45 thermal diffusivity of the sand 46 time factor 42 Personnel, Hiring of 376 Petrographic examination of slag 128, 129
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244 Management of industrial research 1 Mass, Mechanical properties of cast steel as influenced by 223 Mcchanical properties and structure of a Mo-Ni-Cr cast iron 353-359 Mechanical properties of a nodular cast iron, Effect of phosphorus content on 501-508 Mechanism of gamma-range graphitization of gray iron 184, 185 Mechanism of steel penetration 114 Mechanism of steel penetration in sand 53, 54 Mechanization of gray iron foundry 56 Melt quality and fracture characteristics of 85-5-5 and 88-8-4 alloys 67 Melting: aluminum and magnesium-base alloys 139 and refining magnesium 398 copper alloys, Equipment for 465 gray iron under CO. Effect of 482 gray iron under slag, Effect of 481 magnesium alloys 146 malleable iron 244 of 85-5-5-5 brass 361	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: 50 binders and admixtures in sand 45 chemical reactions at steel:sand boundary 48 composition of the molten metal 44 grain size and grain distribution of the sand 46 moisture content of the sand 46 moid washes on 46 permeability of the sand 46 preheating of the sand 47 ramming density 46 sintering point 36 static pressure of molten metal 45 thermal diffusivity of the sand 46 time factor 42 Personnel, Hiring of 376 Petrographic examination of slag 128, 129 Phosphorus content. Effect of, on mechanical proper-
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244 Management of industrial research 1 Mass, Mechanical properties of cast steel as influenced by 223 Mcchanical properties and structure of a Mo-Ni-Cr cast iron 353-359 Mechanical properties of a nodular cast iron, Effect of phosphorus content on 501-508 Mechanism of gamma-range graphitization of gray iron Mechanism of metal penetration 114 Mechanism of steel penetration in sand 53, 54 Mechanization of gray iron foundry 56 Melt quality and fracture characteristics of 85-5-5-5 and 88-8-4 alloys 67 Melting: aluminum and magnesium-base alloys	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: 50 binders and admixtures in sand 45 chemical reactions at steel:sand boundary 48 composition of the molten metal 44 grain size and grain distribution of the sand 46 moisture content of the sand 46 moid washes on 46 permeability of the sand 46 preheating of the sand 47 ramming density 46 sintering point 36 static pressure of molten metal 45 thermal diffusivity of the sand 46 thermal diffusivity of the sand
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244 Management of industrial research	Penetration in sand molds, Determination of metal Penetration of steel: Factors which counteract
Malleable iron castings, Equipment and methods of straightening and dimensional inspection of 418-424 Malleable iron, Relative effects of chromium and silicon contents on rate of anneal of black-heart 244 Management of industrial research 1 Mass, Mechanical properties of cast steel as influenced by 223 Mcchanical properties and structure of a Mo-Ni-Cr cast iron 353-359 Mechanical properties of a nodular cast iron, Effect of phosphorus content on 501-508 Mechanism of gamma-range graphitization of gray iron Mechanism of metal penetration 114 Mechanism of steel penetration in sand 53, 54 Mechanization of gray iron foundry 56 Melt quality and fracture characteristics of 85-5-5-5 and 88-8-4 alloys 67 Melting: aluminum and magnesium-base alloys	Penetration in sand molds, Determination of metal 253 Penetration of steel: 50 Factors which counteract 50 Factors which favor 49 Immersion tests for 37 Mcchanism of 53,54 Rate of 53 Resistance to 53 Penetration of steel, Effect of: 50 binders and admixtures in sand 45 chemical reactions at steel:sand boundary 48 composition of the molten metal 44 grain size and grain distribution of the sand 46 moisture content of the sand 46 moid washes on 46 permeability of the sand 46 preheating of the sand 47 ramming density 46 sintering point 36 static pressure of molten metal 45 temperature of molten metal 45 temperature of molten metal 46 thermal diffusivity of the sand 46 time factor 42 Personnel, Hiring of 376 <t< td=""></t<>

magnesium-rare earth anoys	Scaling and growth tests for nodular from
metal penetration :	School, secondary, training programs for the foundry,
nodular cast iron	Industry participation in
Plugs, Manufacture of bronze boiler drop 296-299	Scum formation on surface of molten gray iron 47
Polymerization of resin bond	Sea coal, Effect of, in solidification of gray iron 48
Pots, Melting, for aluminum	Secondary copper alloys, Refining
Precoat materials for investment casting 315-322	Second-stage annealing of malleable iron
Preheating of sand, Effect of, on penetration of steel 52	Segregation, Mechanical properties of cast steel as
Pressure testing of gray iron castings 304-308	influenced by
Projects, Nonferrous development	Shrinks in malleable castings
	Sieve analyses of silica sands
R	Sieve shakers for sands
Radiographic examination of castings	Silica reduction in gray iron melting
Radiographic studies on cast steel	Silica sands, Sieve analyses of
Radiography	Silicate inclusions in gray iron
as an assistant to foundry practice	Silicon-chromium alloy in complicated iron castings 401-41
Rare earth alloys, Analysis of 534	Silicon content, Influence of, on mechanical and high
Rare earth-magnesium sand casting alloys	temperature properties of nodular cast iron 387-347
Red brass, 85-5-5-5, Fracture characteristics and melt	Silicon, Effect of, on annealing rate of malleable iron 24
quality of	Silicon less in cupola
Refining secondary copper alloys	Silicon pickup in melting iron
	Sintering point, Effect of, on steel penetration into sand 36
Refractories:	Slag analysis, Cupola
Drying and preheating 549	Slag and metal reaction in basic cupola melting 323
Installation of	Slag control in basic cupola operation 323-331
Ladle, and practice in acid electric steel foundry . 546-554	Slag, Dry, formation on surface of molten gray iron . 473, 473
Refractory:	Slag over molten gray iron, Influence of type of 476
consumption in basic cupola	Slag phase in gray iron melting
cost and maintenance in basic cupola 329	Sodium chloride-fluoride, Modification of aluminum-
technique in basic cupola	silicon alloys by
type, Effect of, in melting gray iron 476	Sodium metal, Modification of aluminum-silicon alloys with 3
Remelting aluminum ingot, Effect of temperature and	Solidification of:
holding time in	castings
Research activities:	gray iron in sand molds 425-434
classified	steel against sand and chill walls
Financing of	steel, analogue method of study 446-447
Research defined	Solubility of carbon in molten iron 20
Research director, Functions of	Specification sheets and their various uses 372-375
Research group, Operation of 4	Static pressures, Effect of, on steel penetration into
Research, Management of industrial	molding sand
Research, Origin of	Steam treatment of melten gray iron, Effect of 482
Resin-bonded sands 453	Steel bars, cast, Riser and casting end effects on sound-
Resin core:	ness of
binders	Steel, cast, Mechanical properties of, as influenced by
mixes	mass and segregation
Reverberatory furnaces and melting:	Steel, Effect of vanadium on properties of cast Cr-Mo 261
aluminum	Steel, Penetration of, into molding sand
magnesium	Steel, Solidification of, against sand and chill walls 435-450
Riser and casting end effects on soundness of cast steel bars 171	Strain cracks
Riser efficiency:	Stopper and nozzle assembly for bottom-pour ladle 551
defined	Structure and mechanical properties of a Mo-Ni-Cr
Increasing	cast iron
Risering fundamentals	Structure of gray irons, Some effects of temperature
Rust, Effect of, on analysis of gray iron melt 479	and melting variables on chemical composition and 472-487
	Sulphur dioxide in melting magnesium
S	Sulphide inclusions in gray iron
	Sulphur removal in cupola from iron 21
Sampling of ladles	
Sand:	T
Core, practice	The state of the s
dryer	Technical trainee, Absorbing the
fineness number	Temperature and melting variables, Some effects of,
grain distribution, Effect of, on casting finish 523-528	on chemical composition and structure of gray iron 472-487
handling layout 61-63	Temperature control in melting:
Molding, preparation 60-63	aluminum alloys
Testing of, under impact	copper alloys
Sands:	magnesium alloys
Compaction studies of molding 291-295	Temperature of molten bronze
Molasses-bonded	Testing sand under impact
Oil-bonded	Thermal analysis of cast steels
Resin-bonded	Thermal conductivity defined
Sieve analyses of silica	Thermal considerations in foundry work
Sampling. Method of, molten copper alloys for analysis . 466	
Sawdust in cores, Objections to	A MODELLINE STATE A CONTROL OF THE C
Scab defect on gray iron castings	Thermal studies
0 , 0	Methods of making
Scabs on gray iron castings:	with heat and mass flow analyzer
Cause of	Tilting ladle, U-type, for gray iron
Cure for	Tilting reverberatory furnace for melting aluminum 142
Foundry conditions that promote	Timestudy fatigue factor

courses in								
Trainee, Absorbing the technic 1							462	-464
Training course outline for foundry	V							333
Training programs								
Training programs for the foundry,	In	du	stry	pa	rtic	ci-		
pation in secondary school .							376	-380
Transformation characteristics, Isoth cooling of alloyed white iron							509	-522
T-T-T diagrams for alloyed white i	ron	1						513
U								
Ultrasonic or acoustical tests								313
U. S. Navy nonferrous development	pr	ogr	am				348	-352
U-type tilting ladle								

W	
Water cooling of cupola	25
Wedge tests for gray iron	81
Weigh lorry for weighing cupola charge	64
alloyed, Composition of	510
alloyed, Isothermal transformation characteristics	
on direct cooling of 509	-522
Freezing of, in green sand molds 488	-493
melting	509
photomicrographs 510, 514-516, 519	
85-5-5-5	
alloy, Cooling curves of high purity	366
red brass fracture characteristics and melt quality	67
impurities in	361
88-8-4, bronze, Fracture characteristics and melt quality of 80-10-0-2 bronze, Continuous cast, macroetch showing	67
coarse columnar grains	360

AUTHOR INDEX

4 T.C. C. T.							
A.F.S. COMMITTEE REPORTS—							
Chill Tests and the Metallurgy of Gray Iron							
Fatigue Data Summary—Report No. 2	*						
Scab Defect on Gray Iron Castings				•			
A.F.S. RESEARCH REPORTS—							
Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze .							
Annich, M. E.—Fatigue Data Summary, Report No. 2							
AUSTIN, W. W., JRImprovement of Machinability in High-Phosphorus Gray Cast Iron-P.	art 1	1 .					. 2
BISHOP, H. F., MYSKOWSKI, E. T. and PELLINI, W. SThe Contribution of Riser and Casting			oundn	ess of	Cast	Steel	
Bars							1
BISHOP, H. F., BRANDT, F. A. and PELLINI, W. SSolidification of Steel Against Sand and C	hill	Walls					1
BLAKE, E. ADesign of Core Boxes and Driers for Use on Core Blowing Machines .							2
BOCK, W. K. and Schwartz, H. AFreezing of White Cast Iron in Green Sand Molds							48
BRANDT, F. A, BISHOP, H. F. and PELLINI, W. SSolidification of Steel Against Sand and	Chill	Walls					43
Brosky, S. A. and Johnson, C. BRadiography as an Assistant to Foundry Practice .							13
Brown, B. F. and HAWKES, M. FKinetics of Graphitization in Cast Iron						,	13
CABLE, J. WRecent Advances in Dielectric Core Baking							1!
CARTER, C. LAbsorbing the Technical Trainee							46
CLARK, R. ASilicon-Chromium Alloy in Complicated Iron Castings							10
COLTON, R. A. and MARGOLIS, MEffects of Certain Elements on Grain Size of Cast Copp		se Allo	bys				30
CONGER, G. A., ROTE, F. B. and DELONGE, K. AIsothermal Transformation Characteristics	on	Direct	Cooli	ng o	f Alle	oyed	
White Iron							50
CORNELL, R. CDie Casting Magnesium Alloys			."				39
CULLEN, O E. and LIGHT, R. JMalleable Cast Iron-Annealing Furnaces and Atmosphere							10
DAVENPORT, W. F. and STROTT, APrecoat Materials for Investment Casting							31
DAVIDSON, J. R. and KLINE, B. FManufacture of Bronze Boiler Drop Plugs							29
DELONGE, K. A., ROTE, F. B. and CONGER, G. AIsothermal Transformation Characteristics					Alle	oved	
White Iron							50
DUNPHY, R. P. and PELLINI, W. SSolidification of Gray Iron in Sand Molds							42
Dyke, R. HModification of Aluminum-Silicon Alloys							0
EASTWOOD, L. WMelting Aluminum and Magnesium-Base Alloys							13
EKEY, D. C. and GONYA, H. JDetermination of Metal Penetration in Sand Molds							25
ELSEA, A. R., WHITE, W. H. and RICE, L. PInfluence of Silicon Content on Mechanical and					perties	of	
Nodular Cast Iron	0						33
EXCHANGE PAPERS FROM A.F.S.							
Some Thermal Considerations in Foundry Work							
Mechanical Equipment for Medium-Size Gray Iron Foundry							50
EXCHANGE PAPERS TO A.F.S.							
Basic Cupola Melting and Its Possibilities							20
Modification of Aluminum-Silicon Alloys							28
An Investigation of the Penetration of Steel Into Molding Sand							3
FAIRFIELD, H. H., MURTON, A. E and RICHARDSON, BCore Oil Evaluation Method							270
FAIRFIELD, H. H. and MACCONACHIE, JEffect of Sand Grain Distribution on Casting Finish							523
FLINN, R. A. and KRAFT, R. WImportance of Slag Control in Basic Cupola Operation .							325
FREAR, C. LU. S. Navy Non-Ferrous Development Program							348
FREUND, C. JWhat the College Graduate Can Do For the American Foundryman							529
GERTSMAN, S. L. and A. E. MURTON-Metal Penetration							108
GLASSENBERG, M., MONDOLFO, L. F. and HESSE, A. HRefining Secondary Copper Alloys .							465
GOLDSMITH, J. R., ZIEGLER, N. A. and MEINHART, W. LEffect of Vanadium on the Propert							100
num Steels					, , , ,		261
GONYA, H. J. and EKEY, D. CDetermination of Metal Penetration in Sand Molds							253
GREENLEE, R. HModern Core Sand Practice							412
GRIM, R. E. and JOHNS, W. D. JRCompaction Studies of Molding Seads							291
HASTINGS, C. HChoosing Equipment for Nondestructive Testing							309
HAWKES, M. F. and Brown, B. FKinetics of Graphitization in Case Iron							181
HEINE, R. WOxidation-Reduction Principles Controlling the Composition of Molten Cast	Irons						121
HEINE, R. W. and LANGE, E. ASome Effects of Temperature and Melting Variables on Cher		Compo	osition	and S	tructu	ire	
of Gray Irons							472
Hesse, A. H., Glassenberg, M. and Mondolfo, L. FRefining Secondary Copper Alloys							465
OHNS, W. D. JR. and GRIM, R. ECompaction Studies of Molding Sands							291
OHNSON, C. B. and Brosky, S. A.—Radiography as an Assistant to Foundry Practice							151
ORY, E. L.—Core Practice as Related to Malleable Foundry Losses							300
KRAFT, R. W. and FLINN, R. A.—Importance of Slag Control in Basic Cupcla Operation KRAUSE, D. E.—Chill Tests and the Metallurgy of Gray Iron						*	323 79
						*	296
LINE, B. F. and DAVIDSON, J. R.—Manufacture of Bronze Boiler Drop Plugs							117
	10-1	Comm	olei			*	117
ANGE, E. A. and HEINE, R. WSome Effects of Temperature and Melting Variables on Chen	ncal	compc	sation a	and S	THEIN	I'C	470
of Gray Irons	*	*			*	*	172
IGHT, R. J. and CULLEN, O. E.—Maileable Cast Iron—Annealing Furnaces and Atmosphere							494
ORIA, E. A.—Structure and Mechanical Properties of a Mo-Ni-Cr Cast Iron					*		353
IACCONACHIE, J. and FAIRFIELD, H. HEffect of Sand Grain Distribution on Casting Finish	D.	411		*			523
JARGOLIS, M. and COLTON, R. A.—Effects of Certain Elements on Grain S'ze of Cast Copper-							360
IcQuiston, C. E. and Schneble, A. W. Jr.—A New Method of Evaluating Costs in Jobbing F	ound	1163					210

9_ 8.

7 56

MIGERON, A. E., PARCEULD, H. H. and RICHARDSON, B.—Core Oil Evaluation Method MIGERON, A. E., FARCUILD, H. H. and RICHARDSON, B.—Core Oil Evaluation Method MISSROWS, E. T. BISSTON, H. F. and PELLINI, W. S.—The Contribution of Riser and Casting End Effects to Soundness of Cast Sicel Bars NELSON, K. E. and STRIETER, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys OKEEFE, J., JR.—Increasing Riser Efficiency PASCHERS, V.—Some Thermal Considerations in Foundry Work PASCHERS, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Sicel Bars PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Sicel Bars PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Sized Against Sand and Chill Walls PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Sized Inso Molding Sand PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Sized Inso Molding Sand STRIETERSON, H.—An investigation of the Penetration of Sized inso Molding Sand STRIETERSON, H.—An investigation of the Penetration of Sized inso Molding Sand STRIETERSON, E.—The Relative Effects of Chromium and Silicra Contents on Rate of Anneal of Black-Heart Malleable Iron STRIETERSON, E.—Basic Cupola Melting and its Possibilities GN Nodular Cast Iron STRIETERS, F. S.—Basic Cupola Melting and its Possibilities GN Nodular Cast Iron STRIETER, F. D. and UPTHERONY, E. O. and UPTHERONY, E. C.—Melt Quality and Fracture Characteristics on Direct Cooling of Alloyed White Iron STRIELLEN, R. D. and UPTHERONY, E.—Melton of Sized Insolution Method GNOTE, F. B., SIRILLENG, R. D. and UPTHERONY, E.—A New Method of Evaluation Method STRIELLEN, R. W. J., B.—Industry Participation in Secondary School Training Programs for the Foundry SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segr	
MONDERO, L. F., GLANSTNERG, M. and HESSE, A. H.—Refining Secondary Copper Alloys MOGRE W. H.—Testing of Sand Under Impact MICHON, A. E. and GERTAIAN, S. L.—Metal Percentation MICHON, A. E. and STRIETER, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys OKEEPE, J., Jr.—Increasing Riser Efficiency PASCHEN, V.—Some Thermal Considerations in Foundry Work PASCHEN, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars PELLINI, W. S. and DUNPHY, R. P.—Solidification of Gray Iron in Sand Molds PELLINI, W. S., BISHOP, H. F. and BANDY, F. A.—Solidification of Steel and Molds PELLINI, W. S., BISHOP, H. F. and BRANDY, F. A.—Solidification of Steel Into Molding Sand RAING, F. W. and KENNINSKY, A. O.—Sieve Analyses of Stilica Sands PETTASSON, H.—An investigation of the Penetration of Steel into Molding Sand RAING, F. W. and KENNINSKY, A. O.—Sieve Analyses of Silica Sands RENNING, F. W. and KENNINSKY, A. O.—Sieve Analyses of Silica Sands RENNING, F. W. and KENNINSKY, A. O.—Sieve Analyses of Silica Sands RICKSHAW, E. S.—Basic Cuppola Melting and its Possibilities REINDER, J. E.—Effect of Phorpheriru Content on Mechanical Properties of a Nodular Cast Iron Solica Renormal Composition of Silica Sands RICKSHAW, E. S.—Basic Cuppola Melting and its Possibilities R	
MIGERON, A. E., PARCEULD, H. H. and RICHARDSON, B.—Core Oil Evaluation Method MIGERON, A. E., FARCUILD, H. H. and RICHARDSON, B.—Core Oil Evaluation Method MISSROWS, E. T. BISSTON, H. F. and PELLINI, W. S.—The Contribution of Riser and Casting End Effects to Soundness of Cast Sicel Bars NELSON, K. E. and STRIETER, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys OKEEFE, J., JR.—Increasing Riser Efficiency PASCHERS, V.—Some Thermal Considerations in Foundry Work PASCHERS, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Sicel Bars PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Sicel Bars PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Sized Against Sand and Chill Walls PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Sized Inso Molding Sand PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Sized Inso Molding Sand STRIETERSON, H.—An investigation of the Penetration of Sized inso Molding Sand STRIETERSON, H.—An investigation of the Penetration of Sized inso Molding Sand STRIETERSON, E.—The Relative Effects of Chromium and Silicra Contents on Rate of Anneal of Black-Heart Malleable Iron STRIETERSON, E.—Basic Cupola Melting and its Possibilities GN Nodular Cast Iron STRIETERS, F. S.—Basic Cupola Melting and its Possibilities GN Nodular Cast Iron STRIETER, F. D. and UPTHERONY, E. O. and UPTHERONY, E. C.—Melt Quality and Fracture Characteristics on Direct Cooling of Alloyed White Iron STRIELLEN, R. D. and UPTHERONY, E.—Melton of Sized Insolution Method GNOTE, F. B., SIRILLENG, R. D. and UPTHERONY, E.—A New Method of Evaluation Method STRIELLEN, R. W. J., B.—Industry Participation in Secondary School Training Programs for the Foundry SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segr	
MICHON, A. E. and Gert and, S. L.—Metal Fenetration Micholon, A. E., Farkerilla, H. H. and Reicharbason, B. Coro Oil Evaluation Method Myskowski, E. T., Bishop, H. F. and Pellin, W. S.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Nelson, K. E. and Stheffer, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys O'Keefee, J., Jr.—Increasing Riser Efficiency Paschis, V.—Some Thermal Considerations in Foundry Work Paschis, V.—Some Thermal Considerations in Foundry Work Paschis, V.—Some Thermal Considerations in Foundry Work Paschis, V.—Heat Flow in Moist Sand Pellin, W. S., Bishop, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Pellin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Pellin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls Petilin, W. B. Bartis, W. S.—Service Analyses of Silica Saudo Residual Cast Iron R	MOORE, W. HTesting of Sand Under Impact
Myskowski, E. T., Bishop, H. F. and Pellini, W. S.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Nelson, K. E. and Striefer, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys O'Keefe, J., Jr.—Increasing Riser Efficiency Paschers, V.—Heat Flow in Moist Sand Pellini, W. S., Bishop, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Nelson, W. S., Bishop, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Pellini, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel and Casting End Effects to Soundness of Cast Steel Bars Pellini, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel and Casting End Effects of Cast Steel Bars Raming, F. W. and Krintisky, A. O.—Sieve Analyses of Silica Sauds Raming, F. W. and Krintisky, A. O.—Sieve Analyses of Silica Sauds Render, F. B.—The Relative Effects of Chromium and Silicon Contents on Rate of Anneal of Black-Heart Malleable Iron Eribrize, J. E.—Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron Eribrize, J. E.—Effect of Phosphorus Content on Mechanical Properties of Nodular Cast Iron Eribrize, J. E.—Effect of Phosphorus Content on Mechanical Method Richershaw, L. W. Brieffer, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron Richersham, L. D. and Uptherskove, C.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-84 Bronze Rote, F. B., Concer, G. A. and Delonge, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron Schware, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schware, J. H., Wallace, J. F. and Fartor, R. B.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze Syripe, E.—Elf Effects and Usefulness of Gases in M	MURTON, A. E. and GERT MAN, S. LMetal Penetration
NELSON, K. E. and STRIETER, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys O'KEEFE, J., JR.—Increasing Riser Efficiency PASCHRIS, V.—Some Thermal Considerations in Foundry Work PASCHRIS, V.—Some Thermal Considerations in Foundry Work PASCHRIS, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Ba:s PELLINI, W. S. and DUNPHY, R. P.—Solidification of Gray Iron in Sand Molds PELLINI, W. S. BISHOP, H. F. and BEANDT, F. A.—Solidification of Steel Against Sand and Chill Walls PELLINI, W. S., BISHOP, H. F. and BEANDT, F. A.—Solidification of Steel Against Sand and Chill Walls PETTERSSON, H.—An investigation of the Penetration of Steel Into Molding Sand RABING, F. W. and KRYNITSKY, A. O.—Sieve Analyses of Silica Sands RABING, F. W. and KRYNITSKY, A. O.—Sieve Analyses of Silica Sands REBIDEZ, J. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron STRIED, J. E.—Effect of Phophepherus Content on Mechanical Properties of Nodular Cast Iron RESIBIAN, E. E.—Basic Cupola Melting and its Possibilities RICHARDSON, E.—Here The Phophepherus Content on Mechanical and High-Temperature Properties of Nodular Cast Iron STRICHARDSON, B., MURITON, A. A. and FAIRFFELD, H. H.—Core Oil Evaluation Method ROTE, F. B., SIRELLENG, R. D. and UPTHERGOVE, C.—Melt Quality and Fracture Characteristics on Direct Cooling of Alloyed White Iron RUSSILL, T. W., Ja.—Industry Participation in Secondary School Training Programs for the Foundry SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional I	
Netson, K. E. and Strikter, F. P.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys Nespers, J., Jr.—Increasing Riser Efficiency Paschiris, V.—Some Thermal Considerations in Foundry Work Paschiris, V.—Heat Flow in Moist Sand Pellin, W. S., Bishor, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Pellin, W. S., Bishor, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Pellin, W. S. and Dunyhy, R. P.—Solidification of Gray Iron in Sand Molds Pellin, W. S. and Dunyhy, R. P.—Solidification of Gray Iron in Sand Molds Pellin, W. S. and Dunyhy, R. P.—Solidification of Steel Into Molding Sand Steel Bars Raning, F. W. and Kryntrin, A. O.—Sieve Analyses of Silica Sands Raning, F. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 24. Reider, J. E.—Effect of Phospherus Content on Mechanical Properties of a Nodular Cast Iron Ersinder, J. E.—Effect of Phospherus Content on Mechanical Properties of Anodular Cast Iron Strike, S.—Basic Cupola Melting and its Possibilities Rice, L. P., White, W. H. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron Strike, S.—Basic Cupola Melting and its Possibilities Richer, E. B.—Subschild, A. A. and Farkfeed, H. H.—Core Oil Evaluation Method Richer, E. D. and Uptherrow, E.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze Royae, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schware, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schware, J. H., A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Sierly, A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Sierly, A. and Dock, W. K.—Freezing of White Cast	
Alloys O'KEEFE, J., JR.—Increasing Riser Efficiency PASCHERS, V.—Some Thermal Considerations in Foundry Work PASCHERS, V.—Some Thermal Consideration of Steel Bass and Casting End Effects to Soundness of Cast Steel Ba:s PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Into Molding Sand RAINIGE, F. W. and KANYINSKY, A. O.—Sieve Analyses of Silica Sauds RAINIGE, F. W. and KENYINSKY, A. O.—Sieve Analyses of Silica Sauds REBIDER, J. E.—The Relative Effects of Chromium and Silicin Contents on Rate of Anneal of Black-Heart Malleable Iron RESIGHARY, E. S.—Basic Cupola Melting and its Possibilities REIGHER, J. P., WHITE, W. H. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron RICHARDSON, B., MURTON, A. A. and Fairffeld, H. H.—Core Oil Evaluation Method ROTE, F. B., SHELLENG, R. D. and Uptiezrove, C.—Melt Quality and Fracture Characteristics on Direct Cooling of Alloyed White Iron RUSSILL, T. W., Jr.—Industry Participation in Secondary School Training Programs for the Foundry SAVAGE, J. H., WALLAGE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation SCHINBLE, A. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds SIELLENG, R. D., Uptiezrove, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 BRODZE SAITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings STRUETE, F. P. and NELSON, K. E.—A Castability and Properties of Cast Steel as Influenced by Mass and Segregation SAITH, K. M.—Di	
O'REFER, J., JR.—Increasing Riser Efficiency PASCHRIS, V.—Some Thermal Considerations in Foundry Work PASCHRIS, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bai's PELLINI, W. S., BISHOP, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls PETTERSSON, H.—An investigation of the Penetration of Steel into Molding Sand RABING, F. W. and KRYNITSKY, A. O.—Sieve Analyses of Silica Sauds RABING, F. W. and KRYNITSKY, A. O.—Sieve Analyses of Silica Sauds RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 24. RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 25. RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 26. RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 27. RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 28. RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 28. RESIBBAP, J. E.—The Reclaire Effects of Chromium and Silican Contents on Rechanical and High-Temperature Properties of Nodular Cast Iron 29. RESIBBAP, J. E.—The Reclaire Effects of Contents on Mechanical Malleable Iron 29. RICHARDSON, B., MURTON, A. A. and FAREFELD, H. H.—Core Oil Evaluating Cress in Jobios of Bass and 88-8-4 Bronze 20. RUSSELLE, T. W., Jr.—Industry Participation in Secondary School Training Pregrams for the Foundry 20. RUSSELLE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 20. RUSSEL	
PASCHISS, V.—Some Thermal Considerations in Foundry Work PASCHISS, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Ba:s PELLINI, W. S., BISHOP, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Ba:s PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 42. PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 43. PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 44. PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 45. PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 46. PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 47. PELLINI, W. S., BISHOP, H. F. And BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 48. PELLINI, W. S., BISHOP, H. F. And BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 49. PELLINI, W. S., BISHOP, H. F. And BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 49. PELLINI, W. S., BISHOP, H. F. And BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 40. PELLINI, W. H. H. B. A. O.—Selve Analyses of Silicion Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 40. PELLINI, W. H. H. H. H. A. O. O.—Sieve Analyses of Silicion Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 40. PELLINI, W. H. H. H. And LESA, A. R.—Indication of Steel Against Sand and Casting Steel Against Sand and Science Against Sand Sand Science Aga	O'KEEFE, L. IRIncreasing Riser Efficiency
PASCHERS, V.—Heat Flow in Moist Sand PELLINI, W. S., BISHOP, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars Steel Bars PELLINI, W. S. and DUNPHY, R. P.—Solidification of Gray Iron in Sand Molds PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 429. PELLINI, W. S., BISHOP, H. F. and BRANDT, F. A.—Solidification of Steel Against Sand and Chill Walls 430. PETTERSSON, H.—An investigation of the Penetration of Steel into Molding Sand 731. REHIDER, J. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 732. REHIDER, J. E.—Effect of Phospherus Content on Mechanical Properties of a Nodular Cast Iron 733. RICHARDSON, B., MURION, A. A. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 734. RICHARDSON, B., MURION, A. A. and PAIRFELD, H. H.—Core Oil Evaluation Method 735. ROTE, F. B., SHELLENG, R. D. and Uptimegrove, C.—Melt Quality and Fracture Characteristics on Direct Cooling of Alloyed White Iron 736. RUSSELL, T. W., JR.—Industry Participation in Secondary School Training Pregrams for the Foundry 737. SANAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 737. SCHUNDRIE, A. W. JR. and McQUISTON, C. E.—A New Method of Evaluating Crs's in Jobbing Foundries 738. SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 739. SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 740. STRIETER, F. P. and Nelson, K. E.—A Castability and Properties of Cast Steel as Influenced by Mass and Segregation 740. STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting 740. STRIETER, F. P. and Nelson, K. E.—A Castability and Properties of Cast Steel as Influenced by Ma	
Pellini, W. S., Bishop, H. F. and Myskowski, E. T.—The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars 17. Pellini, W. S. and Dunphy, R. P.—Solidification of Gray Iron in Sand Molds 18. Pellini, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel Against Sand and Chill Walls 18. Pellini, W. S., Bishop, H. F. and Brand, F. A.—Solidification of Steel into Molding Sand 18. Raring, F. W. and Kryntisky, A. O.—Sieve Analyses of Silica Sauds 19. Raring, J. E.—Effect of Chromium and Silicm Contents on Rate of Anneal of Black-Heart Malleable Iron 19. Reinder, J. E.—Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron 19. Reinder, J. E.—Bisci Cupola Melting and its Possibilities 19. Richlard Cast Iron 19. Richlard Cast Iron 19. Richlard Cast Iron 19. Royal Cast Iron 10. Ro	
Steel Ba's PELLINI, W. S. and DUNPHY, R. P.—Solidification of Gray Iron in Sand Molds PELLINI, W. S. and DUNPHY, R. P.—Solidification of Gray Iron in Sand Molds PETTERSSON, H.—An investigation of the Penetration of Steel Against Sand and Chill Walls ARBING, F. W. and KRYMTSKY, A. O.—Solieve Analyses of Silica Sauds RARING, F. W. and KRYMTSKY, A. O.—Solieve Analyses of Silica Sauds REIDER, J. E.—Effect of Phospherus Content on Mechanical Properties of a Nodular Cast Iron PENSITAW, E. S.—Basic Cupola Melting and its Possibilities RICE, L. P., WHITE, W. H. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron RICHARDSON, B., MUNION, A. A. and Farrfild, H. H.—Core Oil Evaluation Method ROTE, F. B., SHELLENG, R. D. and Uptthegrove, C.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze RUSSELL, T. W., JR.—Industry Participation in Secondary School Training Programs for the Foundry White Iron RUSSELL, T. W., JR.—Industry Participation in Secondary School Training Programs for the Foundry Schwage, J. H., WALLACE, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schward, J. H., WALLACE, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schward, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Schward, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Schward, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Striller, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Strill Effects and Usefulness of Gases in Metallurgy Striller, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Strill Effects and Usefulness of Gases in Metallurgy Strill Effects and Their Various Uses Uttherapove, C., Sheller, R. D. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-	PELLINI, W. S., BISHOP, H. F. and MYSKOWSKI, E. TThe Contribution of Riser and Casting End Effects to Soundness of Cast
Pellin, W. S., Bishop, H. F. and Brandt, F. A.—Solidification of Steel Against Sand and Chill Walls Pettersson, H.—An investigation of the Penetration of Steel into Molding Sand Raring, F. W. and Kryntsky, A. O.—Sieve Analyses of Silica Sauds Rehder, J. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron Rehder, J. E.—Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron Petershaw, E. S.—Basic Cupola Melting and its Possibilities Rice, L. P., White, W. H. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron Richardson, B., Murton, A. A. and Fairfeld, H. H.—Core Oil Evaluation Method Rote, F. B., Shelleng, R. D. and Upthersone, C.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Rote, F. B., Concer, G. A. and Delonge, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron Russell, T. W., Jr.—Industry Participation in Secondary School Training Programs for the Foundry Savage, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schingele, R. W.—Development of Foundry Courses in High and Trade Schools Schingele, R. W.—Development of Foundry Courses in High and Trade Schools Schinary, H. A. and Book, W. K.—Freezing of White Cast Iron in Green Sand Molds Shelleng, R. D., Upthersone, C. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Smith, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Striefets, R. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys Striefets, E.—Ill Effects and Usefulness of Gases in Metallurgy Striefers, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys Striefers, E.—Ill Effects and Usefulness of Gases in Metallurgy Triefersove, C., Shelleng, R. D. a	
Pettersson, H.—An investigation of the Penetration of Steel into Molding Sand Raring, F. W. and Kryntsky, A. O.—Sieve Analyses of Silica Sauds Raring, F. W. and Kryntsky, A. O.—Sieve Analyses of Silica Sauds Reimer, J. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 24. Reimer, J. E.—Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron Rice, L. P., White, W. H. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron Richardon, B., Murdon, A. A. and Fairffeld, H. H.—Core Oil Evaluation Method 27. Rother, F. B., Shelleng, R. D. and Uptherried, H. H.—Core Oil Evaluation Method 28. Rother, F. B., Shelleng, R. D. and Uptherried, K. A.—Isothermal Transformation Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Rosel, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 29. Rother, R. W.—Development of Foundry Courses in High and Trade Schools Schiware, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Shelleng, R. D., Uptherrove, C. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Shith, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Strutk, K. M.—Dimensional Checking and Pressure Testing o	
RARING, F. W. and KRYNITSKY, A. O.—Sieve Analyses of Silica Sauds REHDER, J. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 244 REHDER, J. E.—Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron 546 RENSHAW, F. S.—Basic Cupola Melting and its Possibilities RICE, L. P., WHITE, W. H. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 547 RICHARDSON, B., MURTON, A. A. and FAIRFELD, H. H.—Core Oil Evaluation Method 758 RICHARDSON, B., MURTON, A. A. and FAIRFELD, H. H.—Core Oil Evaluation Method 758 RICHARDSON, B., MURTON, A. A. and DELONGE, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron 759 RUSSELL, T. W., JR.—Industry Participation in Secondary School Training Programs for the Foundry 750 SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 750 SCHINEBLE, A. W. Jr. and McQuiston, C. E.—A New Method of Evaluating Crss in Jobbing Foundries 750 SCHINARIZ, H. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds 751 SCHUMANY, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 752 SCHUMANY, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 753 SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 88-	
REHDER, J. E.—The Relative Effects of Chromium and Silican Contents on Rate of Anneal of Black-Heart Malleable Iron 24. Rehder, J. E.—Effect of Phosphcrus Content on Mechanical Properties of a Nodular Cast Iron 82. Rice, L. P., White, W. H. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 83. Richardson, B., Murton, A. A. and Fairfeld, H. H.—Core Oil Evaluation Method 84. Rote, F. B., Shellerg, R. D. and Upthegrove, C.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze 85. Rote, F. B., Conger, G. A. and Delonge, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron 86. Russell, T. W., Jr.—Industry Participation in Secondary School Training Programs for the Foundry 87. Savage, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 88. Schuman, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 88. Schuman, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 88. Schuman, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 88. Schuman, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 88. Schuman, L. N.—Dimensional Checking and Pressure Testing of Gray Iron Castings 88. Schuman, L. N.—Dimensional Checking and Pressure Testing of Gray Iron Castings 89. Street, E. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting 89. Street, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting 89. Street, F. P. and Nelson, K. E.—A Castability and Properties of Cast Steel as Influenced by Mass and Segregation 89. Street, F. P. and Nelson, K. E.—A Castability and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze 89. Street, F.	
Rehber, J. E.—Effect of Phospherus Content on Mechanical Properties of a Nodular Cast Iron Renshaw, E. S.—Basic Cupola Melting and its Possibilities Rice, L. P., White, W. H. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron Richardson, B., Murton, A. A. and Fairfield, H. H.—Core Oil Evaluation Method Rote, B., Shellen, R. D. and Upthersone, C.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Rote, F. B., Concer, G. A. and Delonge, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron Russell, T. W., Jr.—Industry Participation in Secondary School Training Pregrams for the Foundry Schaege, J. H., Wallace, J. F. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Schneble, A. W.—Br. and McQuiston, C. E.—A New Method of Evaluating Crass in Jobbing Foundries Schhorder, R. W.—Development of Foundry Courses in High and Trade Schools Schhorder, R. W.—Development of Foundry Courses in High and Trade Schools Schhorder, R. D., Upthersone, C. and Mote, E. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Smith, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Striketer, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys Strott, A. and Davenport, W. F.—Precoat Materials for Investment Casting Taylor, H. F., Wallace, J. F. and Savage, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Strott, A. and Davenport, W. F.—Precoat Materials for Investment Casting Taylor, C., Shelleng, R. D. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Wallace, J. F., Savage, J. H. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Strott, A. and Brass and Their Various Uses Wallace, J. F., Savage, J. H. and Taylor, H. F.—Mechanical Properties of Cas	
RENSHAW, F. S.—Basic Cupola Melting and its Possibilities RICE, L. P., WHITE, W. H. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron RICHARDSON, B., MURTON, A. A. and FARRFEELD, H. H.—Core Oil Evaluation Method ROTE, F. B., SHELLENG, R. D. and Upthegrove, C.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze ROTE, F. B., CONGER, G. A. and DELONGE, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron RUSSELL, T. W., Jr.—Industry Participation in Secondary School Training Pregrams for the Foundry SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation SCHRODER, R. W.—Development of Foundry Courses in High and Trade Schools SCHROMER, R. W.—Development and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMAN, I. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMAN, I. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., Upthegroup, C. and Rotte, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze STROIT, A. and DAVENDORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 224 WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 325	
RICE, L. P., WHITE, W. H., and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron	
of Nodular Cast Iron RICHARDSON, B., MURTON, A. A. and FAIRFIELD, H. H.—Core Oil Evaluation Method 276 ROTE, F. B., SHELLENG, R. D. and Upthegrove, C.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze ROTE, F. B., CONGER, G. A. and DELONGE, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron RUSSELL, T. W., JR.—Industry Participation in Secondary School Training Programs for the Foundry 376 SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation SCHNEBLE, A. W. JR. and McQuiston, C. E.—A New Method of Evaluating Crs's in Jobbing Foundries SCHOEDER, R. W.—Development of Foundry Courses in High and Trade Schools SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMARIZ, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., Upthegrove, C. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze 67 SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings STROTT, A. and Davenport, W. F.—Precoat Materials for Investment Casting STROTT, A. and Davenport, W. F.—Precoat Materials for Investment Casting Alloys STROTT, A. and Davenport, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 TAYLOR, H. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 837 WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 837	
RICHARDSON, B., MURTON, A. A. and FAIRFFELD, H. H.—Core Oil Evaluation Method ROTE, F. B., SHELLENG, R. D. and UPTHEGROVE, C.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze ROTE, F. B., CONGER, G. A. and DELONGE, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron SUSSELL, T. W., JR.—Industry Participation in Secondary School Training Pregrams for the Foundry SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation SCHORDER, R. W.—Development of Foundry Courses in High and Trade Schools SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHWARTZ, H. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze STRIFTER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
ROTE, F. B., SHELLENG, R. D. and UPTHEGROVE, C.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze ROTE, F. B., CONGER, G. A. and DELONGE, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron. 509 RUSSELL, T. W., Jr.—Industry Participation in Secondary School Training Pregrams for the Foundry 376 SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 SCHROEDER, R. W.—Development of Foundry Courses in High and Trade Schools 332 SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 418 SCHWARTZ, H. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds 5. 48 SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze 56 SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings 57 STRETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys 57 STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting 58 TRYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 TAYLOR, J.—"Spec" Sheets and Their Various Uses 57 UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze 67 WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron 58	RICHARDSON, B., MURTON, A. A. and FAIRFIELD, H. H.—Core Oil Evaluation Method
ROTE, F. B., CONGER, G. A. and DeLonge, K. A.—Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron	
White Iron Russell, T. W., Jr.—Industry Participation in Secondary School Training Pregrams for the Foundry 376 SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 SCHNEBLE, A. W. Jr. and McQuiston, C. E.—A New Method of Evaluating Crs's in Jobbing Foundries 223 SCHUMAN, L. N.—Equipment of Foundry Courses in High and Trade Schools 329 SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 418 SCHWARTZ, H. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., Upthegrove, C. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze STRITT, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings 304 STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROIT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 136 TAYLOR, J.—"Spec" Sheets and Their Various Uses 137 Upthegrove, C., Shelleng, R. D. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 WHITE, W. H., RICE, L. P. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	Bronze
RUSSELL, T. W., JR.—Industry Participation in Secondary School Training Pregrams for the Foundry SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation SCHNEBLE, A. W. JR. and McQUISTON, C. E.—A New Method of Evaluating Costs in Jobbing Foundries SCHOOLER, R. W.—Development of Foundry Courses in High and Trade Schools SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHWARTZ, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5-5 Red Brass and 88-8-4 Bronze SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron S787	
SAVAGE, J. H., WALLACE, J. F. and TAYLOR, H. FMechanical Properties of Cast Steel as Influenced by Mass and Segregation SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries 332 SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries 332 SCHNEBLE, A. W. JR. and McQuiston, C. EA New Method of Evaluating Costs in Jobbing Foundries 348 SCHNEBLE, A. W. JR. and Book, W. KFreezing of White Cast Iron in Green Sand Molds 488 SCHNEBLE, A. NEquipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 488 SHELLENG, R. D. Utthecrove, C. and Rote, F. BMelt Quality and Fracture Characteristics of 85-5-5 Red Brass and Casting Alloys STRICTER, F. P. and Nelson, K. EA Castability and Properties of Cast Steel as Influenced by Mass and Segregation 223 TAYLOR, J"Spec" Sheets and Their Various Uses 135 Utthecrove, C., Shelleng, R. D. and Rote, F. BMelt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze White, W. H., Rice, L. P. and Elsea, A. RInfluence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
SCHNEBLE, A. W. JR. and McQuiston, C. E.—A New Method of Evaluating Costs in Jobbing Foundries Schroeder, R. W.—Development of Foundry Courses in High and Trade Schools Schwarz, R. W.—Development and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings Schwarz, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds Shelleng, R. D., Upthegrove, C. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Smith, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings Spire, E.—Ill Effects and Usefulness of Gases in Metallurgy Strieter, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys Strott, A. and Davenport, W. F.—Precoat Materials for Investment Casting Taylor, H. F., Wallace, J. F. and Savage, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation Taylor, J.—"Spec" Sheets and Their Various Uses Upthegrove, C., Shelleng, R. D. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze Wallace, J. F., Savage, J. H. and Taylor, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 White, W. H., Rice, L. P. and Elsea, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
SCHROEDER, R. W.—Development of Foundry Courses in High and Trade Schools SCHMANN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings 418 SCHWARTZ, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings STRIETER, F. P. and Velson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 223 WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
SCHUMAN, L. N.—Equipment and Methods of Straightening and Dimensional Inspection of Malleable Iron Castings SCHUMARTZ, H. A. and Bock, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings STRIETER, F. P. and Usefulness of Gases in Metallurgy STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses Upthegrove, C., Shelleng, R. D. and Rote, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and Elsea, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron SCHUMAN, L. N.—Equipment and Methods of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
SCHWARTZ, H. A. and BOCK, W. K.—Freezing of White Cast Iron in Green Sand Molds SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings STRIETER, F. P. and Usefulness of Gases in Metallurgy STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and Davenport, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 1372 L'PTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation 1283 WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 1397	
SHELLENG, R. D., UPTHEGROVE, C. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings SPIRE, E.—III Effects and Usefulness of Gases in Metallurgy STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses 1372 UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and Elsea, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
Bronze SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings SPIRE, E.—Ill Effects and Usefulness of Gases in Metallurgy STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron 337	
SMITH, K. M.—Dimensional Checking and Pressure Testing of Gray Iron Castings	
STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and Elsea, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron	SMITH, K. MDimensional Checking and Pressure Testing of Gray Iron Castings
STRIETER, F. P. and Nelson, K. E.—A Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting Alloys STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and Elsea, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron	SPIRE, EIII Effects and Usefulness of Gases in Metallurgy
STROTT, A. and DAVENPORT, W. F.—Precoat Materials for Investment Casting TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron 315	STRIETER, F. P. and Nelson, K. EA Castability and Property Comparison of Several Magnesium-Rare Earth Sand Casting
TAYLOR, H. F., WALLACE, J. F. and SAVAGE, J. H.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation TAYLOR, J.—"Spec" Sheets and Their Various Uses UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. B.—Melt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Centent on Mechanical and High-Temperature Properties of Nodular Cast Iron S37	Alloys
TAYLOR, J.—"Spec" Sheets and Their Various Uses	
UPTHEGROVE, C., SHELLENG, R. D. and ROTE, F. BMelt Quality and Fracture Characteristics of 85-5-5 Red Brass and 88-8-4 Bronze	
Bronze	
WALLACE, J. F., SAVAGE, J. H. and TAYLOR, H. F.—Mechanical Properties of Cast Steel as Influenced by Mass and Segregation . WHITE, W. H., RICE, L. P. and ELSEA, A. R.—Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron	
WHITE, W. H., RICE, L. P. and ELSEA, A. RInfluence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron	
of Nodular Cast Iron	
ZEDER, J. CManagement of Industrial Research	ZEDER, J. CManagement of Industrial Research
ZIEGLER, N. A., MEINHART, W. L. and GOLDSMITH, J. REffect of Vanadium on the Properties of Cast Chromium-Molybdenum Steels	ZIEGLER, N. A., MEINHART, W. L. and GOLDSMITH, J. REffect of Vanadium on the Properties of Cast Chromium-Molybdenum
ZIMNAWODA, H. WMechanical Equipment for Medium-Size Gray Iron Foundry	